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Abstract:

The synthesis of supersaturated bcc Fe$_{44}$Co$_{44}$Sn$_{12}$ by mechanical alloying was followed by various techniques. Two stannides, CoSn$_5$ and FeSn$_2$ form first at interfaces. Tetragonal CoSn$_5$, a very recently discovered stannide, is obtained here by a simple method. These stannides then dissolve and the ternary bcc alloy forms with a progressive and simultaneous dissolution of Co and Sn into bcc Fe.

Keywords: Mechanical alloying, Fe-Co-Sn alloys, X-ray diffraction, Magnetization, Mössbauer spectroscopy

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1- Introduction

Near-equiaxial FeCo alloys are b.c.c. below ~1250 K [1]. The b.c.c. phase (A2) orders to a CsCl type structure (B2) at temperatures below ~1000 K. Iron-cobalt alloys possess exceptional magnetic properties which make them useful as soft magnetic materials [1]. Various ternary or multinary systems based on near-equiaxial FeCo were thoroughly investigated [1]. The addition of a small amount of vanadium to FeCo helps for instance to circumvent its brittleness [1]. We started recently to study metastable (Fe_{50-x/2}Co_{50-x/2})Sn ternary alloys prepared by mechanical alloying (MA) [2-3]. The structures and properties of alloys of the ternary system Fe-Co-Sn are essentially uninvestigated. The magnetic and magneto-optical properties of the compound FeCoSn, with a hexagonal Ni_{2}In type structure, were studied in ref. [4]. Ternary Fe-Co-Sn alloys were electrodeposited from a gluconate electrolyte [5]. Most of the alloys have a tin content larger than 82 at.%. The remaining alloys have compositions scattered around ~Fe_{25}Co_{25}Sn_{50}. An amorphous phase is the dominant phase in all these deposits. Co_{2}FeSn Heusler alloys films were prepared by controlled deposition of successive layers of Co, Fe and Sn [6]. In addition, Fe/MgO/Co_{2}FeSn magnetic tunnel junctions were characterized [6]. Finally, nanocrystalline tetragonal (Fe_{1-y}Co_{y})Sn_{2} were prepared by chemical methods to improve the electrochemical performance of Sn intermetallic compounds as electrode active material for Li-ion batteries [7].

Mechanical alloying was used successfully to enhance the solubility of Sn in bcc Fe from ~9 at% to ~20 at% [8-9]. The equilibrium solubility of Sn in FeCo (A2 or B2) is quite low at thermal equilibrium, being less than 1 at.% [10]. We chose thus MA as our preparation method to increase significantly the dissolution of tin in bcc
FeCo. A clear-cut value cannot be attributed readily to the actual supersaturation of as-milled \((\text{Fe}_{50-x/2}\text{Co}_{50-x/2})\text{Sn}_x\). In our milling conditions, the maximum solubility \(x_m\), \(14 \leq x_m \leq 20\), is likely closer to 20 [2-3]. The problems raised by the determination of \(x_m\), which emphasize analogies between these ternary alloys and as-milled \(\text{Fe}_{100-x}\text{Sn}_x\), will be discussed in detail elsewhere.

The phenomena which occur during MA from elemental powder mixtures were found to remain essentially unchanged for \((\text{Fe}_{50-x/2}\text{Co}_{50-x/2})\text{Sn}_x\) with \(x \leq x_m\), being for instance qualitatively similar for \(x=6, 12\) and 20. The present paper focuses on the mechanisms of mechanosynthesis of one of these alloys, namely \(\text{Fe}_{44}\text{Co}_{44}\text{Sn}_{12}\). The milled powders were characterized as a function of milling time by a combination of various techniques: X-ray and neutron diffractions, magnetic measurements, Mössbauer spectroscopy of \(^{57}\text{Fe}\) and \(^{119}\text{Sn}\).

