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Ti–Mg–Si alloys produced by non-equilibrium processing methods: mechanical alloying and sputtering

L. Dias^a, B. Trindade^{a,*}, C. Coelho^b, S. Patankar^c, C. Draney^c, F.H. Sam Froes^c

^a ICEMS, Department de Engenharia Mecânica, Universidade de Coimbra, 3030 Coimbra, Portugal

^b Dept. Eng. Mecânica, Escola Superior de Tecnologia de Abrantes, Instituto Politécnico de Tomar, Portugal

^c Department of Metallurgy, Mining and Geological Engineering (M3GE), Institute for Materials and Advanced Processes (IMAP),

University of Idaho, Moscow, ID, USA

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Abstract

Mechanical alloying (MA) and magnetron sputtering were used as processing techniques for the synthesis of various Ti–Mg–Si (Ti-rich) lightweight alloys. The samples were analysed by means of electron probe microanalysis (EPMA), X-ray diffraction (XRD), transmission electron microscopy (TEM), differential thermal analysis (DTA) and differential scanning calorimetry (DSC). Results show that a supersaturated low range ordered solid solution of Ti(Mg, Si) was produced during milling of the Ti₇₅Mg₁₃Si₁₂ mixture, which crystallises at about 550 °C, giving rise to a dispersion of Ti₅Si₃ in a α -Ti matrix. A metastable (Ti, Mg)₅Si₃ intermetallic was observed to form during milling of the Ti₆₀Mg₁₀Si₃₀ sample obtained by two different milling processing routes: (i) from elemental Ti, Mg and Si powders and (ii) from titanium hydride (TiH_{1.924}), Mg and Si powders. The (Ti, Mg)₅Si₃ intermetallic decomposes into Ti₅Si₃ and Mg₂Si in the temperature range of 600–700 °C. Contrarily to the results obtained for the mechanically alloyed samples, no low range ordered or intermetallic phases were detected in the as-deposited sputtered thin films. All but the Ti₈₁Mg₆Si₁₃ thin film (lowest Mg content) contain two hcp phases, α -Ti and Mg, with [0 0 0 2] preferred orientations. The Ti₈₁Mg₆Si₁₃ thin film consists of an extended α -Ti solid solution. The heat treatment of the thin films lead to an increase of the structural order of these phases and to the formation of Ti₅Si₃.

Keywords: Ti-Mg-Si system; Mechanical alloying; Sputtering; Thermal behaviour

1. Introduction

Titanium-based alloys have been attractive materials for different structural applications due to their good creep performance, high strength and toughness as well as excellent corrosion resistance. There is an increasing interest in the production of these materials by mechanical alloying (MA) [1], including titanium-based intermetallics, particulate reinforced titanium alloys (MMCs) and ultra-lightweight alloys [2]. Many attempts have been made to reduce the density of Ti alloys in order to develop materials suitable for aerospace applications with high-specific strength (strength to density ratio). In fact, density reduction is three to five times more effective in reducing the structural weight than increasing the strength or modulus and consequently it also increases

E-mail address: bruno.trindade@dem.uc.pt (B. Trindade).

the specific strength [3]. Elements that can be used for this purpose include magnesium, aluminium and silicon. Titanium and magnesium are immiscible in the solid state and do not form intermetallic phases [4]. However, in the case of aluminium and silicon several titanium aluminides [5] and silicides [6] can be formed. These intermetallics are known to have good mechanical properties at elevated temperatures that cannot be met by ceramics or conventional superalloys [7]. Moreover, most of them have good resistance to oxidation at elevated temperatures. The major problem of these intermetallics is their low ductility and fracture toughness at room temperature and, in some cases, insufficient strength and creep resistance at high temperatures. One approach to improve mechanical properties is to fabricate a two phase material consisting of a ductile solid solution (for ductile phase toughening) and a discontinuous reinforcement phase consisting of intermetallic particles (MMCs). The goal of dispersed-phase alloys is to strengthen the alloy matrix by impeding the motion of dislocations. The influence of the

^{*} Corresponding author. Tel.: +351-239-790745;

fax: +351-239-790701.

