Effect of Annealing on the Magnetic Properties of FeCoNiCuNbSiB Soft Magnetic Alloys

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A series of nanocrystalline soft magnetic alloys with nominal compositions of Fe66.8-xCo10Ni10Cu0.8Nb2.9Si11.5B8 (x = 1–15 at%) were developed and studied. Effects of annealing on the soft magnetic properties, crystallization behavior, and domain structure were investigated. The alloys with higher Ni content were prone to exhibit stronger magnetic anisotropy. The Fe66.8Co10Ni10Cu0.8Nb2.9Si11.5B8 alloy exhibited excellent soft magnetic properties, including the low permeability of 2000, low coercivity of about 0.6 A/m, and low remanence of 2.4 mT, together with a temperature gap of 128 K between two crystallization onset temperatures. It has been found that the Ni content and the annealing process possess significant effects on the soft magnetic property of the nanocrystalline alloys. It shows that the developed Fe66.8Co10Ni10Cu0.8Nb2.9Si11.5B8 nanocrystalline alloy exhibits great potentials for applying in the field of common mode chokes or current transformers, due to its ability to resist the direct current.

Keywords: nanocrystallization, annealing, microstructure, soft magnetic properties, magnetic domain

INTRODUCTION

FINEMET-type FeCuNbSiB nanocrystalline alloys exhibit excellent comprehensive soft magnetic properties including high saturation magnetic induction $B_s$, high permeability $\mu$, and low core loss $P$ (Yoshizawa et al., 1988a). Furthermore, since FINEMET-type FeCuNbSiB nanocrystalline alloys can be easily prepared in air at low cost, they are widely used in electronic fields such as high-frequency power transformers, common mode chokes, magnetic amplifiers, and current transformers (Yoshizawa et al., 1988b; Petzold, 2002; Hasegawa, 2004; Herzer et al., 2005; Hasegawa, 2006; Herzer, 2013). Nowadays, the devices used in electronic and information fields, however, are prone to higher power and higher frequency. Then, it is necessary to develop soft magnetic materials with better permeability properties in high-frequency range and good DC resistance ability.

It is known that the performance of permeability vs. frequency in high-frequency regions could be enhanced by a large anisotropy constant $K_u$ when the coercivity property does not deteriorate. This results in the decreases in permeability attenuation with frequency. This characteristic makes the design of the LCR rectifier circuit become easier when the material is processed into an inductor. The anisotropy constant $K_u$ could be evaluated by Herzer et al. (2011)

$$K_u = \frac{H_u I_s}{2}$$

(1)
where $H_k$ is the magnetic anisotropy field and $J_s$ is the saturation magnetic polarization. As indicated by Eq. 1, the magnetic anisotropy constant $K_u$ can be improved by increasing $J_s$. Nevertheless, despite there being myriad works focusing on developing materials with high saturation magnetic polarization (Makino et al., 1995; Ohta and Yoshizawa, 2007; Makino et al., 2009; Kubota et al., 2011; Urata et al., 2011; Chen et al., 2013; Li et al., 2014; Li et al., 2015; Shi and Yao, 2020; Shi et al., 2021), few of these materials can meet the requirements of industrial mass production. Moreover, the increase in magnetic anisotropy constant $K_u$ is very limited.

For FINEMET-type alloys, the magnetic field-induced anisotropy is very weak (Herzer, 1994a). However, it was found that the creep could induce the anisotropy in FINEMET-type alloys (Herzer, 1994b). It shows a $K_u$ value as large as 1,000 J/m$^3$ can be achieved after the FINEMET-type alloys are annealed with a 100-MPa tensile stress applied along the ribbon axis. However, due to the low productivity, inducing anisotropy via creep is impractical for industrial production.

For nanocrystalline alloys processed by magnetic field annealing, adding other ferromagnetic elements such as Co and Ni is considered to be conducive to the induction of in-plane anisotropy (Fujimori et al., 1977; Ohnuma et al., 2003; Yoshizawa et al., 2003). However, the crystallization process would be significantly influenced after the addition/replacement of Co and Ni to Fe-based amorphous alloys (Agudo and Vázquez, 2005). The increment of the Ni content leads to the rise of the second crystallization peak temperature of the amorphous alloy rise, resulting in the deterioration of the thermal stability and soft magnetic property of the alloy. Then, understanding the factors affecting the crystallization behavior and soft magnetic property of the nanocrystalline alloy is important.

