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Magnesium in Synthesis of Porous and Biofunctionalized Metallic Materials

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Abstract

Magnesium particles are used in metallurgic routes, where it can be total or partially evaporated creating pores for ingrowth bone tissue. This book chapter contains the latest findings on the microstructural physical and mechanical properties of β -Ti alloys with Mg additions designed and obtained by the authors. As well as the main new techniques used to fabricate Ti-Mg alloys. An especial emphasis on the microstructure-properties relationship was made to assist on the guide for future efforts of the scientific community towards developing more efficient biomaterials. The β % were related to the low elastic modulus which were in the range of 31–49 GPa close to cortical bone and hardness close to commercial Ti grade 2. The compressive strength was greater than the value of cortical bone. Pore size were in the range of 5–100 μm depending on the sintering temperature, with higher wettability the samples with more porosity. These findings were promising to application of β titanium alloys containing Mg for orthopedic application.

Keywords: oxide magnesium, biological properties, titanium surface, porous metals

1. Introduction

Surgical treatments of bone injuries patients in emergency departments worldwide each year due to involvement in rigorous athletic activities, social instability, traffic accidents, and prolonged human lifespan [1].

Bone defects, mainly induced by traumatic avulsions, sequelae of infection-induced bone sequestration, congenital malformations, or neoplastic resections, confront us with an extreme challenge for reconstructive surgery the need to induce bone regeneration to repair structural bone deficient [2] has inspired research on and development of a vast number of bone repair materials.

Diverse metallic materials are already established as biomaterials due to their high biocompatibility, low toxicity, and good strength–ductility relationship. Examples of these alloys are stainless steel (especially 316 L), cobalt and chromium (CoCr) alloys, and titanium (Ti) alloys [3, 4]. However, the low toxicity and mechanical properties of Ti alloys, specifically the elastic modulus, are more adequate for biomedical uses.

From Ti alloys, the most common for dental and orthopedic applications are materials formed by Ti, aluminum (Al) and vanadium (V) like Ti-6Al-4 V and other β -phase-type alloys as the ones with high contents of β -stabilizers (V, Cr, molybdenum (Mo), Fe, niobium (Nb), and tantalum (Ta)) [5–12]. However, several reports point to the V in the Ti-6Al-4 V as toxic [13, 14], being a motivation for exploring further V-free options. Moreover, the β -phase type Ti alloys have a good combination of mechanical properties and biocompatibility. Besides, the β -Ti alloys have a lower elastic modulus compared to other Ti alloys [15]. Considering that the elastic modulus is a key factor for the success or failure of the implant, this is a remarkable characteristic of these alloys [1]. However, the reported elastic modulus for β -Ti alloys ranges from 69 to 110 GPa [15, 16], being still far from that of human bone (lower than 30 GPa) [17].

To overcome this drawback, several Ti alloys are being developed and most of them are showing promising results in the matter of mechanical properties. A number of these metallic systems are being obtained through powder metallurgy methods to obtain functional porous structures. It has been widely reported that the porous surfaces assist on the fixations and ingrowth of organic tissue, improve the body fluid, reduce the mechanical mismatch due to lower elastic modulus values, and reduce the failure rate of implants [3]. Examples of the above are Ti and indium (In) as (Ti-In) [18], Ti-Mo [7–9], Ti, Nb and tin (Sn) as Ti-Nb-Sn [10, 19], Ti and zirconium (Zr) as Ti-Zr [20], and Ti and silver (Ag) as Ti-Ag [21] alloys. However, some of the previous systems employ alloying elements that are still not widely studied, being a reason why several *in vivo* tests of biocompatibility should be carried out to determine their biological feasibility.

Another route is the design and development of biomaterials based on widely explore elements as magnesium (Mg). This element has multiple advantages for biomaterials as non-toxicity, biocompatibility, biodegradability, increase strength of the bone, and has a low elastic modulus [3, 4, 22–24]. Low concentrations of Mg^{2+} play an important role in cells activity by stimulating the improvement of cell adhesion and extracellular matrix mineralization [25, 26]. Furthermore, Mg is the fourth most abundant element in the human body and is essential in digestion processes [22, 24]. The non-harmful degradation of an Mg, zinc (Zn) and manganese (Mn) as $Mg_{2.5}Zn_{0.2}Mn$ alloy inside the human body has been demonstrated [23]. Based on the above, Mg is a feasible alloying element to boost the biocompatibility and possible control of biodegradability over the time of different biomaterials for medical purposes. The biodegradability of Mg can avoid the need for a second surgical process to remove the implant. The possibility to control such biodegradability is still under intense investigation [23, 27, 28]. Moreover, Mg is a potential alloying element to significantly reduce the elastic modulus. This could reduce the failure rate due to mechanical mismatch between the implant and the bone, and the occurrence of load shielding (absorption of mechanical stress by the implant) [3]. However, one of the main disadvantages of Mg as a biomaterial is that the degradation rate can be faster than the required to allow a complete regeneration of the organic tissue [29]. This is the motivation to explore the use of Mg as an alloying element instead of a matrix. Considering the already explained qualities of Ti biomaterials, it is a good candidate to join with the virtues of Mg.

