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Correlation between microstructure and temperature dependence of magnetic properties in Fe$_{60}$Co$_{18}$Ni$_{6}$B$_{15}$Cu$_{1}$ alloy series


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I. INTRODUCTION

Soft magnetic Fe-based nanocrystalline alloys are among the softest magnetic materials known. The interest of the scientific community on this field was attracted by the work of Yoshizawa et al. in which the fabrication of Fe–Si–Nb–B–Cu (Finemet) alloys was reported. Besides this one, following works focused mainly on other families of alloys: Fe–(Nb,Zr,Hf,…)–B–(Cu) (Nanoperm) and Fe–Co–(Nb,Zr,Hf,…)–B–(Cu) (Hitperm). In the case of Hitperm alloys, they were developed in the aim of extending the soft magnetic properties to higher temperatures as previous systems lost their magnetic softness at the relatively low Curie temperature of the residual amorphous phase. Consequently, high-temperature behavior of Hitperm alloys has been studied in recent years from the application point of view. The peculiar microstructure exhibited by these alloys, in which ferromagnetic nanosized crystallites are embedded in a ferromagnetic residual amorphous phase with a lower Curie temperature, is responsible for their outstanding soft magnetic properties. Herzer successfully explained the main features observed in the magnetic behavior of these nanocrystalline alloys. However, Herzer’s model concerns a single nanocrystalline phase with an average crystal size (D), describing the coercivity dependence on crystal size by the well-known (1/D$^3$) power law, but nanocrystalline systems developed by controlled crystallization of an amorphous precursor alloy always consist at least of two phases: nanocrystals plus residual amorphous matrix. Thus, Herzer’s original model fails to describe the thermal dependence of coercivity close to the Curie temperature of the amorphous phase and its rise observed at the very beginning of nanocrystallization. Some years after Herzer’s work, Hernando et al. and Suzuki and Cadogan extended Herzer’s model to biphasic systems.

In this work, thermal and microstructural dependences of magnetic properties are reported for a low Co containing Hitperm series with partial or total substitution of Zr for Nb. Studies on the effect of the early transition metal (Zr, Nb, Ta, Hf) on the crystallization, microstructure and magnetic properties can be found in literature for Hitperm alloys with medium and high Co content. Results obtained in this study were quantitatively analyzed in the frame of the two phase model developed by Hernando et al. and Suzuki and Cadogan. Therefore, it will be helpful to recall some ideas concerning these models.

The effective magnetic anisotropy, $K_{\text{exp}}$, of a system is proportional to the product of the saturation magnetization, $M_s$, and the coercivity, $H_C$: $K_{\text{exp}} \propto M_s H_C$. 

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Temperature dependence of magnetic properties of nanocrystalline Fe$_{60}$Co$_{18}$Cu$_{1}$B$_{15}$Nb$_{6}$Zr$_{x}$ (x=0, 3, 6) alloys has been studied at different stages of devitrification. Transmission electron microscopy shows nanocrystals of the size $\sim$5 nm, which remains almost constant along the nanocrystallization process. Curie temperature of the residual amorphous phase decreases as nanocrystallization progresses for all the studied alloys. Thermal dependence of the exchange stiffness constant is obtained from the measurement of specific magnetization and coercivity as a function of crystalline fraction and temperature for the three studied alloys. © 2009 American Institute of Physics. DOI: 10.1063/1.3125515
\[ p_C K_{\text{exp}} = \mu_0 H C M_S, \]

where \( p_C \) is a constant (\(-0.2\) for Finemet alloys and 0.64 for cubic crystals\(^{16}\)) and \( \mu_0 \) is the magnetic permeability of vacuum.

On the other hand, the observed anisotropy is a combination of different contributions: mainly magnetocrystalline, \( K_C \), and induced, \( K_I \), anisotropies.

\[ K_{\text{exp}} = \sqrt{\langle K_C \rangle^2 + \langle K_I \rangle^2}. \]

In the previous expression, average values are used due to the complexity of the system. Finally, the crystalline contribution to the anisotropy would be affected by the crystalline fraction, \( X_C \), and can be expressed as\(^{16,19}\)

\[ \langle K_C \rangle = X_C^2 \left( K_C^{\text{crystalline}} \right) = X_C^2 \frac{K_1^2(D)^6}{A^3}, \]

where \( K_1 \) is the magnetocrystalline anisotropy constant of the crystalline phase (\(-10^4 \text{ J m}^{-3}\)) (Ref. 20) and \( A \) is the exchange stiffness constant. The thermal and microstructural dependences of this parameter for the low Co containing Hitperm series studied in this work will be also reported.