Here, effects of Ni addition and annealing process on the structure and properties of FeCoNiCuNbSiB soft magnetic alloys were investigated, and an Fe$^{66.8}_{66.8}$Co$^{10}_{10}$Ni$^{10}_{10}$Cu$^{0.8}_{0.8}$Nb$^{2.9}_{2.9}$Si$^{11.5}_{11.5}$B$^{8}_{8}$ nanocrystalline alloy with good soft magnetic property and large resistance to direct current in the high-frequency region was obtained.

**EXPERIMENTAL PROCEDURE**

Alloy ingots with a nominal composition of Fe$^{66.8-}_{66.8}$Co$^{10}_{10}$Ni$^{10}_{10}$Cu$^{0.8}_{0.8}$Nb$^{2.9}_{2.9}$Si$^{11.5}_{11.5}$B$^{8}_{8}$ ($x = 1, 5, 10, 15$ at%) were prepared by induction melting in a high-vacuum condition. The raw materials were industrial-purity metals of Fe (99.9 mass%), Co (99.9 mass%), Ni (99.9 mass%), Cu (99.9 mass%), Si (99.9 mass%), pre-alloyed Nb-Fe (99.7 mass%), and Fe-B (99.7 mass%). The amorphous ribbons were manufactured by the single-roll rapid solidification method. The ribbon width is 30 mm, and thickness is 22 μm. The ribbons were slit to 6 mm in width. Toroidal cores with the outer diameter of 29 mm and inner diameter of 21 mm were wounded by these slit ribbons then. These cores were annealed at different temperatures in a furnace with 2 L/min flowing argon; meanwhile, a 400-Oe transverse magnetic field was applied. The transverse field means the direction is parallel to the axis of rotation.

The microstructure of as-cast and annealed ribbons was analyzed by X-ray diffractometry (XRD, Bruker D8 Advance) with Cu K radiation ($\lambda = 0.15406$ nm). The thermodynamic characteristics of the ribbons were analyzed by the differential scanning calorimeter (DSC, NETZSCH 404C) at a heating rate of 10 K/min under high-purity argon flow. The B-H loops were measured by a static magnetic property analyzer (MATS-2010SD). The core loss property of the cores was analyzed by the IWATSU B-H Analyzer. The test frequency range was 10–100 kHz, sinusoidal type. The maximum flux density values were 100, 200, and 300 mT, respectively. Images of the magnetic domain structures for samples with different compositions were captured using the magneto-optical Kerr microscope (Zeiss A × 10). All measurements were performed at room temperature.

**RESULTS AND DISCUSSION**

The XRD patterns of the as-cast Fe$^{66.8-}_{66.8}$Co$^{10}_{10}$Ni$^{10}_{10}$Cu$^{0.8}_{0.8}$Nb$^{2.9}_{2.9}$Si$^{11.5}_{11.5}$B$^{8}_{8}$ ($x = 1, 5, 10, 15$ at%) ribbons are shown in Figure 1. Except for the broad peak corresponding to the amorphous phase, no crystallization peak can be found in the XRD patterns, indicating that the microstructure of the as-cast ribbons is amorphous.

The DSC curves of the as-cast Fe$^{66.8-}_{66.8}$Co$^{10}_{10}$Ni$^{10}_{10}$Cu$^{0.8}_{0.8}$Nb$^{2.9}_{2.9}$Si$^{11.5}_{11.5}$B$^{8}_{8}$ ($x = 1, 5, 10, 15$ at%) ribbons are shown in Figure 2. The first and second crystallization onset temperature $T_{x1}$ and $T_{x2}$ and the first and second crystallization peak temperature $T_1$ and $T_2$ are indicated by arrows on the curves. We can find three distinguishing characteristics in the DSC curves. Firstly, $T_2$ decreases gradually with the increase in Ni content, while $T_1$ stays almost unchanged. The temperature gap $\Delta T_2$ ($T_{x2}-T_{x1}$) shrinks down from 165 to 119 K with the increase in Ni content. However, the interval $\Delta T_2$ for the alloys with $x = 5, 10, 15$ are almost the same. This value agrees to the previous work by Agudo in Ref. (Agudo and Vázquez, 2005). Secondly, there is...
an exothermic-peak-like region between $T_{x1}$ and $T_{x2}$ at about 828 K for the alloys with $x/10, 15\%$; this region is considered to be an overlap of the two stages of crystallization. It means that the alloys of Fe$_{56.8}$Co$_{10}$Ni$_{10}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_{8}$ and Fe$_{51.8}$Co$_{10}$Ni$_{15}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_{8}$ tend to get into the second crystallization process much more easily. Also, the stability of the first-stage crystallization phase is weaker. This might be the reason for the $\Delta T_{x}$ of the 10% Ni alloy being a little larger than 5% Ni alloy. Values of $T_{x1}$, $T_{x2}$, $T_1$, $T_2$, and $\Delta T_{x}$ of the as-cast Fe$_{66.8-x}$Co$_{10}$Ni$_{x}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_{8}$ alloys are summarized in Table 1.