2- Experimental

Powder mixtures of elemental Fe (purity: 99.9\%), Co (purity: 99.8\%) and Sn powders (purity: 99.5\%) were mixed together and milled in argon atmosphere in a planetary Fritsch P6 ball-mill at 500 rpm for different milling times. The vials and the seven balls of 15 mm diameter used to mill were made of hardened steel. The powder to ball weight ratio was 1:20. X-ray diffraction (XRD) patterns were recorded using Cu \(K\alpha\) radiation (\(\lambda = 0.1542\) nm). SEM images and X-ray maps (Fe, Co, Sn) were obtained with a JEOL JSM 630 microscope. X-ray maps of particles of a powder milled for 12 h show that Fe, Co and Sn are homogeneously distributed. The latter sample is slightly enriched in Fe coming from milling tools, being actually \(\text{Fe}_{46}\text{Co}_{42}\text{Sn}_{12}\). As this enrichment in Fe depends on milling time, the alloy will be denoted for convenience by its initial composition \(\text{Fe}_{44}\text{Co}_{44}\text{Sn}_{12}\).
Magnetization measurements were carried out in a cryogen-free vibrating sample magnetometer (VSM) of CRYOGENIC Instruments with magnetic field up to 10 T and temperature in the range 1.7 K up to 320 K. The saturation magnetizations were measured at T=5 K. $^{57}\text{Fe}$ and $^{119}\text{Sn}$ Mössbauer spectra were recorded at room temperature (RT), in transmission geometry, by a constant acceleration type spectrometer using sources of $^{57}\text{Co}$ in Rh and of $^{119}\text{Sn}$ in CaSnO$_3$ both with a strength of ~10 mCi. As usual, isomer shifts are referenced to $\alpha$-Fe at RT and to SnO$_2$ for Fe and Sn respectively. Hyperfine magnetic field (HMF) distributions, P(B), were analyzed with a constrained Hesse-Rübartsch method [11] employing Lorentz lines. Here, P(B)$\Delta$B represents the fraction of iron or tin atoms whose field lies between B and B+$\Delta$B. The mean HMF’s are denoted $<B_F>$ and $<B_S>$, where the indices F and S stand for $^{57}\text{Fe}$ and $^{119}\text{Sn}$ respectively. The ‘width’ of a distribution P$_X$(B) (X=F,S) is characterized here by the standard-deviation $\sigma_X=<(B-<B_X>)^2>^{1/2}$.

3- Results and discussion

3.1 First stage: 0.5h \leq t_m \leq 2h, stannide formation

In the MA process of ductile powder mixtures of A and B, particles are repeatedly flattened, fractured and welded. A layered structure of A and B is thus formed and progressively refined and convoluted ([12] and references therein). Interfaces thus play an important role in the synthesis process for short milling times. Three types of interfaces are formed in the case of Fe-Co-Sn: Co-Sn, Fe-Sn and Fe-Co. XRD patterns (fig.1) show broadened diffraction peaks of the starting elements, some of which become rapidly difficult to be evidenced. A broad line due to fcc Co is observed at \sim52° for t_m\geq1h (‘C’ in fig.1). The other clearly visible peaks are associated with two
stannides: CoSn₅ and FeSn₂. During this first stage, fluctuations inherent to the as-prepared mixtures still persist but their magnitude weakens with tₘ (ex: composition fluctuations from particle to particle). Therefore, the phenomena we describe below characterize the average behavior. Deviations from it may exist but they are often difficult to be evidenced and to be depicted with enough precision.

3.1.1 Co-Sn interfaces

The most intense lines at tₘ=0.5h, indexed as ‘1’ in fig.1 and other lines which are not seen at the scale of fig.1, are due to a compound “CoSn₅”, possibly non-stoichiometric (Co₁₋ₓSn₅, z≈0.1-0.2), whose existence was unreported till recently [13]. Nanospheres of CoSn₅ were synthesized through a conversion chemistry route. It is tetragonal (P4/mcc space group) with lattice parameters are a=b=0.69328 nm, c=0.57924 nm [13]. The cobalt stannide formed by MA does not contain iron as confirmed by neutron diffraction patterns and by ¹¹⁹Sn Mössbauer spectra recorded at 15K and at RT. Indeed, the ¹¹⁹Sn spectra of Co₈₈Sn₁₂ milled for tₘ=0.5h differ almost not from those of Fe₄₄Co₄₄Sn₁₂ (fig. 2). Therefore, CoSn₅ does not contribute to ⁵⁷Fe Mössbauer spectra. The intensities of CoSn₅ XRD peaks decrease rapidly. The main line is broad and weak at tₘ=1.5h and is no more observed for tₘ≥2h (fig. 1). We found that CoSn₅ forms too in the early stage of MA of Co-Sn alloys at any Sn content ranging between 6 and 83 at.%. The very rapid formation of CoSn₅ at Co-Sn interfaces can be understood from the fact that Co is an anomalously fast diffuser into Sn [14]. CoSn₅ is non-magnetic as are all tin-rich cobalt stannides.
3.1.2 Fe-Sn interfaces

