Large exchange bias in magnetic shape memory alloys by tuning magnetic ground state and magnetic-field history

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ABSTRACT The exchange bias is of technological significance in magnetic recording and spintronic devices. Pursuing a large bias field is a long-term goal for the research field of magnetic shape memory alloys. In this work, a large bias field of 0.53 T is achieved in the $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16-x}\text{Fe}_x$ ($x = 1, 3, 5$) system by tuning the magnetic ground state (determined by the composition x) and the magnetic-field history (determined by the magnetic field H_{FC} during field cooling and the maximum field H_{Max} during isothermal magnetization). The maximum volume fraction of the interfaces between the ferromagnetic clusters and antiferromagnetic matrix and the strong interfacial interaction are achieved by tuning the magnetic ground state and the magnetic-field history, which results in strong magnetic unidirectional anisotropy and the large exchange bias. Moreover, two guidelines were proposed to obtain the large bias field. Firstly, the composition with a magnetic ground state consisting of the dilute spin glass and the strong antiferromagnetic matrix is preferred to obtain a large bias field; secondly, tuning the magnetic-field history by enhancing H_{FC} and reducing H_{Max} is beneficial to achieving large exchange bias. Our work provides an effective way for designing magnetically inhomogeneous compounds with large exchange bias.

Keywords: magnetic shape memory alloys, martensite, exchange bias, spin glass, antiferromagnetic

INTRODUCTION

The exchange bias (EB) phenomenon is characterized by

a shift of the hysteresis loop along the magnetic field axis. Since its discovery in Co/CoO fine particles by Meiklejohn and Bean in 1956 [1], various systems have been investigated to extend the material families with EB, including multilayers [2], nanostructures [3], thin films [4–6], and bulk compounds with intrinsic phase separation [7]. The EB usually appears in an inhomogeneous magnetic state with ferromagnetic (FM) and anti-ferromagnetic (AFM) subsystems, which is attributed to the field-induced unidirectional magnetic anisotropy formed at the interfaces between the FM and AFM subsystems [8]. Pursuing a large bias field is a long-term goal of EB research because of its significant technological applications in ultrahigh-density magnetic recording, giant magnetoresistance, and spin-valve devices [9–11].

Off-stoichiometric Mn-rich Ni–Mn–Z ($Z = \text{Sb}, \text{Sn}, \text{In}, \text{Ga}$) magnetic shape memory alloys (MSMAs) have complicated magnetic states, which are suitable hosts for investigating EB. In these alloys, the FM interaction originates from the Mn–Mn exchange interaction within the regular Mn sublattice and the AFM interaction originates from the Mn–Mn exchange interaction between the regular Mn sublattice and Z sublattice. The EB behaviours of these MSMAs vary greatly by changing compositions, because the FM and AFM interactions change considerably with the variation of compositions. As a consequence, different magnetic ground states, such as FM/AFM [12], ferrimagnetic/AFM [13], and spin glass (SG)/AFM states [14–16], may form, and the unidirectional

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magnetic anisotropy at the interfaces between the FM clusters and the AFM matrix varies accordingly. Therefore, the magnetic ground states (corresponding to different compositions) of MSMAs play an important role in defining the EB properties.

The EB measurement follows a two-step protocol: i) cooling from a high temperature to the desired testing temperature under a constant magnetic field (H_{FC}), i.e., the field cooling (FC) process; ii) a subsequent isothermal magnetization process by sweeping a magnetic field to the maximum value (H_{Max}). Many investigations have demonstrated that the unidirectional anisotropy can be established below a blocking temperature during the FC process [17–31]. On the other hand, in some Ni-Mn-based MSMAs, the unidirectional anisotropy can also be isothermally created by applying a sufficiently large H_{Max} after a zero-field cooling (ZFC) process [32–35]. A magnetic field-induced transition even occurs from a super SG to a superferromagnetic domain state during the isothermal magnetization process, which leads to spontaneous EB behaviour [34,35]. These experimental observations reveal that the magnetic-field histories (corresponding to H_{FC} and H_{Max}) play a significant role on the EB behaviour.