II. EXPERIMENTAL

Amorphous ribbons (\(-5 \text{ mm wide and } 20–30 \mu m\) thick) with nominal compositions Fe\(_{60}\)Co\(_{18}\)Cu\(_{11}\)B\(_{15}\)Nb\(_{\alpha}\)−Zr\(_{\beta}\) (\(\alpha=0, 3, 6\)) (in the following referred to as Nb, NbZr and Zr alloys, respectively) were prepared by melt-spinning technique. Previous differential scanning calorimetry (DSC) experiments\(^{21}\) showed that the devitrification of these alloys occurs in several transformation steps, evidenced by exothermic processes. In the first one, \(\alpha\)-Fe, Co nanocrystals appear embedded in a residual amorphous matrix enriched in B and Nb and/or Zr. In order to study the nanocrystallization process and its effect on the magnetic properties, several nanocrystalline samples were prepared by heating amorphous as-cast ribbon pieces (6 mm long) at 20 K/min up to selected temperatures in vacuum in a halogen lamp furnace. The final temperatures were chosen in order to achieve 10\%, 30\%, 60\%, and 90\% of the enthalpy ascribed to the nanocrystallization fraction, \( X_C \), and can be expressed as\(^{16,19}\)

FIG. 1. Bright field TEM images of the different studied samples.
changes from 4.2 to 5.7 nm for Nb alloy, from 4.9 to 6.4 nm for Zr alloy and from 4.7 to 6.2 nm for NbZr alloy, with $\Delta(D) = \sigma \sqrt{N} = \pm 0.3$, where $\sigma$ is the standard deviation of the corresponding crystal size distribution of Fig. 2 and $N = 50$, the number of the measured crystals. These small changes are in agreement with the instantaneous growth approximation describing the nanocrystallization kinetics of this kind of alloy. Under this approach, the time required for a nanocrystal to achieve its final size is negligible in comparison with the time of nanocrystallization process.

In the present study, no differences appear after partial or total substitution of Zr for Nb. In all the studied samples, the shape of the crystals was regular (no agglomerates were appreciated) and their average size clearly lower than that of Fe$_{44}$Co$_{44}$Zr$_7$B$_4$Cu$_1$ alloy. These facts indicate that Cu clustering phenomenon is reported. A higher than that of Hitperm alloys containing Zr, where no Cu clustering phenomenon is reported.

**B. Mössbauer spectrometry**

Figure 3 shows, as an example, Mössbauer spectra of as-cast and nanocrystallized samples for NbZr alloy. For all the studied alloys, as-cast samples exhibit a typical very broad sextet characteristic of amorphous ferromagnets but, as nanocrystallization progresses, a new and sharper sextet grows at higher values of velocity, ascribed to $\alpha$-Fe,Co nanocrystals. Although the spectra of amorphous samples could be fitted using a single distribution of magnetic hyperfine fields (HFs), nanocrystallized samples need a more complex model. In the studied case, four discrete sextets and two different HF distributions were used. The discrete sextets (pure crystalline contributions) are necessary to represent the different Fe environment (characterized by the number of Co neighbors) existing in a bcc phase with an expected composition Fe$_{60}$Co$_{18}$Nb$_3$Zr$_3$B$_{15}$Cu$_1$. The two HF distributions can be roughly ascribed to Fe atoms in the amorphous phase (low field contribution) and to Fe atoms at interface regions (high field contribution with HF $< 33$ T). The aim of using two different HF distributions is to preserve the different nature of both contributions: Fe located in amorphous regions and Fe located at interface regions. Pure crystalline contribution and the sum of both HF distributions for NbZr alloy are also represented in Fig. 3 along with the experimental data and the total fitting. The probability distribution for the different hyperfine field contributions obtained from the fitting is also shown as an example for NbZr alloy in this figure.