Until now, few reports on Mg as an alloying element of Ti alloys have been reported [30–33]. Deep research is still needed in the matter of optimizing Mg contents, processing parameters, and designing new systems that reduce the economic and health losses due to the failure of implants. The field of Ti-Mg alloys is emerging and is pointing as highly promising for biomedical purposes.

This book chapter contains the latest findings on the microstructural, mechanical, and biological properties of Ti alloys with Mg additions designed and obtained by the authors. Also, a description the most important techniques to obtain Ti-Mg alloys for biomedical application. An especial emphasis on the microstructure-properties relationship was made to assist on the guide for future efforts of the scientific community towards developing more efficient biomaterials.

2. Chemical and biological properties of magnesium

As the largest dynamic biological tissue in the body, bones are composed of inorganic minerals as magnesium and metabolically active cells surrounded by a large volume of extra cellular matrix, and they form a rigid scaffold that has an irreplaceable role in maintaining life activities, including supporting the body and protecting visceral organs [1]. For bone repair, metallic materials are used to repair or replace the bone tissue damaged. The main materials used in orthopedics include stainless steel and Ti alloys because they are mechanically strong and resistant to fracture. However, there is a potential for the release of metallic ions and/or particles through corrosion and/or wear that trigger inflammatory responses that can reduce biocompatibility and lead to tissue loss [34].

Previously *in vivo* and *in vitro* studies have shown that metallic biomaterials fabricated with Mg exhibit good biocompatibility and free of systemic inflammation reaction or affection of the cellular blood composition. In addition, high mineral apposition rates and increased bone mass were found around degrading Mg implants in bone [35] also, it can present a benefic influence in metallic materials once the bone-implant interface mineralization and osseointegration are significantly greater for metallic materials like titanium and magnesium alloys (Ti-Mg), hence have shown to stimulate new bone formation when used as bone fixture.

Mg is the fourth most plentiful cation in the human body, and is an element essential in many metabolic processes, involved in the regulation of eukaryotic cell proliferation, structural functions are correlates with the enhancement of protein synthesis. Furthermore, is primarily stored in bone tissue, controlling growth of bone cells and accelerates the bone healing [36, 37], which has characteristics of bio-degradability, in the physiological environment can be eliminated and also the corrosion product of Ti-Mg alloys (Mg^{2+}) does not cause unexpected complications because excessive Mg^{2+} are easily eliminated in the urine. Moreover, alloys fabricated using Mg elemental can present mechanical properties similar to those of bone, due to fabrication pores material, decreasing the elastic modulus [38, 39]. Once the bone resorption around stress-shielded are in bone fixation treatments is an important consideration within clinical sectors. Because of its versatility, metallic biomaterials based on Mg can contribute to biological properties and improve the osseointegration process [40].

Mg^{2+} is distributed in three major compartments of the body: ~65% in the mineral phase of bone, 34% in muscle, ~1% in plasma and interstitial fluid [41] and it has a radius of 0.65Å (Mg^{2+}). Usually, at the physiological pH range, Mg^{2+} is hydroxylated with six H_2O molecules with a large hydration energy to form a complex with a large radius of approximately 5Å, also strongly interacts with phosphate ligands from nucleotides such as ATP and DNA due to its high charge/radius ratio [42]. Coordination number and spatial distribution of water molecules surrounding the Mg^{2+} influence its binding thermodynamics with the protein. As it can be associated with water or phosphate complex binds protein with three to five coordinating oxygen plays a major role in the protein-binding ligands.

Its charge density is approximately $(.99/.65)^3$ 3x more than that of ion calcium (Ca^{2+}) and its affinity for electronegative ligands, almost always oxygen in biological systems, is much greater. Further, the Mg-oxygen bond length is approximately 2.05 Å. When Mg elemental is octahedrally coordinated by six oxygen atoms, the oxygen-oxygen distance is $2.05 \times 2^{0.5} = 2.9 \text{ Å}$, an optimal van der Waals contact [43]. In biological systems, Mg ions exist in 3 different states: bound to proteins, complexed to anions, and free (Mg^{2+}). Only free Mg has biological activity [44]. The adult human body contains approximately 24 g (1 mol) of Mg in cells × 280 mg in extracellular fluids and the skeleton represents the body's largest Mg store (approx. 60% of total Mg), divided into two subcompartments. Thus, bone functions as a large Mg reservoir, helping to stabilize its concentration in serum. About one fourth of total Mg^{2+} is located in skeleton muscle [45]. Bone is continuously under load, which can cause bone defects such as fatigue cracks. Such defects can be dealt with by the body's own healing mechanism. Thus, such critical size defects (CSDs) have to be treated with an implant.