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dispersoids in these systems depends on their size. Small dispersoids (1-100 nm) tend to increase yield strength while medium-size dispersoids (100-1000 nm) have a strong inhibiting effect on recrystallisation and grain growth [8]. Far from equilibrium synthesis processes such as mechanical alloying and sputtering have been used to produce MMCs materials with non-equilibrium structures, e.g. amorphous alloys, supersaturated solid solutions, metastable crystalline phases and nanocrystalline materials [9]. Cavusoglu et al. [10] synthesised a low-density Ti-Mg-Si alloy by MA and showed that it is possible to obtain intermetallic phases during subsequent annealing. According to these authors, three phases, i.e. TiN_{0.3}, Ti₅Si₃ and MgO were detected after annealing at 1100 °C. The reaction of titanium powders with contaminant nitrogen presented in the argon used during the DTA runs negated the existence of a titanium solid solution in the final structure. In a study on the production of Ti-Mg-Si thin films by sputtering, Trindade et al. [11] claimed the formation of a fine dispersion of titanium silicides in the titanium matrix during annealing of the as-deposited structure. In a recent work on HIP'd Mg-Ti-Si (Mg-rich) samples obtained by MA, Dias et al. [12] observed the formation of intermetallic nanophases during the synthesis process. Mixtures of $Mg + Mg_2Si$ and $Mg_2Si + Ti_5Si_3$ were obtained for the $Mg_{88}Ti_4Si_7$ and Mg₆₀Ti₁₀Si₃₀ HIP'd samples, respectively.

In this publication, we will present results on the synthesis and characterisation of various Ti–Mg–Si low-density alloys obtained by two far from equilibrium processes, i.e. mechanical alloying and sputtering. A comparison will be made between the phases formed during synthesis and subsequent annealing of both sets of samples. The influence of the silicon amount on the formation of intermetallic structures will also be presented.

2. Experimental

MA was used to obtain samples with nominal compositions of $T_{175}Mg_{13}Si_{12}$ and $Ti_{60}Mg_{10}Si_{30}$ (at.%) from Mg, Si and Ti powders with a nominal purity of 99.6, 99.5 and 99% and an average particle size of 60, 10 and 75 µm, respectively. Another $Ti_{60}Mg_{10}Si_{30}$ sample was also obtained using titanium hydride ($TiH_{1.924}$ with a nominal purity of 99% and an average particle size of 150 µm) instead of titanium in order to study the influence of hydrogen in the MA process. This compound has been used to accelerate the milling process and to reduce contamination, avoiding the use of a process control agent addition [10].

A planetary ball mill hardened steel vial and balls were employed for milling process. About 24 g of the powder blend was loaded into the vial. The weight ratio of ball-to-powder was approximately 20:1 during the process. The vial was sealed using a rubber 'O' ring and filled with hydrogenated argon gas. To avoid contamination all handling of powder blends was conducted under an argon atmosphere. The MA process was carried out at a rotation speed of 500 rpm. The milling process was interrupted for 10 min, each 5 min to cool the vial. Small amounts of powder were drawn out at pre-determined time intervals (1, 5, 10 and 20 h) for structural and chemical analysis. The as-MA'd samples were characterised by means of XRD. The X-ray patterns were obtained using Co K α radiation. DSC and DTA were used to evaluate the thermal stability of the ternary Ti–Mg–Si samples. The heating rate in both cases was 40 °C min⁻¹. Chemical homogeneity of the particles formed during the milling process was followed as a function of milling time by EPMA.

Ti–Mg–Si thin films with $\approx 3 \,\mu$ m thickness were co-deposited by d.c. magnetron sputtering from two targets (Ti and Mg) with silicon foils superimposed. Steel and glass samples were used as substrates. The sputtering parameters used in the depositions were the following: $p_{\text{ultimate}} = 10^{-4} \text{ Pa}$; $p_{\text{dep.}} = 1 \text{ Pa}$; $P_{\text{Mg}} = 2.22 \,\text{W cm}^{-2}$, P_{Ti} and target–substrate distance variable. The substrates were biased ($-70 \,\text{V}$). The thin films were studied in as-deposited state and after isothermal annealings at elevated temperatures, in a hydrogenated argon atmosphere (5% H₂), by means of EPMA, XRD and transmission electron microscopy (TEM). For TEM analysis the films were thinned on both sides by ion milling in an argon atmosphere. A 300 kV TEM was used for the thin foil observations.

3. Results and discussion

3.1. Mechanical alloying

Fig. 1 shows the XRD patterns from the mixtures obtained for different milling times. For all the samples the line reflections of the different phases gradually broaden with milling time implying a decrease of their structural order.