The cores were annealed at 838 K with transverse magnetic field for 15, 30, and 45 min, respectively. The $B-H$ loops of the cores are shown in Figure 3. The hysteresis loops of the annealed alloys show good linearity. With the increment of Ni content, the anisotropic field $H_A$ rises from 200 to 380 A/m. The values of the magnetic anisotropy constant $K_u$ can be obtained from Eq. 1.

During the calculation of $K_u$ $J_s$ was estimated using the values of $B_{800}$ ($B_{800}$ denotes the magnetic induction intensity of alloys under an applied magnetic field of 800 A/m), and the $B_{800}$ values are 1.26, 1.15, 1.11, and 1.04 T for the alloys of Ni = 1%, Ni = 5%, Ni = 10%, and Ni = 15%, respectively. When the Ni content increases to 15%, the alloy exhibits a lower permeability and larger magnetic anisotropy field $H_A$. However, the coercivity of the alloy is significantly increased, and the linearity of the hysteresis loop gets worse. The coercivity ($H_c$), remanence ($B_r$), initial permeability ($\mu_i$), and anisotropy constant ($K_u$) properties are shown in Table 2. With the prolongation of annealing time, the permeability of the alloy gradually decreases. The strength of the magnetic anisotropy field becomes stronger. Moreover, the ability to resist the external magnetic field is enhanced. The $H_c$ and $B_r$ for the Fe$_{66.8-x}$Co$_{10}$Ni$_{x}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_{8}$ alloy decrease with the increasing annealing time. The $H_c$ and $B_r$ of Fe$_{61.8}$Co$_{10}$Ni$_{5}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_{8}$ alloy increase firstly and then get smaller with the increase in heat treatment time. $H_c$ and $B_r$ increase definitely with the heat treatment time when the Ni content is high (x = 10, 15%). When the Ni content reaches 15%, the coercive $H_c$ and the remanence $B_r$ increase greatly. When the heat treatment time increased from 15 to 45 min, $H_c$ and $B_r$ increased from 15.81 A/m and 28 mT to 163.8 A/m and 330 mT, respectively. With the increase in Ni content, the relative initial
permeability $\mu_i$ decreases and the magnetic anisotropy $K_u$ is enhanced.

The XRD patterns of Fe$_{66.8-x}$Co$_{10}$Ni$_{x}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_8$ ($x = 1, 5, 10, 15$ at\%) alloy ribbons which annealed at 838 K for 30 min are shown in Figure 4. The grain size of the crystalline phase was estimated using the Scherrer formula as follows (Scherrer, 1918),

$$D = \frac{K\lambda}{\beta\cos\theta}$$

(2)

where $D$ is the grain size, $K$ is the Scherrer constant ($\sim 0.89$), $\lambda$ is the wavelength of the X-ray, $\beta$ is the full width at half maximum (FWHM) of the diffraction peak, and $\theta$ is the Bragg angle. The crystallized volume fraction phase can be estimated by using the following equation (Langford, 1978; Verdon et al., 1998; Zhang et al., 1998; Cerqueira et al., 2000; Aronin et al., 2010):

$$V_{cr} = \frac{I_{cr}}{I_{cr} + I_{am}}$$

(3)

TABLE 2 | Magnetic properties of Fe$_{66.8-x}$Co$_{10}$Ni$_{x}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_8$ ($x = 1, 5, 10, 15$ at\%) alloy cores.

