INTRODUCTION

Since the nanocrystalline Fe_{73.5}Si_{13.5}B_{9}Nb_{3}Cu_{1} alloy was first found by Yoshizawa in 1988, nanocrystalline soft magnetic alloys (Makino, 2012; Ohta & Yoshizawa, 2008; Suzuki et al., 1991a, 1991b; Willard et al., 1998) have attracted great attention due to their excellent soft magnetic properties such as high saturation magnetic flux density \( B_s \), high initial permeability \( \mu_i \), low coercivity \( H_c \) and low core loss \( P \). From then on, three typical nanocrystalline alloy systems have been developed under the trade name of Finemet (Hitachi Matels Ltd., Japan) (Yoshizawa et al., 1988), Nanoperm (MH&W International Corp., USA) (Suzuki et al., 1991a, 1991b) and Hitperm (Carnegie Mellon University, USA) (Willard et al., 1998). Among these alloys, Nanoperm exhibits high \( B_s \) of 1.5–1.7 T but contains expensive and easily oxidized elements (e.g. Zr or Hf). Hitperm possesses high Curie temperature but the \( H_c \) is up to 200 A/m and the large amount of additive element Co increases material cost. Compared to the above alloys, Finemet exhibits high \( \mu_i \) of 100,000, low \( H_c \) of 0.53 A/m, as well as convenient production process, thus has been widely applied in electronic components such as transformers, inductors, sensors, filters etc.

The soft magnetic properties of nanocrystalline alloys result from the uniform nanocrystal structure which effectively reduces the magneto-crystalline anisotropy (Herzer, 1990, 1992, 1997, 2005, 2013). In addition, it has been pointed out that magnetic field annealing can further optimize soft
magnetic properties because the uniaxial anisotropy ($K_u$) induced by magnetic field has been utilized for tailoring the shape of hysteresis loops (Herzer, 1997, 2005, 2013; Ito et al., 2007; Suzuki & Herzer, 2012). The excellent soft magnetic properties of nanocrystalline alloy, such as low $H_c$ and low core loss ($W$), contribute to energy conservation and environmental protection. On the other hand, inductance ($L$) is also an important parameter of nanocrystalline magnetic core for electronic components application. With larger $L$, the nanocrystalline magnetic core is more sensitive and effective impedance to the alternating current (AC) interference signal, which is more beneficial to application in signal screening, noise filtering and current stabilizing. However, there are few reports about the influence of magnetic field annealing conditions on magnetic properties of nanocrystalline magnetic core with high inductance. Thus, in this work, we investigated the magnetic properties variation for a certain size magnetic core under no field annealing (NFA) and transverse magnetic field annealing (TFA), respectively, with an essential prerequisite of obtaining or keeping high inductance. The relationship between microstructure and magnetic properties under TFA and NFA were also investigated.

MATERIALS AND METHODS

Amorphous ribbons with nominal composition of Fe$_{73.5}$Cu$_1$Nb$_3$Si$_{15.5}$B$_7$ were prepared by single roller melt spinning. The width and thickness were 10 mm and 20 μm, respectively. Thermal stability was evaluated by differential scanning calorimeter (DSC) under a flowing highly purified argon atmosphere with a heating rate of 0.67°C/s. Melt-spun ribbons were coiled into toroidal core with 20 mm outer diameter and 14 mm inner diameter. Toroidal core samples were annealed under vacuum to develop nanocrystalline structure by transverse magnetic field furnace and the magnetic field was adjusted by current intensity. The core samples were placed in furnace chamber to ensure magnetic field direction parallel to the width of ribbon. Microstructure of annealed samples was identified by X-ray diffraction (XRD) with Cu Kα radiation and transmission electron microscopy (TEM). Grain size was measured by Nano Measurer software (Fudan University, China) for statistical analysis. $L_x$ was measured at a high frequency of 100 kHz and voltage 0.3 V by inductance-capacitance-resistance digital bridge. $B_1$ was measured by a vibrating sample magnetometer under a maximum applied field of 800 kA/m. $H_c$ and maximum permeability ($\mu_m$) were measured by DC B-H loop tracer under a maximum applied field of 800 A/m. Core loss was measured using AC B-H analyzer. The value of inductance and magnetic properties were calculated by averaging three samples annealed in the same condition.