As mentioned above, the EB is dependent on three key parameters: composition, H_{FC} , and H_{Max} . Numerous efforts have been devoted to achieve large EB in MSMAs by tuning these factors. In previous studies, only one or two of the three factors have been tuned to improve EB in separate MSMAs [27,34], which provides partial information on determining the bias field. The achievement of maximal EB requires tuning all of the above factors. However, a systematic study of the dependence of EB on these three factors in the same MSMA system is lacking so far. Thus, the guidelines for achieving large EB are far from being established.

To explore the effective means of obtaining large EB in MSMAs, the EB of $Ni_{50}Mn_{34}In_{16-x}Fe_x$ ($x = 1, 3, 5$) was systematically studied with different magnetic ground states (corresponding to different x) and magnetic-field histories (corresponding to different H_{FC} and H_{Max}). The $Ni_{50}Mn_{34}In_{16}$ MSMA with a cluster SG (CSG)/AFM state [34] was chosen as the terminal alloy because it not only shows the EB effect, but also exhibits many other functional properties, such as giant magnetothermal conductivity [36], kinetic arrest [37], and the magnetocaloric effect [38]. The substitution of Fe for In can change the AFM interactions between the Mn and In/Fe atoms so as to tune the magnetic ground state of the system. We found that the ground state of $Ni_{50}Mn_{34}In_{16-x}Fe_x$ changes

from CSG/AFM to dilute SG (DSG)/AFM, as the Fe content x changes from 1 to 5. More importantly, the systematic measurements of EB behaviours with different x , H_{FC} , and H_{Max} reveal that two guidelines should be orderly followed to obtain the large bias field. First, the composition with a magnetic ground state of DSG/AFM, in which small FM clusters embedded in a strong AFM matrix, is preferred to obtain large EB; second, tuning the magnetic-field history by enhancing H_{FC} and reducing H_{Max} is beneficial to achieving large EB, because the maximum volume fraction of FM/AFM interfaces and strong interface interaction can be obtained under these two conditions. This further results in strong unidirectional magnetic anisotropy and large EB. A maximum bias value of 0.53 T is obtained for $x = 5$ under $H_{FC} = 1.2$ T and $H_{Max} = 1$ T. Our work contributes to the comprehensive understanding of the EB behaviour of bulk compounds with inhomogeneous magnetic phases, which provides an effective strategy for designing MSMAs with large EB.

EXPERIMENTAL SECTION

Polycrystalline ingots of nominal $Ni_{50}Mn_{34}In_{16-x}Fe_x$ ($x = 1, 3, 5$) MSMAs were prepared by arc melting stoichiometric amounts of the high-purity constituent elements under an Ar atmosphere. To achieve high homogeneity, the alloys were further annealed at 1173 K for 12 h in an evacuated quartz tube and then quenched in room-temperature water. Differential scanning calorimetry (DSC, TA instruments, Q2000) was used to determine the martensitic transition temperature with a temperature variation rate of 10 K min^{-1} . Furthermore, X-ray powder diffraction (XRD) was performed at room temperature to detect the structure of the samples by using a Bruker D8 Powder diffractometer with the Cu $K\alpha_1$ line ($\lambda_1 = 1.5418\text{ \AA}$). The temperature dependence of the magnetization ($M-T$), magnetic hysteresis ($M-H$) loops, and alternating current (AC) susceptibility were measured by using a superconducting quantum interference device (Quantum Design, MPMS-XL-5) magnetometer. Two different protocols, namely, ZFC and FC, were adopted in the $M-T$ measurements. The ZFC magnetization was obtained by cooling a sample from 400 to 3 K in zero field and then applying a magnetic field of 10 mT to measure the magnetization as the sample warmed up. The FC magnetization was subsequently obtained by measuring the magnetization with the same field of 10 mT upon both cooling and heating. All of the $M-H$ loops were isothermally recorded with the measurement field H_{Max} at 5 K after cooling with a field H_{FC} from 400 K. The tem-

perature dependence of the AC susceptibility χ' was measured using a sinusoidal magnetic field with an amplitude of 0.5 mT and a frequency range from 1 to 333 Hz.