3. Most promising techniques to fabricate Ti-Mg alloys

Ti alloys can be fabricated using as alloying element Mg to create porosity, leading to formation of biocompatible scaffolds with lower elastic modulus by metallurgical and additive manufacturing process. Thus, the stiffness of Mg-based implants can be more easily tailored to match that of bone, which reduces the risk of stress shielding, a phenomenon that will be discussed in the next section.

3.1 Fabrication via powder metallurgy

One of the most conventional techniques to create porous metallic materials is by the powder metallurgy. Conventional manufacturing of Ti alloys via powder metallurgy involves: (1) blending of the powder to achieve a uniform particle size distribution and, if needed, mixing of the Ti powder with the required alloying elements; (2) shaping of the powder blend (this can be achieved via different manufacturing methods where the simplest and cheapest is cold uniaxial pressing); and (3) solid-state sintering (i.e. heat treatment at high temperature, below the melting point of Ti) generally performed under vacuum [46]. This technique is cost effective, allows for better control of powder size and introduction of desirable pores. When it comes to pore size and shape, these are related to the size of the starting powder, its shape, its size and the shape of the spacer used to promote porosity. Among all of them, Mg is a good candidate in the manufacture of Ti scaffolds, as its solubility in Ti is low. Furthermore, Mg^{2+} increases osteoconductivity [47] and does not present any type of biomedical inconvenience, such as toxicity.

The main works carried out using Mg are for Ti-Al-V alloys [48]. Wen et al., (2001) reported that Mg foam prepared with a porosity of 50% showed a compressive strength of 2.33 MPa, with Young's modulus of 0.35 MPa, respectively [49]. Zhuang et al. [50] also evaluated the mechanical properties of porous Mg manufactured by powder metallurgy and the scaffolds with porosity from 36 to 55% showed a Young's modulus value in the range of 3.6 to 18.1 GPa, closer to that of natural bone [50]. They also investigated the effect of porosity on biodegradation. In their study, it was reported that materials with greater porosity degraded more quickly, due to greater interconnectivity and surface exposure, conditions that maximize chemical reactions. Although Mg materials have a low elastic modulus, mechanical resistance and corrosion are limiting factors for their use [51]. However, the use of Mg together with

other elements to form porous alloys such as Ti can be an interesting alternative. For the production of porosity using this technique, it is necessary to control parameters such as temperature and times of the sintering steps, in addition to the particle size of the powders, due to its commitment to mechanical properties [52]. Such variables strongly influence the morphology of the pores, that is, it can provide the same amount of porosity, but with different shapes and sizes. The mechanical properties are also affected, mainly those related to the ductility and dynamic properties of the material, as they depend on the porosity characteristics [53]. The pores attenuate the applied force, do not distribute it over a larger area and cause local stress accumulation, so that they can even serve as sites for crack nucleation [54]. The effect of porosity on mechanical properties depends mainly on the following factors: volume fraction of the pores and their interconnection, size, morphology and distribution. The most important parameters are the total porosity, the shape of the pores and/or contacts during sintering [55].

Porosity has a noticeable and well-recognized effect on mechanical properties. Porosity can increase stress concentration and cause fractures. It was demonstrated by Danninger et al. that this parameter is directly related to the mechanical properties of an alloy [56]. These factors can be controlled by adjusting the sintering parameters, compaction pressure and particle size [57]. Optionally, functional porosity can be introduced by adapting the particle size of the starting powders and the sintering conditions. In addition, powder metallurgy allows flexibility in alloy design, mixing pure Mg powders with different elemental or alloy powders. Due to the high affinity of Mg for oxygen, all handling of powders and samples, as well as subsequent sintering, must be carried out under a protective atmosphere of argon or under vacuum [58] conditions, residual pores can vary between 2% and 45%. When approaching porosities close to 45%, interconnection (percolation) of the pores appears.

3.2 Fabrication via additive manufacturing

In recent years, interest has increased in the application of additive manufacturing of Mg alloys for its biomedical application. This can allow the obtaining of complex shapes adapted to the patient, since it would be a personalized manufacture. However, and due to the physical properties of Mg and its alloys, the application of additive manufacturing by melting the alloy has not been easy, since the boiling temperature of magnesium is very low ($\sim 1091^{\circ}\text{C}$). Despite this, and due to the manifest interest in the biodegradation properties of Mg motivate its application as biodegradable implants [59], seeking the best combination between resistance to corrosion, wear, mechanical properties and biocompatibility [60].