The industrial annealing process for nanocrystalline magnetic core included five stages: 1) heat the furnace from room temperature to 330°C; II) keep the furnace temperature at 330°C for 60 min, then heat the furnace to 480°C; III) keep the furnace temperature at 480°C for 60 min, then heat the furnace to annealing temperature ($T_a$); IV) keep the furnace temperature at selected $T_a$ for 60 min; V) cool down to room temperature with a cooling rate about 0.167°C/s. In this study, we developed the TFA process based on the industrial annealing technology by adding transverse field through different annealing stages. Four TFA conditions were carried out, named TFA1, TFA2, TFA3 and TFA4, respectively. TFA1: applying transverse field through annealing stages II, III, IV and V; TFA2: applying transverse field through annealing stages III, IV and V; TFA3: applying transverse field through annealing stages IV and V; TFA4: applying transverse field only through annealing stage V.

RESULTS AND DISCUSSION

In order to investigate the crystallization behavior, isothermal crystallization annealing was performed on Fe$_{73.5}$Cu$_1$Nb$_3$Si$_{15.5}$B$_7$ amorphous ribbon samples at different temperature. Fig. 1 shows XRD patterns of ribbon samples annealed for 10 min at 480°C, 550°C, 710°C, which are the temperatures lower than the first crystallization temperature $T_{x1}$, between $T_{x1}$ and the first crystallization peak temperature $T_{x1p}$, just higher than the second crystallization temperature $T_{x2p}$, respectively, as shown on the DSC curve in the insert. The XRD pattern of as-quenched ribbon is also shown for comparison. It is seen from the XRD patterns that the α-Fe(Si) nanocrystals precipitated in the amorphous matrix after annealed at 480°C, but the grain size is too tiny and the volume fraction is too small that only one diffraction peak is detected, which suggests the grain just start to nucleate from...
the amorphous matrix. Three characteristic diffraction peaks of α-Fe(Si) can be identified clearly after annealed at 550°C, which is extremely close to \( T_x \). In contrast, after annealed at 710°C for 10 min, several Fe-B, Fe-Nb compounds, such as (Fe,Si)_3B, Fe_2Nb and Fe_5SiB_2, precipitated and deteriorated soft magnetic property. So it is confirmed that \( T_x \) corresponds to the precipitation temperature of single α-Fe(Si) soft magnetic phase and \( T_{x2} \) is the crystallization temperature of (Fe,Si)_3B, Fe_2Nb and Fe_5SiB_2. The difference between \( T_x \) and \( T_{x2} \) is 170°C, reflecting the alloy exhibits a preferable thermal stability, which is advantageous for developing nanocrystalline alloys, as it is reported that a wide crystallization temperature difference is beneficial for the precipitation of nanocrystals with uniform structure, which can improve the soft magnetic properties (Fan et al., 2013; Makino et al., 1995). Combining the crystallization behavior results with the industrial manufacturing conditions, the annealing temperature \( (T_a) \) in this study are selected as 540°C, 550°C, 560°C, and 570°C, respectively, at which the precipitation of α-Fe(Si) is promoted while the precipitation of hard magnetic phases is inhibited (Hono et al., 1999). The annealing time is fixed as 60 minutes for fully crystallization in order to obtain an ultrathin nanocrystalline structure (Yoshizawa & Yamauchi, 1990).

Fig. 2 shows the dependence of \( L_a \) and \( H_s \) of nanocrystalline magnetic core samples on \( T_a \) under different isothermal annealing conditions. The value of as-quenched alloy is also shown for comparison. The isothermal annealing process, as sketched in the insert in Fig. 2, was designed based on the industrial annealing technology described in the experimental procedures, including five stages: I) heat the furnace from room temperature to 330°C, which is far below \( T_{x1}; \) II) keep the furnace temperature at 330°C for 60 min to ensure heat uniformity in furnace, then heat the furnace to 480°C; III) reserve the furnace temperature at 480°C for 60 min, which not only keeps inner and outer of the magnetic core samples heat uniformity, but also increases the density of α-Fe(Si) nucleates, then heat the furnace to \( T_a \) (540°C, 550°C, 560°C, 570°C, respectively); IV) keep the furnace temperature at each \( T_a \) for 60 min, to ensure samples fully crystallization; V) cooling to room temperature. It can be seen that \( L_a \) of the core is nearly zero without annealing, but significantly improved after annealing. The value of \( L_a \) increases with increasing \( T_a \) firstly and reaches its maximum of 14.7 μH at \( T_a = 550°C \), then decreases slightly along with increasing \( T_a \). Meanwhile, the \( H_s \) without annealing reaches as high as 10 A/m, but decreases to 1.1 A/m after annealed at 540°C for 60 min. When \( T_a \) rises to