| Alloys  | 15 min | 30 min | 45 min |
|--------|--------|--------|--------|
| $H_c$ (A/m) |       |        |        |
| $x = 1$ | 7.93   | 0.54   | 0.40   |
| $x = 5$ | 3.21   | 0.41   | 0.65   |
| $x = 10$| 0.21   | 0.61   | 0.84   |
| $x = 15$| 15.81  | 19.41  | 163.8  |
| $B_r$ (mT) |       |        |        |
| $x = 1$ | 55.71  | 30.55  | 23.92  |
| $x = 5$ | 31.91  | 23.41  | 23.88  |
| $x = 10$| 2.35   | 2.35   | 7.44   |
| $x = 15$| 27.6   | 63.2   | 329.5  |
| $\mu_i$ |       |        |        |
| $x = 1$ | 3,300  | 5,100  | 5,400  |
| $x = 5$ | 3,050  | 2,940  | 2,660  |
| $x = 10$| 2,290  | 2,000  | 2,230  |
| $x = 15$| 1,310  | 810    | 580    |
| $K_u$ (J/m$^3$) |       |        |        |
| $x = 1$ | 108    | 126    | 129    |
| $x = 5$ | 163    | 170    | 174    |
| $x = 10$| 204    | 225    | 200    |
| $x = 15$| 232    | 257    | 355    |

FIGURE 3 | B–H loops of Fe$_{66.8-x}$Co$_{10}$Ni$_{x}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_8$ ($x = 1, 5, 10, 15$ at\%) cores annealed at 838 K for 15 min, 30 min, and 45 min.

FIGURE 4 | XRD patterns of Fe$_{66.8-x}$Co$_{10}$Ni$_{x}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_8$ ($x = 1, 5, 10, 15$ at\%) ribbons annealed at 838 K for 30 min.
and the crystallized volume fraction \( V_{cr} \) corresponding quantitative correlation can be expressed as below:

\[
\tan \theta = \frac{\tan \theta_0}{1 + \frac{V_{cr}}{L_0}}
\]

Anisotropy is positively correlated with grain size. The magnetic anisotropy is mainly determined by the grain size \( \langle K \rangle \).

During the amorphous-to-crystalline transformation process, \( \alpha-(Fe, Ni) \) atoms partially substitute the crystallographic positions of \( \alpha-(Fe, Co, Ni) \) atoms in the precipitated crystalline phases, leading to the formation of \( \alpha-(Fe, Ni), \alpha-(Fe, Co), \) and \( \alpha-(Fe, Co, Ni) \) phases with a bcc structure. Although there is an overlap area between the two crystallization temperatures at about 833 K, there is no other phase but bcc. Maybe the content of the possibly existing phase is very small that we cannot find it through XRD results. The crystallized volume fraction is about 25%, and the grain size is about 11–15 nm. The main diffraction peak positions \((\theta)\) in XRD patterns for the \( x = 1, 5, 10, 15 \) at% alloys are 44.980°, 45.000°, and 44.960° respectively. The magnetic anisotropy was enhanced then. The rise in \( H_c \) of soft magnetic alloys is mainly due to the increment in grain size.

The XRD patterns of the \( x = 15 \) alloy ribbons annealed at 838 K for 15, 30, and 45 min are shown in Figure 5. The XRD patterns for the \( x = 1, 5, 10, 15 \) at% alloys are 44.92°, 44.94°, 45.03°, and 45.00° respectively. Therefore, the drastic increase in coercivity is not caused by the difference in the degree of lattice distortion, either.

The theory proposed by Luborsky (Luborsky et al., 1976) suggests that the origin of the induced anisotropy \( K_u \) in the amorphous alloys is the atomic pairing orientation, which is the same as that of the crystalline material. Since the crystallization process is a thermal activation process, the coupling orientation of \( (Fe, Ni) \) atoms would be enhanced as the annealing time increases. The magnetic anisotropy was enhanced then. The rise in \( H_c \) could be ascribed to the enhanced magnetic anisotropy or the change in magnetic domain structure after the enhancement magnetic anisotropy.