Another factor influencing the scarce development of additive techniques in Mg alloys is the ease of obtaining coupling parts by injection casting processes or the extrusion capacity of these alloys despite the difficulty imposed by their hcp crystal-line structure. In addition, the great reactivity with oxygen that Mg presents must be considered and also limits the application of rapid heating techniques that could cause the combustion of the metal. However, some of the technologies applied to other materials have been applied, by means of specialized teams that require work in protective atmospheres, which ensure the possibility of handling these alloys [61].

The potentially most interesting techniques for the manufacture of magnesium alloys are powder bed fusion (PBF), especially selective laser melting (SLM), widely used in the development of different Mg alloys [62, 63]. Powder bed fusion (PBF) is an AM process in which thermal energy is used to selectively fuse regions of a powder

bed [64]. The powder bed contains metal, polymer, or ceramic powder as feedstock. An energy source directed towards the powder bed selectively scans and melts the top layer of the powder bed. The powder bed then lowers and a fresh layer of powder is spread over the melted layer. This process continues until the entire structure has been formed by stacking melted layers of powder.

Another way to use additive manufacturing related to magnesium alloys is to have a porous structure obtained by additive manufacturing as in the case of Perets et al. [65] who obtain the Ti-6Al-4 V mesh by SLM and then infiltrate the Mg elemental into the holes. In this way they can obtain structures that do not collapse, although their resistance is not increased. The magnesium will present an accelerated corrosion by galvanic effect and depending on the size of the pores, an osseointegration will be available as it is a very biocompatible set.

Degradation capability of Mg gives a feature of bioactivity in bone formation that leading Balog et al. develop a bioactive metal system compound by structural material for dental implants, via extrusion from a powder mixture of Ti and Mg (4 and 12%) in low temperature [66]. Adding Mg, is possible to obtain a bioactive system and a decreasing of elastic modulus, further promote a good osteointegration due to the Mg resorption and the presence of pores where the bone ingrowth can be formed.

Mechanical properties in Ti parts that receive infiltration of Mg depends on the amount of Mg and matrix used. Studies published by Jiang et al. about infiltration of Mg in a scaffold of Ti-Mg (99,9%) was possible to control density from compaction pressure with volume fractions up to 60% Ti, which confers stiffness similar to those of cortical bone [67]. The use of non-degradable Ti matrices, as described in previous sections, is necessary as a non-degradable support due to its excellent biocompatibility, high resistance to corrosion and excellent mechanical properties. Similarly, efforts have been made to obtain porous Ti by means of spacer techniques [68] or by additive manufacturing processes such as selective laser melting (SLM) [69, 70]. Biodegradable Mg-based alloys are advantageous as fillers for bioactive implants because the release of Mg ions during corrosion *in vivo* is non-toxic and, furthermore, may have a beneficial effect on tissue regeneration and osteoblast response [71]. However, excessive *in vivo* corrosion of Mg can result in a premature loss of mechanical integrity and it is for this reason that a balance is struck between the structural integrity provided by titanium alloys and the degradation of Mg alloys. The aim is therefore to achieve synergy between both alloys and therefore to improve or control the biodegradation of magnesium through alloys that allow its corrosion rate to be controlled and can adapt it to the rate of bone growth, as proposed by Perets et al. with Mg-2.4% Nd-0.6% Y-0.3% Zr [65] alloy infiltration working in simulated environments.

4. Microstructural features of Ti-Mg alloys

The Ti-alloys are already recognized as the most promising materials for dental and orthopedic implants. This is due to their excellent biocompatibility, good mechanical properties, superior corrosion resistance, and no allergic issues [72]. Adding Mg provides multiple biological advantages to Ti biomaterials. Some of these advantages are an elastic modulus closer to that of human bone, high biocompatibility, and low toxicity. The mismatch between elastic modules of human bone and dense metallic biomaterials is considered the main cause of implant failure. Typically, the elastic modulus of human bone has a maximum value of 30 GPa, while that for metallic biomaterials ranges from 100 GPa for chemically pure Ti to

230 GPa for Co-Cr alloy [15]. From the metallic biomaterials, the near β -Ti alloys have reported some of the lowest elastic modulus values ranging from 80 to 110 GPa [14]. Additionally, porous β -Ti alloys have obtained elastic modulus ranging from 43 to 75 GPa [4, 37, 73, 74], and it is expected that the addition of Mg will decrease even more such values [75]. An elastic modulus near to that of human bone improves the performance of multiple biomaterials used for dental and orthopedic implants. Considering that the microstructural effects of Mg additions are important to define the mechanical behavior of biomaterials, it is important to study the microstructure.