The second stannide (‘2’ in fig.1) is tetragonal FeSn$_2$ with a CuAl$_2$ type structure ($I4/mcm$ space group) and $a=b=0.6539$ nm, $c=0.5325$ nm. The most intense XRD peak of FeSn$_2$ is still seen at $t_m=3h$. Its decrease starts at $t_m=2h$. This distannide may become enriched in Co when MA proceeds. The XRD line positions, which remain essentially unchanged with $t_m$, indicate however that Co remains a minor alloying element (fig.1 of [7]). FeSn$_2$ is repeatedly reported to form at Fe-Sn interfaces in all kinds of experiments and notably during the mechanosynthesis of Fe-Sn alloys [8-9]. The clear asymmetry of the central part of the $^{57}$Fe Mössbauer spectra observed for $t_m\leq2h$ (fig. 2) is consistent with the formation of antiferromagnetic FeSn$_2$ [15] which becomes superparamagnetic when nanostructured. Its $^{57}$Fe spectrum consists then in a single line at 0.50 mm/s [7,15]. FeSn$_2$ forms more slowly than CoSn$_5$.

3.1.3 Fe-Co interfaces

For short milling times, Fe-Co interfaces favor the formation of solid solutions whose compositions fluctuate. The lattice parameter of the bcc phase (‘F’ in fig. 1) remains essentially constant for $t_m\leq2h$ (fig. 3). This result can only be explained by the formation of Fe-rich bcc Fe-Co alloys, with a Co content less than ~25 at.%, a concentration range in which the lattice parameter of bcc Fe-Co alloys varies very little [16]. These Fe-Co alloys remain thus unobserved in $^{119}$Sn Mössbauer spectra. Simultaneously, the $^{57}$Fe Mössbauer spectra evolve slightly (fig. 2). The associated HMF distributions $P_F(B)$ go through a maximum for the field of bcc $\alpha$-Fe at RT, 33.1T (fig.4). Most of the central part of $P_F(B)$ is indeed due to unalloyed bcc Fe. The amplitude of $P_F(B)$ and the mean-field $<B_F>$ decrease (table 1). Simultaneously, the standard deviation increases because of the progressive growing of both low-field and
high-field tails (fig.4). From the concentration dependence of the mean $^{57}$Fe HMF in Fe-Co alloys [16], the low-field tail is attributed to hcp or fcc Co-rich Co-Fe alloys, with a Co content larger than ~80 at.%, while the high-field tail is ascribed to bcc Fe-rich Fe-Co alloys with a Co content less than ~25 at.% in agreement with XRD.

The decrease of the saturation magnetization for $t_m \leq 2h$ (fig. 3) is primarily due to the formation of the previous stannides. $^{119}$Sn Mössbauer spectra for $t_m=1.5h$ and 2h shows the presence of a magnetic component (fig.2). Some HMF distributions are plotted in fig.4 for $^{119}$Sn and $^{57}$Fe. Once CoSn$_5$ begins to disappear, tin is released and starts to dissolve locally into hcp and fcc cobalt. A similar phenomenon occurs later with FeSn$_2$ and bcc Fe. The HMF’s of $^{119}$Sn dissolved in bcc Fe, hcp Co and fcc Co at RT are respectively 7.9T [17], 4T [18] and 1.7T [18]. The HMF distribution of fig.4 ($t_m=1.5h$) is consistent with the transient formation of Fe and Co alloys with limited contents of Sn. The evaluation of the relative amounts of Sn engaged in the various phases from Mössbauer spectra must be cautious as long as $\beta$-Sn has not disappeared because the latter phase has a very small Lamb-Mössbauer factor at RT (~0.04, [19]). Further, its disappearance is better judged from Mössbauer spectra than from XRD because of nanometer-sized grains. In any case, no significant amounts of concentrated bcc Fe-Sn or Fe-Co-Sn can be formed during this first step because the lattice parameter would otherwise increase [2-3, 17] in contradiction with experiment.