The average values derived from the fitting procedure (magnetic hyperfine field, $\langle$HF$\rangle$; isomer shift, $\langle$IS$\rangle$; intensity ratio between the second and third lines of the sextets, $\langle R_{23} \rangle$) and the area fraction of pure crystalline contributions, $A_{\text{cryst}}$, are shown in Fig. 4 as a function of $X_{\text{DSC}}$. The continuous increase of $A_{\text{cryst}}$ with $X_{\text{DSC}}$ is due to the close relationship between these two parameters. However, they are sensitive to different features of the crystallization process and some information can be extracted from a deeper comparison. $X_{\text{DSC}}$ is the enthalpy needed to achieve a certain nanocrystalline microstructure normalized to the enthalpy released after completing the nanocrystallization process. $A_{\text{cryst}}$ is the fraction of Fe atoms in pure crystalline sites (without taking into account Fe atoms at interface). As a first approximation, both magnitudes can be considered proportional to the crystalline volume fraction. Moreover, a constant composition of the formed $\alpha$-Fe,Co phase along the nanocrystallization process supports this approximation. Although this approximation is well described for nanocrystalline samples by a linear dependence between both magnitudes, this linearity cannot be extended to $X_{\text{DSC}}=0$ (see Fig. 4). In fact, the evolution of
for these nanocrystalline systems based on continuous nucleation and instantaneous growth processes. This mechanism predicts faster crystallization kinetics than for conventional nucleation and growth approximation for this kind of alloys.

For Nb and NbZr alloys, the angle decreases as nanocrystallization progresses, for Zr alloy it remains almost constant independent of $X_{\text{DSC}}$.

**C. Thermomagnetic gravimetry**

Figure 5 shows TMG plots for as-cast and nanocrystalline samples of the three studied alloys. For as-cast samples, a first fall to zero between 600 and 700 K is observed for all the studied alloys due to the Curie transition temperature of the amorphous phase. About 750 K, a rise in the relative magnetization indicates the onset of nanocrystallization, in agreement with DSC results. As nanocrystallization progresses, magnetization does not fall to zero because of the formed ferromagnetic $\alpha$-Fe,Co phase which has a higher Curie temperature than the amorphous one and higher than the explored range. The rise in magnetization shifts to higher temperatures in agreement with the enhancement of thermal stability during nanocrystallization.

**D. Thermal dependence of the initial permeability**

Figure 6 shows initial permeability, $\mu$, versus temperature. These in situ measurements were performed starting from an amorphous as-cast sample and both thermal treatment and measurement were done at the same time during several heating-cooling cycles, progressively increasing the maximum temperature of the $n$th cycle, $T_{\text{heat}}(n)$. After comparing the cooling branch of the $n$th cycle with the heating one corresponding to the $(n+1)$th cycle, it can be observed that $\mu$ is reversible up to $\sim T_{\text{heat}}(n)$ of the $n$th cycle. Heating to temperatures below crystallization onset yields an increase of $\mu$ due to strain relaxation and, for these amorphous samples, a clear Hopkinson peak is observed at the Curie temperature of the amorphous phase. Heating above nanocrystallization onset leads to an irreversible rise of $\mu$ ascribed to formation of $\alpha$-Fe,Co phase. After cooling down nanocrystallized samples, for low crystalline volume fractions, $\mu$ at room temperature increases with respect to that of amorphous samples and reaches a maximum. Moreover, the Hop-
kinson peak is no longer observed but the fall of $\mu$ is continuous and smoother as $T_{\text{heat}}$ increases. As nanocrystallization progresses, the value of $\mu$ at room temperature decreases with respect to the maximum value achieved at very low crystalline fractions when the softest sample is obtained at room temperature. However, the value of $\mu$ at high temperature increases, leading to a smaller coefficient of the dependence of $\mu$ with temperature. Heating above the second transformation stage observed by DSC, an irreversible fall in $\mu$ is observed, ascribed to the formation of boride phases, which magnetically harden the material.

E. Thermal dependence of coercivity

Figure 7 shows the temperature dependence of coercive field, $H_C$, for nanocrystalline samples as a function of the degree of nanocrystallization. In agreement with initial permeability results, the softest sample is obtained for very low crystalline fraction, independent of the studied alloy. For these samples, $X_{\text{DSC}}=0.1$, a clear increase in $H_C$ is observed at elevated temperatures due to the ferroparamagnetic transition of the amorphous phase. As this phase becomes paramagnetic, the exchange coupling between nanocrystals is lost and only dipolar interactions prevent the system from becoming a set of isolated superparamagnetic particles. The anhysteretic superparamagnetic state is expected to be achieved at higher temperatures than those explored here. As nanocrystallization progresses, the observed relative increase in $H_C$, defined as $R_{H_C}=[H_C(T_{\text{max}})-H_C(300 \text{ K})]/H_C(300 \text{ K})$, where $T_{\text{max}}$ is the temperature at which the maximum value of $H_C$ is achieved, drastically drops from $X_{\text{DSC}}=0.1$ to 0.9 (from $R_{H_C} \sim 1800\%$ to $150\%$ and $170\%$ for the Zr and NbZr alloys, respectively, and from $R_{H_C} \sim 1100\%$ to $65\%$ for Nb alloy).