Fig. 3. The dependence of \( L_a \) and \( H_s \) on (A) four transverse magnetic field annealing (TFA) conditions and (B) TFA1 with different current intensity, together with that of no field annealing (NFA) for Fe_{73.5}Cu_{1}Nb_{3}Si_{15.5}B_{7} nanocrystalline magnetic core. The insert in (A) is four TFA annealing conditions named TFA1, TFA2, TFA3, and TFA4, respectively.
550°C, \( H_c \) increases to 1.9 A/m. As \( T_a \) continues increasing to higher than 560°C, \( H_c \) decreases to nearly 0.8 A/m.

Considering the nanocrystalline magnetic core used in high frequency requires high \( L_s \), the \( T_a \) was fixed at 550°C. But the \( H_c \) under this annealing condition reached maximum value, so we tried to decrease the coercivity by TFA. Fig. 3A shows the dependence of \( L_s \) and \( H_c \) of nanocrystalline magnetic core on four TFA conditions, together with that of NFA. The four TFA conditions were carried out by applying transverse magnetic field through different annealing stage, respectively, which named TFA1, TFA2, TFA3, and TFA4 (detail described in the experimental procedures), as shown in the insert in Fig. 3A. The DC intensity in each TFA conditions are all set as 100 A to ensure a constant magnetic field. According to the results, the \( H_c \) of sample after NFA is 1.9 A/m. However, when applied transverse magnetic field, \( H_c \) decreases remarkably and exhibits the minimum value of 0.7 A/m under TFA1. Meanwhile, \( L_s \) almost remains the same value under four TFA conditions. Therefore, the optimum TFA mode is TFA1.

To investigate the dependence of \( L_s \) and \( H_c \) on different magnetic field, we adjusted DC current (\( I \)) from 60 A to 140 A under TFA1. The experimental results are shown in Fig. 3B. Similarly, \( L_s \) has a slight fluctuation variation, while \( H_c \) exhibits sensitive to the current intensity. When \( I \) is 60 A, the value of \( H_c \) is 1.8 A/m, just a little lower than that of NFA, which suggests the effect of magnetic field is extremely weak under this condition. As \( I \) increases to 80 A, \( H_c \) decreases rapidly to 0.85 A/m, and then shows a gradually decreasing tendency with increasing current intensity. It is clearly seen that \( H_c \) is as low as 0.6 A/m when \( I \) increases to 140 A. Hence, we find that at TFA1 condition, the current intensity larger than 80 A is effective for decreasing the \( H_c \). The reason is that uniaxial anisotropy \( K_u \) induced by transvers field annealing dominates over other anisotropies and reduces the coercivity (Herzer, 1997; Suzuki & Herzer, 2012). This is the result we are looking for, that is, not only effectively reducing the \( H_c \), but also maintaining the high \( L_s \).