Concerning the Ti₇₅Mg₁₃Si₁₂ sample, the XRD peaks of Si and Mg gradually weakened with milling time. After 5 h of milling the peaks of the Mg phase are no longer visible and only traces of the Si phase can be detected. This means that these elements suffered an effective decrease in the grain size and/or they formed a Ti(Mg, Si) solid solution with titanium. This last hypothesis is consistent with previous results on the MA of Ti-Mg [12,13] and Ti-Al-Si [14] samples. In fact, it is recognised that it is possible to obtain Ti(Mg) and Ti(Si) supersaturated solid solutions by MA of Ti and Mg or Si. The Ti-Mg binary phase diagram indicates a value of 1 at.% as the maximum solubility of Mg in the α -Ti phase. However, various authors claim higher solubilities in MA'd samples [12–14]. Dodd et al. [13] and Dias et al. [12] report values of 17.9 and 50 at.% for the maximum solubility of Mg in the α -Ti phase, respectively. According to the Ti–Si phase diagram, the solubility of Si in the α -Ti phase is less than 0.5 at.%. In a very recent work on Ti-Al-Si samples synthesised by MA [14], the solid solubility achieved for Si in the α -Ti phase was reported to be 37.5%. Once the Ti(Mg,



Fig. 1. XRD patterns of the mechanical alloyed mixtures for different milling times: (a) $Ti_{75}Mg_{13}Si_{12}$; (b) $Ti_{60}Mg_{10}Si_{30}$; (c) $(TiH_{1.924})_{60}Mg_{10}Si_{30}$.

Si) solid solution is formed, further milling gives rise to a continuous decrease in grain size and destabilisation of the crystalline structure occurs by an increase of the number of structural defects such as vacancies, dislocations, local stresses and grain boundaries [15]. After 10 h of milling, the diffraction patterns of Ti₇₅Mg₁₃Si₁₂ sample are formed by a broad peak near $2\theta = 46^{\circ}$ (Fig. 1a) typical of short range ordered materials. An average crystallite size of ≈ 2 nm was estimated for the this sample milled for 20 h, using the Williamson and Hall method [16]. There is some kind of order in these materials, which extends over a length scale of about 20 Å, intermediate between the short-range order of liquids and the long-range order of crystals [17].

In the $Ti_{60}Mg_{10}Si_{30}$ and $(TiH_{1.924})_{60}Mg_{10}Si_{30}$ samples, the structural changes that take place during milling (Fig. 1b and c) are different from those observed in the $Ti_{75}Mg_{13}Si_{12}$ sample (Fig. 1a). Milling of both the Ti₆₀Mg₁₀Si₃₀ and (TiH_{1.924})₆₀Mg₁₀Si₃₀ samples give rise to the formation of theTi₅Si₃ intermetallic phase. However, the time required for its appearing is not the same for these two samples. In the case of sample without hydrogen, X-ray peaks of the intermetallic phase are detected on milling for 1h while milling periods higher than 5 h are necessary to produce this phase from elemental TiH1.924 and Si powders. The presence of hydrogen in the as-blended mixture in the form of titanium hydride retards the formation of the Ti₅Si₃ phase. This means that the titanium hydride first decomposed during milling (hydrogen degassing) and later reacted with silicon to form the intermetallic phase.

Hydrogen can be easily removed from titanium by vacuum annealing [18]. Reported work on TiH_{1.924}-Mg-Si alloys [10] indicates that the temperature required for hydrogen degassing depends on the synthesis process. DTA runs performed on the as-blended TiH1.924-Mg-Si powders revealed a wide endothermic peak with an onset at 563 °C and a maximum at 638 °C associated with the decomposition of the titanium hydride. However, when the powders were mechanically alloyed for 8h using a SPEX 8000 high-energy ball mill the authors claim that this endothermic peak was shifted to lower temperatures (onset at 330 °C and maximum at 350 °C). The results obtained in the present investigation show that the TiH_{1.924} phase can be decomposed prior to annealing, i.e. during MA. It is very difficult to predict the local temperature of the powders during impact of the balls. Some models [19] suggest modest temperature rises $(\approx 100-200 \,^{\circ}\text{C})$ but if a large exothermic heat reaction is involved and high-milling speeds are used, much higher local temperatures can be produced during the MA process. The local temperatures reached during synthesis of our samples were sufficient to promote the decomposition of the $TiH_{1,924}$ phase and subsequent formation of Ti₅Si₃. The formation of this intermetallic is associated to a high heat of reaction of -577.4 kJ mol⁻¹. The high intensity of milling used in this work, together with the formation of the Ti₅Si₃ intermetallic during synthesis explains why the Ti₆₀Mg₁₀Si₃₀ and (TiH_{1.924})₆₀Mg₁₀Si₃₀ samples are both formed by the same metastable structure, (Ti, Mg)₅Si₃, after 20h of milling. Moreover, the results show that the processing conditions we used in this work gave rise to a more energetic milling process than the work of Cavusoglu et al. [10] in which