F. Saturation magnetization

Saturation magnetization values were obtained from magnetization curves, $M(H)$, using a VSM with a maximum applied field of 1.5 T. The linear behavior of $M(H)$ at high field ($H>0.5$ T) was extrapolated to $H=0$ and the saturation magnetization was measured as the intersection of this line with $H=0$. Figure 8(a) shows the values obtained at room temperature for the different alloys as a function of the nanocrystallization fraction. Figure 8(b) shows the relationship (close to linear) existing between specific magnetization, $\sigma_s$, and $\langle HF \rangle$, also observed for other amorphous and nanocrystalline systems.

A linear increase is observed in Fig. 8(a) as nanocrystallization progresses, which can be understood from a very simple two phase model described by the following expression:
the value corresponding to the fully amorphous system. Although this behavior has been reported for some nanocrystalline systems [e.g., (FeCr)_{73.5}Si_{11.5}B_{3}P_{11}Mo_{0.5}Cu_{1}] (Ref. 29)], the opposite behavior is also reported for similar alloys [e.g., Fe_{68.5}Co_{10}Ge_{2}Zr_{13}B_{9}Cu_{11} (Ref. 30), Fe_{68.5}Zr_{13}B_{9}Cu_{11} (Ref. 31)]. On the other hand, an initial rise in Curie temperature for samples heated up to the early stages of nanocrystallization followed by a continuous decrease of the Curie temperature of the amorphous phase as nanocrystallization progresses is also observed in some nanocrystalline systems [e.g., Fe_{68.5}Co_{10}Ge_{2}Zr_{13}B_{9}Cu_{11}, Fe_{68.5}Mo_{13}Si_{1}, B_{3}P_{11}Mo_{0.5}Cu_{1}] (Ref. 32)]. In fact, this decrease is opposite to the expected increase due to the presence of nanocrystals with a high Curie temperature (α-Fe, Co), which should polarize the amorphous matrix. Compositional changes in the amorphous matrix, impoverishment in Fe and enrichment in Nb and/or Zr as the nanocrystallization progresses, seem to be more important effects and yield a reduction in the Curie temperature in the studied case.

This compositional effect could be clarified considering that the concentration of an element i in the amorphous matrix, \( C_{\text{am}} \), and in the crystalline phase, \( C_{\text{cryst}} \), are linked with the average concentration in the alloy, \( C_{\text{total}} \):

\[
C_{\text{total}}^{\text{am}} = X_{C} C_{\text{cryst}}^{7} + (1 - X_{C}) C_{\text{am}},
\]

where \( X_{C} \) is the fraction of the crystalline volume. \( X_{C} = f X_{\text{DSC}} \) (\( f = 0.54 \) for this alloy) and \( \sigma_{s}^{\text{cryst}} \) and \( \sigma_{s}^{\text{amorph}} \) are the specific magnetization of the crystalline and amorphous phases, respectively. Therefore, a linear increase of the total specific magnetization, \( \sigma_{s}^{\text{total}} \), with \( X_{\text{DSC}} \) is obtained with a slope \( s = f(\sigma_{s}^{\text{cryst}} - \sigma_{s}^{\text{amorph}}) \). Although the constant composition of the nanocrystals along the nanocrystallization process supports \( \sigma_{s}^{\text{cryst}} \) to be constant, this approach cannot be so easily assumed for \( \sigma_{s}^{\text{amorph}} \), as the composition of this phase continuously changes. Therefore, \( \sigma_{s}^{\text{amorph}} \) is rather considered as an average value. The results obtained from the linear fitting are \( \sigma_{s}^{\text{amorph}} = 132.4 \pm 0.5, 145.6 \pm 1.4 \), and \( 136.0 \pm 0.8 \) emu/g; and \( \sigma_{s}^{\text{cryst}} = 171 \pm 3, 168 \pm 7, \) and \( 164 \pm 4 \) emu/g for Nb, Zr, and NbZr alloys, respectively. In the case of \( \sigma_{s}^{\text{cryst}} \), changes among the different studied alloys are within the error bar as the expected composition of the crystalline phase developed in the three studied alloys is the same. However, this value is smaller than that of pure crystalline Fe at room temperature, 218 emu/g, ascribed to the nanocrystalline nature of the crystalline phase studied.