The permeability vs. frequency curves of the \( x = 1, 5, 10, 15 \) at% alloy cores which were annealed at 838 K for 30 min are shown in Figure 6. With the increase in Ni content, the initial permeability of the alloy becomes lower and the magnetic anisotropy is enhanced. The decay of relative initial magnetic
permeability became slower with the increase in frequency. The permeability of the alloy with a Ni content of 10% is about 2,000, and the magnetic field induced \( K_u \) is about 225 J/m\(^3\). The permeability attenuation of these alloys was 47.8%, 35.2%, 19.2%, and 10.2% from 10 to 1,000 kHz for the alloy with Ni content of 1%, 5%, 10%, and 15%, respectively. These results agree with the previous work by Ohnuma in Ref. (Herzer, 1994b) that, with the increase of \( K_u \), the alloy exhibits more excellent permeability-vs.-frequency characteristics in the high-frequency band.

The domain structures of Fe\(_{66.8-x}\)Co\(_{10}\)Ni\(_x\)Cu\(_{0.8}\)Nb\(_{2.9}\)Si\(_{11.5}\)B\(_8\) (\( x = 1, 5, 10, 15 \) at%) alloy ribbons annealed at 838 K for 30 min are shown in Figure 7. The double arrow line in Figure 7 denotes the width direction of the ribbon, which is parallel to the direction of the transverse magnetic field during annealing. The single arrow line is parallel to the orientation of the polarized light in the magneto-optical Kerr microscope. The light-colored magnetic domains are oriented in the same direction as the polarized light, while the dark magnetic domains are in the opposite direction of the polarized light. It can be concluded from the figure that the size of the magnetic domain decreases when the magnetic anisotropy intensity becomes stronger with the increase in the Ni content. The parallelism between the magnetic domain and the direction of the applied magnetic field becomes stronger. The local magnetic domain structure becomes more uniformly distributed. From the perspective of technical magnetization, the magnetic domain structure inside the material because the sum of the different energies tends to be minimized. Moreover, the corresponding energies can be categorized as the exchange energy, the magneto-elastic energy, the magneto-crystalline anisotropy, the demagnetizing field energy, and the applied magnetic field energy. This result agrees with the work by Herzer (Herzer et al., 2011) that combines the domain structure and creep-induced \( K_u \).

The changing process of \( x = 10\) alloy’s domain structure during the whole magnetization process is shown in Figure 8. The direction distribution and test method of the magnetic domain structure photo are the same as those in Figure 7. The magnetic domain structure in each magnetized state is shown in the photograph indicated by the arrow in the figure. The magnetic domain of the alloy ribbon is fine 90° structure, and the whole magnetization process is basically a reversible movement of the domain walls. According to the magnetization theory of soft magnetic materials, the magnetization process can be separated into five steps (Tahara and Sugeno, 1973; Iványi, 1997). First is the initial magnetization (\( \mu_i \)), or the so-called linear magnetization region, where \( B \) varies linearly with \( H \). Second is the Rayleigh region, where the magnetization process is no longer linear. Third is the maximum permeability region, where \( B \) increases sharply with \( H \). Fourth is the near saturation region. Fifth is the paramagnetic region. In the magnetization process of Fe\(_{61.8}\)Co\(_{10}\)Ni\(_5\)Cu\(_{0.8}\)Nb\(_{2.9}\)Si\(_{11.5}\)B\(_8\) and Fe\(_{56.8}\)Co\(_{10}\)Ni\(_{10}\)Cu\(_{0.8}\)Nb\(_{2.9}\)Si\(_{11.5}\)B\(_8\) alloy ribbons, the first region is very big, and the second to fourth regions become very small. So, the magnetization process is kept linear until saturation. For the Fe\(_{51.8}\)Co\(_{10}\)Ni\(_{15}\)Cu\(_{0.8}\)Nb\(_{2.9}\)Si\(_{11.5}\)B\(_8\) alloy ribbon, more Ni elements may be entrapped in the \( \alpha-(Fe) \) and \( \alpha-(Fe)_3Si \) crystallization phases to replace the Fe atom position during the heat treatment. It will result in a more serious lattice distortion, so the energy of the alloy system increases. In addition, the addition of the Ni element increases the number of magnetic interaction (Fe, Ni) atom pairs in the alloy and enhances the magnetic anisotropy. In order to reduce the energy

**FIGURE 7** | (A) alloy of x = 1%; (B) alloy of x = 5%; (C) alloy of x = 10%; (D) alloy of x = 15%.
of the alloy, the magnetic field size of the soft magnetic alloy becomes smaller. This makes the B–H loop of the alloy with strong magnetic anisotropy show poor linearity, and its coercivity and remanence are larger.