Fig. 2. Standard deviation of the chemical composition results as a function of milling time: (a) $Ti_{75}Mg_{13}Si_{12}$; (b) $Ti_{60}Mg_{10}Si_{30}$; (c) $(TiH_{1.924})_{60}Mg_{10}Si_{30}$.

the existence of the TiH_{1.924} hydride is reported, even after 8 h of milling. No signs of magnesium were detected in the XRD patterns of milled samples for 20 h. As reported earlier Mg can be introduced in the α -Ti phase by MA. Furthermore, Mg forms with Si an intermetallic phase, Mg₂Si. Thus, it is likely that a metastable (Ti, Mg)₅Si₃ had been formed during milling instead of Ti₅Si₃. The lattice parameters, a and c, of this new hexagonal phase were calculated from the respective XRD patterns. These values are 0.744 and 0.515 nm for the Ti₆₀Mg₁₀Si₃₀ sample and 0.749 and 0.516 nm for the $(TiH_{1.924})_{60}Mg_{10}Si_{30}$ sample, respectively. The grain size of the (Ti, Mg)₅Si₃ phase after 20 h milling of the two samples are in the nanometric range (15 and 6 nm for the Ti₆₀Mg₁₀Si₃₀ and (TiH_{1.924})₆₀Mg₁₀Si₃₀ samples, respectively). Nanocrystalline intermetallic structures have been reported in different materials obtained by different synthesis processes. The interest in nanocrystalline intermetallics is motivated by theoretical suggestions [20-22], and some experimental data [20], which indicate that reducing the grain size of intermetallics might improve their mechanical properties, e.g. ambient temperature ductility. Microhardness is another property that is enhanced as the grain size decreases.

The powders taken from the vial at 1, 5, 10 and 20 h were also chemically analysed by EPMA. Fig. 2 shows the standard deviations (σ) of the results obtained from 10 points in each sample as a function of milling time. As can be seen, σ decreases as the time increases, meaning that the powders gradually become chemically homogeneous during MA. The presence of hydrogen retards the chemical homogenisation of the (TiH₁ 924)₆₀Mg₁₀Si₃₀ sample when compared to the sample with the same composition but without hydrogen (Ti₆₀Mg₁₀Si₃₀). This is particularly true in the first few hours of MA for the elements Ti and Si which agrees with the XRD results. The overall chemical composition of the samples milled for 20 h is also presented in Fig. 2. Some slight differences of the overall chemical composition occurred during the synthesis process. As expected, the (TiH_{1.924})₆₀Mg₁₀Si₃₀ sample is the one which exhibits a chemical composition closest to the starting composition. In fact, one of the aims of using titanium hydride instead of elemental titanium is to reduce powder cold-welding to the container wall and milling balls.



Fig. 3. DTA curves of the 20h milled powders: (a) $Ti_{75}Mg_{13}Si_{12};$ (b) $Ti_{60}Mg_{10}Si_{30};$ (c) $(TiH_{1.924})_{60}Mg_{10}Si_{30}.$

3.2. Thermal stability of the milled powders

Fig. 3 shows the DTA curves of the three 20 h milled powders up to 900 °C. The characteristic peaks in the curves indicate that phase transformations occurred during heating. In order to investigate these reactions, various thermal runs



Fig. 4. XRD patterns of the mixtures annealed at different temperatures: (a) $Ti_{75}Mg_{13}Si_{12}$; (b) $Ti_{60}Mg_{10}Si_{30}$; (c) $(TiH_{1.924})_{60}Mg_{10}Si_{30}$.

were performed for each composition up to different maximal temperatures (see arrows in Fig. 3). After each DTA run the powders were analysed using XRD (Fig. 4). The DTA curve of the Ti₇₅Mg₁₃Si₁₂ sample (Fig. 3a) shows a sharp exothermic peak, with a maximum at 580 °C, corresponding to the crystallisation of the low range ordered Ti(Mg, Si) phase into α -Ti+Ti₅Si₃. At 645 °C a small endothermic peak can be observed which might be associated with melting of Mg and formation of MgO. This oxidation might occur since Mg cannot react with the silicon to form the intermetallic phase Mg₂Si because the silicon was completely consumed in the formation of the titanium silicide Ti₅Si₃. Concomitantly an eventual oxidation of the α -Ti phase occurs during annealing which is expressed by the displacement of the major X-ray diffraction peak (Fig. 5) to lower angles as a function of temperature. The grain size of the Ti₅Si₃ intermetallic phase after heating up to 900 °C is 12 nm.