Due to the above, the effect of low (3 mass%) Mg additions into a Ti-34Nb-6Sn alloy was studied [28]. The selection of Nb and Sn was done for its β -stabilizing effect on Ti alloys [9, 76]. The β -phase has shown better mechanical compatibility with the human bone in comparison to the α -phase. This is due to lower elastic modulus values and a high strength-to-weight ratio [13, 14]. An elastic modulus of the implant that is close to that of the human bone decreases the mismatch of mechanical stress through the interface and avoids the damage of the organic tissue cells. Based on the above, the closer the elastic modulus between both, bone and implant material, the lower the probability of crack nucleation and failure of the implant [72]. Furthermore, a high strength-to-weight ratio allows reducing the thickness of biomedical metallic implants. Besides, the selected contents of Nb and Sn also contribute to obtaining elastic modulus of the Ti alloy of ~60 GPa [9, 74, 77]. The addition of 6 wt.% of Sn into the Ti-Nb alloy showed a good combination of corrosion resistance, strength, hardness, and lower elastic modulus [4]. On the other hand, considering the low elastic modulus of Mg (from 39 to 46 GPa) [78], it was added to reduce the elastic modulus of the Ti alloy. Besides, Mg is a natural component of human bones and is a required element for the metabolism process, i.e., Mg exhibits great biocompatibility, non-toxicity, and can stimulate hard tissue recovery [2]. The reported biodegradability of Mg^{2+} is one more of its valuable advantages [79]. The possibilities of its use in dental and orthopedic implants can be highly beneficial from controlling its degradation rate and ensuring its mechanical integrity during desire clinical periods. Moreover, the corrosion products of Mg, trigger the osteoconductivity of the bone [75]. For the last, the β -phase Ti alloys have shown a superior electrochemical performance that provides better resistance in corrosive environments as oral or body fluids [13, 32]. This is due to the surface TiO_2 passive film that inherently protects these alloys [10, 32].

From the above-mentioned selection of components, four Ti alloys were prepared by powder metallurgy method. A typical four-stage route of milling – mixing – compaction – sintering was used. For this, measured amounts of titanium hydride (TiH), niobium hydride (NbH), and atomized Sn were used to obtain a mass ratio of Ti, Nb, and Sn of 60:34:6. The powders were mixed in a planetary ball mill and grounded at 200 rpm for 40 min. Posteriorly, the mixed powders were dried under a vacuum. Details of processing parameters can be found in previous work [28]. Half of the mixed powder was saved with the abovementioned chemical composition, while the other half was mixed with Mg powder in a 3 mass%. All the dried powders, whether with or without Mg, were compacted at 200 MPa for 15 s. To evaluate the microstructural and mechanical effect of Mg addition, two different sintering temperatures were used, 900°C and 1100°C. The sintering was carried out in a high vacuum resistive furnace for 2 h. A scheme of the elaboration process is shown in **Figure 1**.

Finally, two Ti-34Nb-6Sn (TNS) and two Ti-34Nb-6Sn/Mg (TNS/M) alloys were produced. The identification of the obtained alloys indicates the sintering temperature as postfix: TNS₉₀₀, TNS₁₁₀₀, TNS/M₉₀₀, and TNS/M₁₁₀₀.

Figure 1. Representation of the methodology to elaborate the Ti-34Nb-6Sn and Ti-34Nb-6Sn/Mg alloys [28].

For microstructural analysis, the samples were subjected to conventional metallographic preparation until a mirror-like surface. Final polishing with oxide polishing suspension (OPS) solution and hydrogen peroxide (10:2) was applied. Rietveld refinement of X-ray diffraction measurements was carried out for quantifying the present phases and estimating the lattice parameters. A Bruker/D2Phaser with Cu-K α radiation was used at 30 kV and 10 mA. The measured 2θ range was 20 and 90° with a step size of 0.02° every 10 s. The Rietveld refinement was carried out by MAUD software (version 2.94) [80]. The morphology and chemical distribution of phases were studied by field emission scanning electron microscopy (FESEM) (ZEISS-ULTRA 55) and an energy-dispersive X-ray spectroscopy detector (EDS) (Oxford Instruments Ltda.).

Considering that the mechanical properties play a determining role in the performance of biomaterials, the elastic modulus was estimated by impulse excitation technique (ATCP, Sonelastic®). Hardness measurements were obtained using a load of 147 N by the Rockwell method (BECLA), using a spherical steel indenter with a diameter of 0.16 cm. More details of the whole methodology can be found in previous works [28–31].

As result, the four alloys resulted in tri-phasic microstructures of α -Ti (under hexagonal compact (hcp), structure), β -Ti (under body centered cubic (bcc) structure), and segregation of Nb. The microstructures can be observed in **Figure 2**, where the light gray color corresponds to the matrix of β -phase, the dark gray to the α -phase, and the bright particles to the Nb segregation. As it was expected from the increment of temperature, bigger grain sizes can be observed in the samples sintered at 1100°C (**Figure 2c** and **d**) compared to those sintered at 900°C. Both, α and β -phases, are randomly distributed in the microstructure, however, the linear chemical composition through the microstructure is not homogeneous. This is due to lower contents of Nb and Sn, especially Nb, in the α -phase (**Figure 2e**). Due to the well-known β -stabilizer nature of both alloying elements [1, 3–10], those chemical gradients were expected. On the other hand, the Nb segregation occurrence was reduced with the sintering temperature (**Figure 2c** and **d**), which indicates a better Nb diffusion in the matrix when temperature increases. However, the Nb particles are continuously observed at both sintered temperatures. From the Ti-Nb diagram phase, Nb has low solubility in Ti, so the continuous presence of Nb segregates was expected [3].