3.2 Second stage: $t_m \geq 2.5h$

For $t_m > 2h$, the lattice parameter “a” of the bcc phase and the saturation magnetization increase and reach a plateau at $t_m > 8h$ (fig.3). Similarly, the $^{57}$Fe Mössbauer lines broaden progressively, $<B_F>$ decreases and $\sigma_F$ increases (fig.2, table 1). $<B_S>$ increases first and then it levels off at a field of 10T for $t_m \geq 8h$ (table 1), a
value discussed in detail in [2-3]. These results imply the formation of a bcc ternary alloy progressively enriched in Sn. This tin enrichment must be fast enough as compared to the speed of dissolution of cobalt into the bcc phase to prevent i) a decrease of “a”, which occurs for Co contents larger than ~28 at.% in bcc Fe-Co alloys [16], ii) an overshoot of the average magnetic moment per transition metal atom <μ> and iii) an ensuing increase of <B⊥> (36T for ~30 at.% Co [16]). Indeed, 

<μ(0)>=(μFe+μCo)/2=1.97μB/at(Fe,Co) for as-prepared Fe₄₄Co₄₄Sn₁₂ powder mixtures while the final moment of our ternary alloy is 2.01μB/at(Fe,Co) (fig.3). Both are smaller than <μ> for bcc Fe-Co with ~30 at.% Co, 2.46μB/at(Fe,Co). The moment <μ(tm)> increases during the second stage on the one hand because of the progressive dissolution of the stannides formed during the first stage and on the other because of the formation of a ternary bcc alloy whose composition evolves progressively to the final one. When the tin HMF’s in bcc Fe-Co-Sn alloys increase on average, then the inner lines of the associated magnetic subspectra move apart and a central peak becomes visible (tm>3.5h, fig. 2). Its intensity goes through a maximum and then decreases almost to zero. Simultaneously, a low intensity and broad XRD peak appears at ~30° for 2.5h ≤ tm ≤ 6h (fig.1 and inset for tm=4h).

The halo at ~30° might be due to an amorphous tin oxide. The XRD pattern (Cu Kα) of amorphous SnO has its main component centered at 30° [20] while the first broad peak of amorphous SnO₂ is found at ~26° [21]. An amorphous SnOₓ oxide, with 1<z<2, would be needed to explain the central ⁰³¹⁹Sn peak (fig. 2). The first halo of such an oxide is expected however to shift to angles smaller than 30°. The halo at ~30° might come from another amorphous phase. Indeed, the Cu Kα XRD patterns reported for amorphous Co₁₀₀₋ₓSnₓ alloys prepared by different methods (sputtering, 26≤x≤45 [22], solvothemal route, x=40 [23], MA, x=15, 35, 50 [24]) are all similar
with a broad first halo at ~31°. A similar peak is observed on the XRD pattern of MA Fe_{75}Sn_{25} (fig.5 of [9]). A second halo is seen at ~43° [22] and would be hidden here by the main bcc peak. The halo positions observed on CoKα XRD patterns of amorphous Fe-Co-Sn [5], once converted to CuKα radiation, are in full agreement with the previous value. Thus, the small peak (inset of fig.1) might be too attributed to an amorphous Co-Fe-Sn alloy but the transient central ^{119}Sn peak would still remain to be explained. At this point, it is difficult to decide which of the two attributions of the small intensity XRD peak is to be retained. Further, both might coexist. Additional characterization is needed to solve this question. Finally, the supersaturated bcc phase in dynamical equilibrium in our milling conditions is found for t_m ≥ 8h.