Since the soft magnetic properties of the Fe51.8Co10Ni15Cu0.8Nb2.9Si11.5B8 alloy are very poor, some further heat treatment was applied to the cores. The deterioration of the coercivity property was originated in the distortion of the lattice. It was assumed that the distortion could be reduced by the decrease in the crystallization degree. We chose a temperature that is higher than the crystallization temperature and definitely lower than the overlap region. Moreover, some other temperatures that are lower than the crystallization temperature were set. The heat treatment temperatures were 783, 683, and 643 K then, and the heat treatment time was 15 min. The heat treatment was finished when the furnace was cooled to 473 K, and the transverse magnetic field of 400 Oe was applied. The XRD patterns of the Fe51.8Co10Ni15Cu0.8Nb2.9Si11.5B8 alloy ribbons after heat treatment are shown in Figure 9. The crystallization temperature of the alloy is 750 K, and the first crystallization peak temperature is 777 K. Some α-Fe-crystallized phases were formed after the heat treatment at 783 K for 15 min. After
calculation, the crystallization ratio is less than 5%. Both 683 and 643 K were subjected to stress relief annealing, and the alloy ribbons possess a typical amorphous structure after heat treatment.

The B–H loops of these Fe$_{51.8}$Co$_{10}$Ni$_{15}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_{8}$ alloy are shown in Figure 10. The $\mu$, $H_c$, and $B_r$ properties are listed in Table 3. With the increase in heat treatment temperature, the linearity of the B–H loop gradually increased; meanwhile, the remanence and coercivity decreased. When the heat treatment temperature rises from 643 to 783, the coercivity of the alloy decreases from 7.29 to 3.63 A/m, and the remanence is reduced from 250.9 to 12.2 mT. However, when the heat treatment temperature increased to 838 K, the coercivity of the alloy increased to 15.81 A/m, and remanence increased to 27.6 mT. It seems that the trace crystallization is beneficial to obtain excellent soft magnetic properties for the Fe-based alloy with high content of Ni. To obtain better soft magnetic properties, it is necessary to resist the direct current is necessary.

**SUMMARY**

The crystallization behavior, microstructure, soft magnetic properties, and domain structure for Fe$_{66.8-x}$Co$_{10}$Ni$_{x}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_{8}$ alloys with the Ni content ranging from $x$ = 1 to $x$ = 15 have been systematically studied, and results are summarized as follows:

1. The increment of the Ni content results in the decline of the crystallization temperature of the secondary crystallization phase and decreases the thermal stability of the α-Fe phase.

2. With the increase in Ni content, the magnetic anisotropy of the alloy is enhanced. However, when the Ni content is ≥15%, the coercivity and remanence properties of the alloy deteriorate sharply. The low coercivity and remanence properties can be obtained by reducing the heat treatment temperature, and the B–H loop linearity of the alloy is improved.

3. A large magnetic field-induced $K_u$ of 225 J/m$^3$ can be obtained through optimized annealing, and the Fe$_{66.8-x}$Co$_{10}$Ni$_{x}$Cu$_{0.8}$Nb$_{2.9}$Si$_{11.5}$B$_{8}$ nanocrystalline alloy exhibits a linear B–H loop and its initial permeability $\mu_i$ is as low as 2,000. It makes the alloy a promising material in the field of common mode chokes or current transformers, where the ability to resist the direct current is necessary.

**DATA AVAILABILITY STATEMENT**

The original contributions presented in the study are included in the article/Supplementary Material; further inquiries can be directed to the corresponding author.

**AUTHOR CONTRIBUTIONS**

ZL, the first person to complete the thesis work. KY, provided ideas for this work and finished the thesis together. TL, provided materials to finish the experimental works. XL, participated the testing work. SW, provided suggestions for the completion of the thesis.

**FUNDING**

This work was supported by the National Key Basic Research and Development Program (Grant No. 2016YFB0300502), National Natural Science Foundation of China (11875270), and Youth Innovation Promotion Association CAS (2018015).

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**Conflict of Interest:** TL was employed by the company China Iron and Steel Research Institute Group, Advanced Technology and Materials Co., Ltd.

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