XRD analyses were performed on Fe\(_{73.5}\)Cu\(_1\)Nb\(_3\)Si\(_{15.5}\)B\(_7\) nanocrystalline magnetic core samples under NFA condition and TFA1 with different current intensity (results are not shown here). Three characteristic diffraction peaks corresponding to (110), (200), (211) plane, respectively, of \( \alpha \)-Fe(Si) phase can be identified for all the NFA and TFA1 samples and there is almost no obvious difference for each sample from the XRD patterns. The average grain size (\( D \)) calculated based on (110) plane with Scherrer’s equation was 13 nm for all NFA and TFA1 samples. To further understand the microstructure, bright-field TEM and selected area electron diffraction (SAED) analyses were carried out on NFA and TFA1 sample (\( I=140 \) A), as shown in Fig. 4. The grain distributions are quite uniform and the volume fractions of nanocrystals are pretty high, which results in identical ultrafine microstructure feature. The diffraction patterns of these two samples from SAED are almost the same. To acquire accurate description of grain size, statistical analysis was carried out by analyzing more than 200 grains for each sample. It is found that the average grain size of NFA and TFA1 (\( I=140 \) A) samples are 13 nm and 12 nm, which is consistent with XRD results. However, the grain size of NFA sample mainly range in 10~16 nm, while most nanocrystals of TFA1 (\( I=140 \) A) sample are measured as 12 nm. What is more, the mean square deviation (\( \sigma \)) is 1.80 for TFA1 (\( I=140 \) A) sample, which is much smaller than that of NFA sample (\( \sigma_{\text{NFA}}=2.46 \)), reflecting a more uniform nanocrystalline structure. Recently, it was found that
the crystallization reaction rate of Fe-based nanocrystalline (Migliaferini et al., 2015; Onodera et al., 2015) is increased by applying extra high magnetic field annealing. It is suggested that the nucleation activation energy for primary crystal phase is decreased under field annealing, which results in the increase of nucleation rate and density, whereas the crystallization mechanism is not changed (Fujii et al., 2008; Onodera et al., 2015), so the grain size distribution is more uniform. That is why the mean square deviation of TFA1 ($I=140$ A) sample is less than that of NFA.

Fig. 5 exemplifies the characteristics hysteresis loops under NFA and TFA1 ($I=60$ A, 80 A, 140 A) conditions. Obviously, flat loops are obtained after TFA and that the bigger current intensity is, the longer and narrower hysteresis loop becomes. Flat loops means lower magnetic remanence ratio and coercivity and more difficult to be saturated with increasing magnetization, indicating the improvement of anti-DC bias properties of nanocrystalline magnetic core products. Fig. 6 summarizes the dependence of specific magnetic properties ($B_s$, $\mu_m$ and $H_c$) and grain size ($D$) on DC current intensity ($I$) for Fe$_{73.5}$Cu$_1$Nb$_3$Si$_{15.5}$B$_7$ nanocrystalline magnetic core samples. The saturation magnetic flux density ($B_s$) almost stays invariant with current change. It is reported that $B_s$ can be calculated by equation (Ohta & Yoshizawa, 2007)

![Fig. 5. The characteristics hysteresis loop for Fe$_{73.5}$Cu$_1$Nb$_3$Si$_{15.5}$B$_7$ nanocrystalline magnetic core under no field annealing (NFA) and transverse magnetic field annealing 1 (TFA1) with direct current were 80 A and 140 A, respectively.](image)

![Fig. 6. The dependence of magnetic properties ($B_s$, $\mu_m$ and $H_c$) and grain size ($D$) on current intensity for Fe$_{73.5}$Cu$_1$Nb$_3$Si$_{15.5}$B$_7$ nanocrystalline magnetic core.](image)
\[
B_s = B_{sc} V/V + B_{am} V/V
\]

where \( V/V \) and \( V/V \) stand for the volume fraction of the crystalline phase and the amorphous phase, and \( B_{sc} \) and \( B_{am} \) are the saturation magnetic flux densities in the crystalline and amorphous phases, respectively. The constant value of \( B_s \) indicates magnetic field annealing does not influence the final volume fraction of crystalline phase for any NFA and TFA1 samples.

Meanwhile, the maximum permeability \( (\mu_m) \) and \( H_c \) shows similar variation tendency with obvious reduction under TFA conditions and reaches the minimum value of 34,000 and 0.6 A/m at \( f \) was 140 A, respectively. While \( \mu_m \) and \( H_c \) shows great decrease with increasing current density from 0 A to 80 A, the mean grain size \( (D_{sc}) \) has no obvious difference which is about 13 nm for all samples. The result seems to be contrary to the well-known random anisotropy model proposed by Herzer in 1997, 2005, 2013, which suggests that when crystal size is smaller than exchange correlation length \( (D<<L_{sc}) \), the magneto-crystalline anisotropy \( K_u \) is essentially reduced by exchange interaction. The overall average anisotropy \( <K> \) is approximated by \( <K>=|K_u| x^6 \)

\[
<K>=|K_u| x^6
\]