IV. DISCUSSION

A. Curie temperature of the amorphous phase

Curie temperature was measured from both TMG and permeability measurements. In each case, the derivative of the signal was calculated and the inflexion point was taken as indicative of the Curie temperature (Fig. 9). Both techniques agree describing a continuous decrease in the Curie temperature with the progress of the nanocrystallization process from the expected composition of the crystalline phase developed among the different studied alloys are within the error bar as being the maximum crystalline fraction achieved, \( f \).

FIG. 9. (Color online) Curie temperature as a function of the progress of nanocrystallization: (a) from TMG plots vs crystallization fraction and (b) from permeability measurements vs the thermal span from the crystallization onset temperature and the maximum temperature achieved during treatment. The inset of (a) corresponds to the data as a function of the enrichment in Nb/Zr and B in the amorphous matrix. Lines correspond to linear fitting of the data.

\[
\sigma_{s}^{\text{total}} = X_{C} \sigma_{s}^{\text{cryst}} + (1 - X_{C}) \sigma_{s}^{\text{amorph}},
\]

where \( X_{C} \) is the crystalline volume fraction, \( X_{C} = f X_{\text{DSC}} \) (\( f = 0.54 \) for this alloy) and \( \sigma_{s}^{\text{cryst}} \) and \( \sigma_{s}^{\text{amorph}} \) are the specific magnetization of the crystalline and amorphous phases, respectively. Therefore, a linear increase of the total specific magnetization, \( \sigma_{s}^{\text{total}} \), with \( X_{\text{DSC}} \) is obtained with a slope \( s = f(\sigma_{s}^{\text{cryst}} - \sigma_{s}^{\text{amorph}}) \). Although the constant composition of the nanocrystals along the nanocrystallization process supports \( \sigma_{s}^{\text{cryst}} \) to be constant, this approach cannot be so easily assumed for \( \sigma_{s}^{\text{amorph}} \), as the composition of this phase continuously changes. Therefore, \( \sigma_{s}^{\text{amorph}} \) is rather considered as an average value. The results obtained from the linear fitting are

\[
\sigma_{s}^{\text{amorph}} = 132.4 \pm 0.5, 145.6 \pm 1.4, \text{ and } 136.0 \pm 0.8 \text{ emu/g}; \quad \text{and} \quad \sigma_{s}^{\text{cryst}} = 171 \pm 3, 168 \pm 7, \text{ and } 164 \pm 4 \text{ emu/g for Nb, Zr, and NbZr alloys, respectively. In the case of } \sigma_{s}^{\text{cryst}}, \text{ changes among the different studied alloys are within the error bar as the expected composition of the crystalline phase developed in the three studied alloys is the same. However, this value is smaller than that of pure crystalline Fe at room temperature, 218 emu/g, ascribed to the nanocrystalline nature of the crystalline phase studied.}
of transformation. Therefore, a linear behavior of $(\mu_0 M_s H_C)^2$ versus $X_{DSC}^4$ is expected with a value at $X_{DSC}=0$ equal to $(p_C(K_d))^2$ and a slope:

$$m = p_C^2 f^2 \frac{K_0^2(D)}{A^6}.$$  

(9)

Figure 10 shows, as an example, this plot for Zr alloy using data obtained at different temperatures. For each measurement temperature, $T_m$, the expected linear behavior is fulfilled (regression factor $r>0.9$) when $T_m$ is below the Curie temperature of the amorphous phase of samples, $T_{C,am}$, with $X_{DSC}<0.9$. It is worth noting that linearity is not lost when $T_m$ is above $T_{C,am}$ for $X_{DSC}=0.9$.