RESULTS AND DISCUSSION

The martensitic transformation of the $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16-x}\text{Fe}_x$ ($x = 1, 3, 5$) alloys was identified by DSC measurements, as shown in Fig. 1a–c. The DSC curves of these samples display large exothermic and endothermic peaks with distinct thermal hysteresis upon cooling and heating procedures, revealing the occurrence of the martensitic transformation. The martensitic transformation temperature T_M is indicated by the arrows (Fig. 1a–c). It shifts to higher temperatures with increasing Fe content x . Fig. 1d–f present the ZFC and FC curves of the samples, which characterize their magnetic transition behaviours. The ZFC and FC curves of all the samples show an abrupt increase at high temperatures, demonstrating the existence of the FM phase in austenite. Its FM transition

temperature is denoted by T_c (Fig. 1d–f). After slight cooling below T_c , the ZFC and FC curves exhibit a sharp drop at T_M due to the martensitic transition. In the low-temperature martensitic region, the ZFC curves of the samples exhibit a peak at T_p , below which the ZFC and FC curves deviate from each other. This demonstrates the existence of a magnetic freezing transition. It has been shown that AFM interactions are greatly enhanced below the martensitic transition in Mn-rich Ni–Mn–In MSMA [39], leading to a sharp drop of magnetization at T_M . Moreover, weak FM interactions may also exist in the martensitic state [40], which give rise to the formation of many FM clusters coexisting with the surrounding AFM matrix. Owing to the spin frustration caused by the inherent competition between FM and AFM interactions, these FM clusters undergo a freezing transition into the SG state below T_p [19,27,30].

The freezing transition of the $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16-x}\text{Fe}_x$ ($x = 1, 3, 5$) samples was further characterized by the temperature dependence of AC magnetic susceptibility χ'

Figure 1 DSC curves of $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16-x}\text{Fe}_x$ MSMA for (a) $x = 1$, (b) $x = 3$, and (c) $x = 5$. Temperature dependence of magnetization curves with ZFC (open circles) and FC (solid squares) protocols under a field of 10 mT for (d) $x = 1$, (e) $x = 3$, and (f) $x = 5$. Real part (χ') of AC susceptibility vs. temperature for (g) $x = 1$, (h) $x = 3$, and (i) $x = 5$, which was measured in a frequency range of 1–333 Hz at a magnetic field strength of 0.5 mT. Insets of (g–i) show that the $\ln(\tau)$ vs. T_p (solid spheres) curve conforms to the power-law (red line).

(Fig. 1g–i). The susceptibility peak temperature T_p displays apparent frequency dispersion for all the three samples, which is the signal of the SG transition. The frequency dependence of T_p for the SG transition conforms to the critical-slowing-down power law $\tau = \tau_0 (T_p/T_g - 1)^{-z\nu}$ [27,30]. In this formula, τ is the relaxation time of locally correlated spins and equals $(2\pi f)^{-1}$ (f is the frequency); τ_0 , T_g , and $z\nu$ are the microscopic relaxation time, the SG transition temperature, and the dynamic critical exponent, respectively. As shown in the insets in Fig. 1g–i, the dependence of $\ln\tau$ on T_p for the three samples can be well fitted by the critical-slowing-down power law. The τ_0 values of these samples obtained by fitting are plotted in Fig. 2a and decreases from $\sim 10^{-9}$ to $\sim 10^{-11}$ s as x increases from 1 to 5. It was reported that τ_0 is approximately 10^{-9} s for CSG [34,35], while it lies in the range of $\sim 10^{-12}$ to 10^{-14} s for DSG [41–43]. Thus, the results in Fig. 2a reveal the low-temperature glassy state of $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16-x}\text{Fe}_x$ changes from CSG to DSG with increasing x .