where \( x \) denotes the crystalline volume fraction, \( L_o \) denotes natural ferromagnetic exchange length and \( L_o \approx 20\sim40 \) nm for Fe-based alloys. Since \( H_c \) is proportional to \( <K> \) and permeability \( \mu \) is proportional to \( <K>^{-1} \), hence \( H_c \propto D^6 \) while \( \mu \propto D^6 \). The fact is that the basic anisotropy \( <K> \) is attributed to the magneto-crystalline anisotropy \( K_u \), magnetic field induced anisotropies \( K_u \) and magneto-elastic anisotropy \( K_v \).<K> is mainly determined by \( K_u \) under NFA, and equation (2) is valid only under this condition. However, when transverse magnetic field is applied, uniaxial anisotropy \( K_u \) induced with axis parallel to the direction of magnetic field, dominates over other anisotropies and wide transverse slab domains are formed, which largely reduce the \( H_c \) and \( \mu_m \) (Flohrer & Herzer, 2010; Herzer, 2013; Suzuki & Herzer, 2012). Herzer also has proved that the major contribution to \( H_c \) under magnetic field annealing arises from spatial fluctuations of the anisotropy only and the simpler domain configuration (Herzer, 1995). Furthermore, there is a slightly reduction of \( H_c \) and \( \mu_m \) when the current intensity increases from 80 A to 140 A, corresponding to the variation of the flat loop as shown in Fig. 6, which indicates that the \( H_c \) and \( \mu_m \) are closing to the limit value with the increasing magnetic field under TFA1 condition.

Core loss is extremely important for nanocrystalline magnetic core applied for high frequency electronic components. Fig. 7 shows the dependence of core loss \( (P) \) on frequency \( (f) \) at 0.2 T for Fe_{73.5}{\text{Cu}}_{1}{\text{Nb}}_{3}{\text{Si}}_{15.5}{\text{B}}_{7} \) nanocrystalline magnetic core. \( P \) increases gradually with increasing \( f \). The increasing rate is small when \( f \leq 20 \) kHz but accelerate rapidly after that. However, samples under TFA1 condition show obviously slower increasing tendency and the stronger of current intensity is, the slower tendency is. The core loss for TFA1 \( (I=140 \) A) sample at 20 kHz is 1.6 W/kg, which is significantly less than that of commercial Finemet alloy of 2.1 W/kg (Yoshizawa et al., 1988). The advantage of low core loss at high frequencies such as 80 kHz and 100 kHz compared to those of NFA magnetic core is extremely obvious. Therefore, the combination of high inductance, low core loss, as well as excellent soft magnetic properties promises application of this nanocrystalline magnetic alloy wide application prospect for high frequency transformers of switched power supplies and interface transformers in telecommunication network.

CONCLUSIONS

Magnetic properties of Fe_{73.5}{\text{Cu}}_{1}{\text{Nb}}_{3}{\text{Si}}_{15.5}{\text{B}}_{7} \) nanocrystalline magnetic core under NFA and TFA conditions aiming at decreasing the coercivity \( (H_c) \) while keeping the high inductance \( (L) \) were investigated. The main results are summarized as follows: 1. Fe_{73.5}{\text{Cu}}_{1}{\text{Nb}}_{3}{\text{Si}}_{15.5}{\text{B}}_{7} \) nanocrystalline magnetic core annealed with transverse magnetic field applied during the whole annealing process with annealing temperature 550°C (TFA1) exhibits the lowest \( H_c \) while keeping high \( L_s \) compared other annealing conditions. \( H_c \) also shows remarkable decrease with increasing DC intensity under TFA1 condition while \( L_s \) keeps almost the constant value. 2. The average grain size and crystallization degree of the nanocrystalline magnetic core under NFA and TFA1 conditions exhibit no obvious difference, but the latter shows a more uniform grain size distribution. 3. Uniaxial anisotropy \( K_u \) induced in transverse magnetic field, flattens the hysteresis.
loop greatly and lowers the maximum permeability and coercivity by dominating over other anisotropies. 4. The nanocrystalline magnetic core under TFA1 with \( J=140 \) A exhibits high inductance of 13.8 \( \mu \)H at 0.3 V at 100 kHz and excellent magnetic properties with low coercivity of 0.6 A/m and low core loss (\( W \) at 20 kHz) of 1.6 W/kg under flux density of 0.2 T.

**CONFLICT OF INTEREST**

No potential conflict of interest relevant to this article was reported.

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