The phases percentages estimated by Rietveld refinement from XRD measurements and the total porosity obtained by Archimedes method for the four alloys are presented in **Table 1**. The TNS₉₀₀, TNS₁₁₀₀, and TNS/M₁₁₀₀ samples showed similar phases percentages. However, the TNS/M₉₀₀ showed a reduced β to α transformation during sintering. Additionally, both TNS/M samples obtained higher porosity

Figure 2. Microstructure of a) TNS₉₀₀, b) TNS/M₉₀₀, c) TNS₁₁₀₀, and d) TNS/M₁₁₀₀, as well as e) linear chemical gradients through α and β phases representative of the four studied samples. Adapted from [28].

Sample	α (mass%)	β (mass%)	Total porosity (%)
TNS ₉₀₀	20.1 ± 0.3	79.9 ± 0.0	22 ± 1
TNS ₁₁₀₀	23.8 ± 0.5	76.1 ± 0.0	11 ± 1
TNS/M ₉₀₀	35.3 ± 0.0	64.7 ± 0.0	29 ± 1
TNS/M ₁₁₀₀	22.4 ± 0.0	77.6 ± 1.4	20 ± 0.5

Table 1. Phases percentages and total porosity of TNS and TNS/M samples sintered at 900°C and 1100°C.

percentages in comparison with the Mg-free alloys. This was a clear suggestion about the reduction of diffusional processes when Mg is added into the Ti-Nb-Sn system. Thus, Mg addition has an apparent α -phase stabilization effect.

Furthermore, the increment of porosity with the Mg content could be related to the highest content of oxygen from the intrinsic passivation layer of Mg^{2+} . When temperatures increase, the release of gas also increases, generating pores at the microstructure [81]. Additionally, the Mg powders acted as a spacer in the TNS/M samples. Comparing the low melting point of Mg (~650°C) with that of Ti (~1668°C), it is evident that a fraction of Mg is evaporated during sintering, while the Ti content remains constant. The partial evaporation of Mg assisted in the formation of the pores during sintering. It is well known that the porosity tends to decrease at higher sintering temperatures during powder metallurgy methods [82]. Thus, the reduced porosity for the samples sintered at 1100°C compared to the samples sintered at 900°C was an expected result. The porosity could influence the mechanical properties of the alloy, especially in the strength and elastic modulus. The mechanical properties will be discussed in the next Section 3.

Posteriorly, the previous Ti-34Nb-6Sn and Ti-34Nb-6Sn/Mg alloys were also reported through the same powder metallurgy methodology, except for sintering temperatures of 700°C and 800°C [29–31]. These samples will be identified as TNS₇₀₀, TNS/M₇₀₀, TNS₈₀₀, and TNS/M₈₀₀ for this book chapter. Compared to the previous TNS/M₉₀₀, and TNS/M₁₁₀₀, the same three constitutive phases, α , β and Nb segregation, were observed at the TNS and TNS/M alloys sintered at 700°C and 800°C. For

comparison purposes, representative EDS measurements of TNS/M₈₀₀ and TNS/M₉₀₀ are shown in **Figure 3**. **Figure 3a** is representative of the distribution of the elements in the samples sintered at 700°C and 800°C, while **Figure 3b** represents the distribution of the elements in samples sintered at 900°C and 1100°C. Similar distributions of the alloying elements were observed in both cases, except for the Nb segregates. **Figure 3a** shows a greater presence of Nb segregates in comparison with that representative of sintering above 900°C (**Figure 3b**). This can be explained by the lower solubility of Nb in Ti below 1100 K (~827°C) [3]. The solubility, together with the effect of temperature, could also be related to the smaller grain size in the samples sintered below 900°C (**Figures 2 and 3**). While the β -phase matrix of the TNS₁₁₀₀ has an average grain size of 15 μm , that of the TNS₇₀₀ has an average of 6 μm . This could be due to the lower solubility of Nb at lower sintering temperatures generated more segregated particles through the microstructure. Those particles could act as a pin for grain growth. The pin-like behavior of Nb segregates in a Ti matrix has been reported before [16].