A qualitative explanation could be given after the model developed by Hernando et al.,\(^\text{15}\) where $A$ is substituted by $\gamma A$, being

$$\gamma = e^{-\Lambda/L_{am}},$$  

(10)

being

$$\Lambda = \langle D \rangle \left( \frac{1}{X_C} \right)^{1/3} - \langle D \rangle,$$  

(11)

where $L_{am}$ is the exchange correlation length of the amorphous matrix and $\gamma$ is a parameter (from 0 to 1) that describes the ability of the amorphous matrix for transmitting the exchange interaction between nanocrystals. For low values of $D/L_{am}$, $\gamma$ is independent of crystalline fraction,\(^\text{15}\) which supports the linear fitting observed for low temperatures in Fig. 10. Increasing the temperature, $L_{am}$ decreases and its dependence on the crystalline fraction can no longer be neglected as the $D/L_{am}$ ratio increases (see Fig. 1 in Ref. 15), explaining the deviation from the linearity of the high temperature data of Fig. 10 for low crystalline fractions. Values of $\Lambda$ could also be very small for large values of crystalline volume fraction and small crystal size as those reported here (5 nm), leading to $\gamma\sim 1$. In fact, single phase model developed by Herzer only fails for low crystalline fractions close to the Curie temperature of the amorphous phase, where a maximum in coercivity is only explained by two phase models. At large crystalline volume fractions ($X_{DSC} = 0.6$ in the present paper), this maximum is not detected and the effect of the amorphous phase could be neglected considering that the nanocrystals are so close to each other that the exchange interaction is possible.

The slopes described by expression (9) were obtained for different temperatures and taking into account the linear temperature dependence of $K_1 = aT + b$, $[K_1$ decreases from 46 to 13 kJ/m$^3$ increasing $T$ from 273 to 673 K (Ref. 35)], values of the exchange stiffness constant can be obtained as a function of the temperature from

$$A(T) = p_C^2 f^2 \frac{(aT + b)^{4/3}(D)^2}{m(T)^{1/6}},$$  

(12)

Figure 11 shows $A$ as a function of the temperature, neglecting the effect of $f$ and $p_C$, as $(f^2 p_C)^{1/3} \sim 1$. The obtained values are almost composition independent, in agreement with the expected similar composition of the crystalline phase formed in all the studied alloys and are of the order of that of pure Fe ($A = 1.74 \times 10^{-11}$ J m$^{-1}$) (Ref. 28) and those of amorphous alloys.\(^\text{15}\) It is worth mentioning that these results were obtained without any free parameter but just using the experimentally measured values of coercivity, saturation magnetization, volume fraction, and crystal size obtained in this study and the thermal dependence of $K_1$.\(^\text{35}\) The observed continuous decrease in $A$ as temperature increases could also be qualitatively explained by the two phase model developed by Hernando et al.,\(^\text{15}\) for a given value of $X_C$ and $D$, $\gamma$ decreases continuously from 1 to 0 approaching the Curie temperature of the amorphous phase but its dependence on the crystalline fraction is out of consideration in this study, where results are obtained as an average from samples with different $X_{DSC}$. However, for $\langle D \rangle = 5$ nm and $L_{am} \sim 30$ nm, $\gamma$ changes $\sim 15\%$ from $X_C=0.1$ to 0.9.\(^\text{15}\)

The values of $\langle K_{1i} \rangle$ exhibit large error bars but following the assumptions done on the independence of this parameter...
on $X_{\text{DSC}}$, $p_C(K_C) \sim 2$ kJ m$^{-3}$ can be estimated without distinguishable difference between the three studied compositions.

V. CONCLUSIONS

The temperature dependence of magnetic properties has been studied for nanocrystalline Fe$_{80}$Co$_{10}$Cu$_{3}$B$_{15}$Nb$_x$Zr$_{1-x}$ ($x$=0, 3, 6) samples at different stages of devitrification. The size of the formed nanocrystals ($\sim$5 nm) is almost constant with nanocrystallization progress, supporting the instantaneous growth approximation to describe the kinetics of the transformation.

A good correlation could be found between the different parameters related with the transformed fraction: $X_{\text{DSC}}$, $A_{\text{cryst}}$, $\langle H_F \rangle$, and $\sigma_S$. Soft magnetic properties at room temperature are enhanced for low crystalline fraction samples, but deteriorate for samples with a high crystalline fraction. However, the thermal dependence of magnetic properties is reduced as nanocrystallization progresses, which is a positive factor for technical applications in a broad temperature range.

The Curie temperature of the residual amorphous phase decreases as nanocrystallization progresses, being higher for 6 at. % Zr alloy than for Nb containing alloys for a constant value of $X_{\text{DSC}}$.

The thermal dependence of the magnetic anisotropy can be interpreted in the framework of the extended two phase model of random anisotropy applied to nanocrystalline systems. Results yield the thermal dependence of the exchange stiffness constant with changing composition and microstructure without using any free parameter.

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