The DTA curves for the $Ti_{60}Mg_{10}Si_{30}$ and $(TiH_{1.924})_{60}$ Mg₁₀Si₃₀ samples show one exothermic peak in the range 600 °C–750 °C. However, the temperature at which this peak appears is somewhat different for the two samples, being lower for the sample containing titanium hydride powder. XRD analysis performed after annealing at temperatures before and after this exothermic peak shows that it can be attributed to the decomposition of the metastable (Ti, Mg)₅Si₃ intermetallic into the stable $Ti_5Si_3 + Mg_2Si$ phases (Fig. 4b and c).

The different thermal stabilities of these two samples indicate that their as-milled structures do not present the same level of metastability as a result of accumulation absorption of energy during MA and/or the same chemical composition. The existence of hydrogen during ball milling of the $(TiH_{1.924})_{60}Mg_{10}Si_{30}$ alloy may be another



Fig. 5. Displacement of the α -Ti (101) peak of the Ti₇₅Mg₁₃Si₁₂ sample as a function of the annealing temperature.



Fig. 6. X-ray diffraction patterns of the as-deposited thin films.



Fig. 7. Parameter c of the (Ti) and (Mg) phases for the as-deposited thin films.

fact that might influence the metastability of the (Ti, $Mg)_5Si_3$ phase. Supposing that a certain amount of hydrogen might still be present in this sample after 20 h of milling it is reasonable to accept that hydrogen would contribute to higher metastability of the (Ti, $Mg)_5Si_3$ intermetallic. In fact, a detailed observation of the DTA curve of this sample (Fig. 3c) reveals a very broad, weak and extensive endothermic peak between 200 and 400 °C that



Fig. 8. TEM results of the as-deposited thin films: (a) TEM image of $Ti_{68}Mg_{20}Si_{12}$ sample; (b) corresponding SADP.



Fig. 9. DSC thermogram of the $Ti_{68}Mg_{20}Si_{12}$ thin film.

could be associated with its dehydriding. This dehydriding process was also observed by Cavusoglu et al. [10] on a $(TiH_{1.924})$ –Mg–Si ball-milled sample in the same temperature range.



Fig. 10. XRD patterns of the (a) $Ti_{68}Mg_{20}Si_{12}$ and (b) $Ti_{81}Mg_6Si_{13}$ thin films samples annealed at 500, 600 and 650 $^\circ C.$

The XRD patterns of the $Ti_{60}Mg_{10}Si_{30}$ and $(TiH_{1.924})_{60}$ Mg₁₀Si₃₀ samples heated up to 900 °C (Fig. 4b and c) reveal that both samples are formed by $Ti_5Si_3 + Mg_2Si$ with grain sizes of 102 and 41 nm for the former sample and 34 and 27 nm for the sample obtained with titanium hydride, respectively. These results show that similar structures are obtained from different powders mixtures.

3.3. Thin films obtained by sputtering

The X-ray diffraction patterns of the as-sputtered films are presented in Fig. 6. All but the $Ti_{81}Mg_6Si_{13}$ thin film (lowest Mg content) exhibited two hcp phases, α -Ti and Mg, with [0002] preferred orientations, in agreement with previous results on binary Ti-Mg sputtered films [9]. This texture corresponds to the closely packed planes (basal planes). The $Ti_{81}Mg_6Si_{13}$ thin film is formed by a α -Ti solid solution. As the samples become richer in magnesium the peaks of the (Mg) phase become more intense. The interplanar distances d(0002) of both phases were used to calculate the c parameter of the hcp structures of the as-deposited thin films (Fig. 7). While the α -Ti d(0002) interplanar distance remains almost constant as a function of chemical composition and do not differ significantly from the equilibrium position given by the ICDD card (2.341 Å), the same is not true for the hcp Mg phase. In fact, the values of the d(0002)interplanar distance obtained are lower than what would be expected (2.605 Å according to the ICDD card). This is particularly true for the $Ti_{75}Mg_{14}Si_{11}$ sample (Ti/Mg = 5.4). With increasing Mg content the d(0002) values of the hcp

Mg phase tend to follow Vegards law, corroborating the results of previous work on the Ti–Mg binary system [9].