From **Table 2**, the TNS₇₀₀, TNS/M₇₀₀, TNS₈₀₀, and TNS/M₈₀₀ samples presented higher porosity percentages in comparison with the samples sintered at 900°C and 1100°C (**Table 1**). Besides, the samples sintered at 800°C were more compacted than the samples sintered at 700°C. This was a confirmation about the higher the sintering temperature, the lower the porosity percentage. It is well-known that increasing the sintering temperature favors the diffusional processes and more compacted microstructures with higher relative densities, i.e., lower porosity percentages [82]. Besides, **Table 1** and **Table 2** showed an increment of porosity with the Mg additions for samples sintered at the same temperature. As it was explained before, the partial evaporation of Mg and its passivating oxide, contributed to the increment of porosity. Among the most studied spacers for powder metallurgy methods are carbamide, sodium chloride, ammonium hydrogen carbonate, and Mg [83]. From those, Mg has shown superior advantages over the organic spacers due to its good biocompatibility

Figure 3. Comparison between alloying elements distribution of the Ti-34Nb-6Sn/Mg alloy representative of a) the sintered at 700°C and 800°C and b) the sintered at 900°C and 1100°C. adapted from [28, 30].

Sample	α (mass%)	β (mass%)	Total porosity (%)
TNS ₇₀₀	45 ± 0.3	55 ± 0.3	23 ± 0.5
TNS ₈₀₀	31 ± 0.2	69 ± 0.2	21 ± 1.4
TNS/M ₇₀₀	41 ± 0.1	59 ± 0.1	38 ± 0.6
TNS/M ₈₀₀	33 ± 0.3	67 ± 0.3	28 ± 0.3

Table 2.
Phases percentages and total porosity of TNS and TNS/M samples sintered at 700 °C and 800 °C.

and good mechanical properties [79]. The increment of pore formation with the Mg additions could be beneficial for the implant by decreasing the elastic modulus. This topic will be covered in Section 3.

Besides, from **Figure 3**, a larger pore size for the samples sintered below 800°C is notable in comparison with that for the samples sintered above 900°C. Macropores of ~100 µm in average were acquired (**Figure 3a**) for the samples sintered below 800°C. This was contrasting with the pores in the range from 5 to 35 µm obtained in the samples sintered above 900°C. This could be related to lower atomic diffusion resulting from the lower sintering temperature that reduced the bonding ratio in the samples. Porosity improves the bonding between bone and implant material, encouraging the anchorage and growth of the organic tissue [15, 75]. Thus, porous materials facilitate tissue generation enabling body fluid transmission [84]. Considering that allowing successful osseointegration is one of the main requirements of dental and orthopedic implant materials, the porous structures are highly promising for those applications. It has also been reported that macro-pores are beneficial for multiple biological processes as cell attachment, ingrowth of osteoblasts, vascularization, and osteoconductivity [85]. However, an adequate vascularization requires pores with diameters larger than 100 µm, specifically in the range from 100 to 500 µm [15, 86]. From the above and the fact that increments of porosity percentage used to assist on the decrement of elastic modulus [75], the alloys sintered at temperatures below 800°C could be more appropriate for dental or orthopedic implant applications. However, other mechanical properties as strength and hardness are also crucial for the performance of metallic implants. The mechanical behavior will be described in the following Section 5. This will contribute to clarifying the current concerns about the role of porosity *in-vivo* environments.

5. Mechanical behavior of Ti-Mg alloys

It is well known that the microstructural features have a strong effect on the mechanical behavior of metallic components. In this section, the mechanical performance of the Ti-34Nb-6Sn and Ti-34Nb-6Sn/Mg alloys will be explained based on their microstructural features described in Section 4.

As it was explained before, the Mg additions into dental or orthopedic biomaterials should assist the adequate biocompatibility and a good mechanical strength-elastic modulus relationship. Lower elastic modulus is expected with Mg additions. Moreover, lower elastic moduli are expected from higher porosity percentages, this is, for lower sintering temperatures. Simultaneously, higher hardness values can be expected from the more compacted samples, i.e., the less porous. To evaluate these hypotheses, hardness and elastic modulus measurements were carried out.

From previous works [28–30], **Table 3** presents the Vickers hardness as a function of sintering temperature for the TNS and TNS/M samples. A general tendency to increase hardness as a function of sintering temperature was observed. This is congruent with the lower total porosity percentage in the samples sintered at higher temperatures. As it was explained in Section 2, the temperature increment fabrication atomic diffusional processes that result in bonding improvement and density increment. To observe the effect of temperature on the total porosity and hardness of the sintered alloys, **Figure 4** compares those three values. A general decrement of porosity with the increment of temperature can be observed in the TNS (**Figure 4a**) and TNS/M (**Figure 4b**) samples. Simultaneously, the higher density of the samples resulted in higher hardness values for all the samples.