4- Conclusion

The mechanical alloying of supersaturated bcc Fe_{44}Co_{44}Sn_{12} was studied as a function of milling time. Two stannides, CoSn_{5} and FeSn_{2}, form first at interfaces. Tetragonal CoSn_{5}, a very recently discovered stannide [13], is obtained here by a simple method. Its very rapid formation results from the anomalously fast diffusion of Co into Sn. These stannides dissolve and the ternary bcc alloy form with a simultaneous dissolution of Co and Sn into bcc Fe until the dynamic equilibrium phase is obtained. The MA mechanisms are qualitatively similar for all supersaturated bcc alloys we studied (Sn content: x ≤ 20). The as-milled alloys (x≤20) order by annealing during 15h at 673K as proven unambiguously by neutron diffraction patterns and ^{119}Sn Mössbauer spectra (in preparation).
Acknowledgements

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References


Table 1: Mean $^{119}$Sn (S) and $^{57}$Fe (F) HMF’s and standard-deviations at different milling times ($<B_S>=0T$ and $<B_F>=33.1T$ at $t_m=0$)

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<th>$t_m$(h)</th>
<th>$&lt;B_S&gt;$ (T)</th>
<th>$\sigma_S$ (T)</th>
<th>$t_m$(h)</th>
<th>$&lt;B_F&gt;$ (T)</th>
<th>$\sigma_F$ (T)</th>
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</thead>
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<td>1.5</td>
<td>6.9 ± 0.2</td>
<td>3.4 ± 0.2</td>
<td>0.5</td>
<td>32.9 ± 0.2</td>
<td>1.7 ± 0.2</td>
</tr>
<tr>
<td>2</td>
<td>8.2 ± 0.3</td>
<td>4.7 ± 0.3</td>
<td>2</td>
<td>32.4 ± 0.3</td>
<td>2.4 ± 0.3</td>
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<tr>
<td>$\geq$8</td>
<td>10.1 ± 0.3</td>
<td>4.5 ± 0.3</td>
<td>12</td>
<td>31.2 ± 0.2</td>
<td>3.3 ± 0.3</td>
</tr>
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</table>

Figure Captions

Figure 1- Evolution with milling time, $t_m$ (h) (indicated on the left), of the X-ray diffraction patterns (CuKα) of mechanically alloyed Fe$_{44}$Co$_{44}$Sn$_{12}$.

Figure 2- Room-temperature $^{119}$Sn (left) and $^{57}$Fe (right) Mössbauer spectra of mechanically alloyed Fe$_{44}$Co$_{44}$Sn$_{12}$ as a function of milling time $t_m$ (h) (indicated on the left).

Figure 3- Evolution with milling time, $t_m$ (h), of two characteristics of mechanically alloyed Fe$_{44}$Co$_{44}$Sn$_{12}$: left) lattice parameter of the bcc phase evidenced in X-ray diffraction patterns (fig. 1) right) saturation magnetization at 5K.

Figure 4- Hyperfine magnetic field distributions at RT of Fe$_{44}$Co$_{44}$Sn$_{12}$ for the indicated milling times: $^{119}$Sn (left) and $^{57}$Fe (right).
Figure 1- Evolution with milling time, $t_m$ (h) (indicated on the left), of the X-ray diffraction patterns ($\text{Cu}K\alpha$) of mechanically alloyed $\text{Fe}_{44}\text{Co}_{44}\text{Sn}_{12}$.
Figure 2- RT $^{119}$Sn (left) and $^{57}$Fe (right) Mössbauer spectra of mechanically alloyed Fe$_{44}$Co$_{44}$Sn$_{12}$ as a function of milling time $t_m$ (h) (indicated on the left).

Figure 3- Evolution with milling time, $t_m$ (h), of two characteristics of mechanically alloyed Fe$_{44}$Co$_{44}$Sn$_{12}$: left) lattice parameter of the bcc phase evidenced in X-ray diffraction patterns (fig. 1), right) saturation magnetization at 5K.
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