Contrarily to the results obtained for the MA'd samples, no low range ordered structures were detected in the as-deposited thins films. However, the grain size of the (Ti) and (Mg) solid solution, calculated by the Sherrer equation on the basis of the FWHM of the (0002) XRD peaks, is in the range of 2–3 nm. This means that these thin films consist of nanometric grains with intermediate range order.

Fig. 8 shows transmission electron microscopy results obtained from the $Ti_{68}Mg_{20}Si_{12}$ thin film, as a typical example of the as-deposited thin films with a Ti/Mg atomic ratio equal or higher than 5.4. The films are characterised by low-contrast images (Fig. 8a). There is a marked difference between these images and those obtained for the films of the Ti–Mg binary system for similar Ti/Mg atomic ratios [9]. This suggests that silicon is responsible for the grain size decrease in these ternary films. The corresponding electron diffraction pattern (Fig. 8b) support this conclusion as it is formed by two large weak halos characteristic of structures with a decreased structural order.

3.4. Thermal stability of the sputtered thin films

In order to evaluate the thermal stability of the sputtered thin films, thermal analysis was performed using DSC. Fig. 9 shows the thermogram obtained from the $Ti_{68}Mg_{20}Si_{12}$ sample up to 650 °C, in an exothermic peak with an onset at 545 °C and a maximum at 575 °C is present. In order to find out the reaction associated to this peak, the sample was



Fig. 11. TEM images and corresponding SADP's of the 600 °C annealed films. (a) Ti₈₁Mg₆Si₁₃, (b) Ti₆₈Mg₂₀Si₁₂.

annealed at maximal temperatures of 500, 600 and 650 °C for 1 h and analysed by XRD after cooling to room temperature. The X-ray diffraction patterns of the $Ti_{68}Mg_{20}Si_{12}$ annealed samples are shown in Fig. 10a. These heat treatments lead to an increase of the structural order of the α -Ti and (Mg) phases and to the formation of Ti_5Si_3 . The appearance of this intermetallic is responsible for the exothermic peak observed in the DSC thermogram that occurred in the temperature range 545–590 °C.

The sample containing the lowest Mg content was also annealed at 500, 600 and 650 °C. The X-ray diffraction patterns obtained after cooling (Fig. 10b) reveal a peak at 2θ = 39.9°, corresponding to the (0002) plane of the hcp Mg phase. This means that annealing resulted in the decomposition of the metastable Ti(Mg) solid solution into two stable phases, i.e. α -Ti + (Mg).

Fig. 11 shows TEM images of the two sets of thin films annealed at 600 °C/1 h: one corresponding to the $Ti_{81}Mg_6Si_{13}$ sample (Fig. 11a) and the other to $Ti_{75}Mg_{14}Si_{11}$ (Fig. 11b). This last image is a typical example of as-deposited samples formed by two structural phases. Some grains of about 40 nm size can be observed, implying a higher structural range order of the annealed films when compared to the same films in the as-deposited condition. However, the grains are not big enough to obtain clear and independent electron diffraction patterns from the different phases. The selected area diffraction patterns (SADP's) obtained can be attributed mainly to the Ti_5Si_3 phase. However some spots are also visible, which can be index to the α -Ti phase. Similar structure was obtained for the corresponding MA'd and annealed sample.

4. Conclusions

Different Ti–Mg–Si ternary samples were synthesised by two far from equilibrium techniques, i.e. mechanical alloying and sputtering. Extended Ti(Mg, Si) low range ordered solid solution was obtained after milling of the Ti₇₅Mg₁₃Si₁₂ mixture. This structure gives rise to the Ti₅Si₃ intermetallic during annealing. A metastable (Ti, Mg)₅Si₃ intermetallic was observed to form during synthesis of the Ti₆₀Mg₁₀Si₃₀ sample obtained by two different milling processing routes: (i) from elemental Ti, Mg and Si powders and (ii) from titanium hydride (TiH_{1.924}), Mg and Si powders. The (Ti, Mg)₅Si₃ intermetallic decomposes into Ti₅Si₃ and Mg₂Si in the temperature range of 600–700 °C.

In contrast to mechanically alloyed samples, no low range ordered or intermetallic phases were found to be formed in the sputtered thin films. All but the $Ti_{81}Mg_6Si_{13}$ thin film (lowest Mg content) present two hcp phases, α -Ti and Mg, with [0002] preferred orientation. The $Ti_{81}Mg_6Si_{13}$ thin film is formed by a metastable α -Ti solid solution. The heat treatment of these films lead to an increase of the structural order of these phases and to the formation of Ti_5Si_3 .

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