For comparison purposes, the hardness of elemental Ti ranges from 1.3 to 2.0 GPa [16], while the hardness in the TNS and TNS/M samples ranges from 0.9 to 4.0 GPa. Besides, the minimum recommended hardness value for metallic biomaterials is 1.2 GPa for avoiding high wear damage susceptibility during chewing and daily oral processes [87]. However, the hardness of the natural human teeth ranges from 2.2 to 3.9 GPa [88]. Hardness values near to the ones of natural teeth could assist in decreasing wearing between teeth and implant. From the sintered TNS and TNS/M alloys, the TNS₇₀₀, TNS₈₀₀, TNS/M₇₀₀, and TNS/M₈₀₀ have hardness values below 1.5 GPa, this is, below the minimum acceptable for avoiding wear damage and being within the range of the hardness of natural teeth. As result, those four alloys cannot serve as a feasible biomaterial for dental applications. However, the hardness of the human bone ranges from ~0.3 to ~0.75 GPa [89, 90]. This means that all the sintered samples are within the acceptable hardness range for being applied as orthopedic implants.

From **Figure 4b**, it is also possible to observe a slower decrement rate of total porosity with the sintering temperature in comparison with that for the samples free of Mg additions (**Figure 4a**). This could be related to the abovementioned effect of Mg as a spacer. The partial evaporation of Mg created additional pores compared to the created in the Mg-free samples (TNS). As result, the hardness increment with sintering temperature in Mg-added (TNS/M) alloys also showed a slower rate of increment in comparison with the TNS samples.

As it was discussed in previous Section 4, the elastic modulus plays a key role in the success rate of dental and orthopedic implants. Considering that the elastic modulus is a measure of the stiffness of the material, it determines the resistance to

Sample	Hardness (HV)
TNS ₇₀₀	146 ± 16
TNS ₈₀₀	153 ± 12
TNS ₉₀₀	309 ± 44
TNS ₁₁₀₀	411 ± 25
TNS/M ₇₀₀	92 ± 10
TNS/M ₈₀₀	120 ± 20
TNS/M ₉₀₀	226 ± 85
TNS/M ₁₁₀₀	344 ± 24

Table 3.
Comparison of Vickers hardness (HV) between the TNS and TNS/M samples sintered at 700 °C, 800 °C, 900 °C, and 1100 °C.

Figure 4.
Hardness and total porosity as a function of sintering temperature for the a) TNS and b) TNS/M samples.

deform a material in the elastic range. For increasing the feasibility of the implant, the constituent alloy should have an elastic modulus near to that of the human bone. The elastic modulus of human bone ranges from 5 to 30 GPa [15, 91].

For evaluating the elastic modulus of the sintered alloys, **Figure 5** presents a comparison between the values obtained for the TNS and TNS/M samples sintered at different temperatures. Considering that the samples sintered at 700°C and 800°C were discarded as potential materials for dental implant applications, the samples TNS₇₀₀, TNS₈₀₀, TNS/M₇₀₀, and TNS/M₈₀₀ were not included in **Figure 5**. Lower elastic modulus can be observed in the samples with Mg addition compared to the TNS systems. This result was congruent with the lower hardness values of the TNS/M samples (**Figure 4**), which implies that these alloys are softer than the TNS for similar sintering conditions. As it was explained before, the lower hardness in the TNS/M samples resulted from the spacer-like behavior of the Mg powders. Besides, a tendency to increase the elastic modulus with the sintering temperature was observed. Being congruent with lower porosity percentages measured in the TNS₁₁₀₀ and TNS/M₁₁₀₀ samples.

Comparing the obtained elastic moduli in the studied samples with the ones reported for human bone, the TNS/M₉₀₀ sample can be the most adequate for its use in dental or orthopedic implants. Besides, the TNS/M₉₀₀ alloy joins an acceptable hardness for biomedical implants and has adequate porosity features for triggering the anchorage between organic tissue and implant material. This is, the TNS/M₉₀₀ sample

Figure 5.
Elastic modulus as function of temperature for samples sintered at 900°C and 1100°C with (TNSM) and without (TNS) Mg addition [28].

combined the best microstructural and mechanical properties to be a potential biomaterial. However, the performance of these alloys under in vivo environments should be described to determine the feasibility of the studied alloys for biomedical purposes.

6. Conclusion and future perspectives

Development of biomaterials needs to focus on the biointerface construction to match the structure of the host tissue and to meet the mechanical requirements of specific tissue. In order to do that, metallurgic and additive manufacturing techniques present great potential in the development of Ti-Mg alloys, with complex shape formed by pores to be more biocompatible. It is critical to manipulate the surface by physical and chemical parameters to achieve the clinical purpose of the biomaterial, leading to a fast integration to the bone, due to the stimulating biological functions. In this review was showed that β -Ti-Nb-Sn alloy can be fabricated using Mg to create high content of porosity (20–38%), with an elastic modulus between 31 and 49 GPa close to bone tissue, and hardness close to commercial materials and higher than different parts of skeleton.

In this sense continuous studies and researches in this field is of great relevance for materials applied in life sciences.

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Conflict of interest

The authors declare no conflict of interest.

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