The variation of the low temperature glassy state is due to the change of AFM interaction of the $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16-x}\text{Fe}_x$ system, which is achieved by tuning the lattice parameter through the substitution of Fe for In. The room-temperature XRD patterns (Fig. S1, SI) show the $L2_1$ parent phase for $x = 1$ and the six-layered modulated (6M) monoclinic martensite phase for $x = 5$ [44]. The $L2_1$ lat-

tice parameter ($a = 5.999$ Å) obtained from the Rietveld refinement for $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{15}\text{Fe}_1$ ($x = 1$) is smaller than that reported for $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16}$ ($a = 6.006$ Å) [36]. Such a decrease in the lattice parameter is due to the substitution of small-radius Fe for large-radius In atoms. With the increase of x , the lattice parameter of the system gradually decreases, which is beneficial to promoting the AFM interaction between the regular Mn sublattice and the In/Fe sublattice in its martensite phase. This characteristic is also consistent with the results in Fig. 2b, where the saturated magnetization of martensite (M_s^M) at 5 K considerably decreases with increasing x . Since the AFM interaction becomes stronger, the volume fraction of the AFM matrix extends and the size of the FM clusters shrinks correspondingly, resulting in the gradual change from CSG into DSG with increasing x .

The phase diagram in Fig. 2c summarises the evolution of the structural transformations and the magnetic phases as a function of x for the $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16-x}\text{Fe}_x$ system. The substitution of Fe for In shifts T_M to a higher temperature, which agrees with previous results for Ni–Mn–In–Fe MSMAs [45]. T_c of austenite insignificantly changes within $1 \leq x < 3$, while it overlaps with T_M for $3 < x \leq 5$. The low temperature magnetic state for $x = 1$ is CSG/AFM with large FM clusters embedded in the AFM matrix (Fig. 2d). The clusters of the CSG become smaller (Fig. 2e) for $x = 3$. The low temperature magnetic state for

Figure 2 (a) Dependence of τ_0 on x obtained by fitting the AC susceptibility in Fig. 1. (b) Saturated magnetization (M_s^M) in martensite at 5 K as a function of x . (c) Phase diagram for $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16-x}\text{Fe}_x$ ($x = 1$ –5) MSMAs. (d–f) show schematically that the low-temperature magnetic state evolves from CSG/AFM to DSG/AFM as x increases from 1 to 5. The yellow circles represent the FM clusters with different sizes. The blue background represents the AFM matrix, the color of which gradually becomes darker, representing the AFM interaction becoming stronger.

$x = 5$ changes to DSG/AFM, where many atomic spins are embedded in the AFM matrix (Fig. 2f). The SG transition temperature T_g rapidly decreases with increasing x , which is consistent with the fact that large clusters tend to freeze at higher temperatures.

Fig. 3 displays the dependence of the M - H curves on the Fe content x (corresponding to the magnetic ground states), H_{FC} , and H_{Max} (corresponding to the magnetic-field history) for the $Ni_{50}Mn_{34}In_{16-x}Fe_x$ system. All M - H curves were measured at 5 K. The M - H curves for different x under the same H_{FC} (0.4 T) and H_{Max} (1 T) are compared in Fig. 3a. The inset reveals that all the three samples exhibit EB with the M - H curves shifted toward the negative magnetic field axis. The increase in x changes the M - H curves from an FM-like shape to a paramagnetic (PM)-like shape, because the change of magnetic state from CSG/AFM to DSG/AFM is accompanied with a considerable weakening of magnetism. Fig. 3b compares the M - H curves under different H_{FC} but with the same x (3) and H_{Max} (1 T). Its M - H curve changes from PM-like to FM-like shape and its EB behaviour also changes greatly with increasing H_{FC} , because the FM subsystems are strengthened and the corresponding magnetization is enhanced by increasing H_{FC} . The M - H curves for different H_{Max} but with the same x (3) and H_{FC} (0.4 T) are compared in Fig. 3c. Interestingly, the shapes of the M - H curves are similar but their shifts are quite different for various H_{Max} . This characteristic demonstrates that H_{Max} can also change the magnetic state and thus affects the EB behaviour of the system.

In order to obtain a better understanding of the dependence of EB on x , H_{FC} , and H_{Max} , the bias field H_{EB} as a function of H_{FC} for different H_{Max} and x are depicted in Fig. 4a. H_{EB} was calculated by using $H_{EB} = -(H_L + H_R)/2$, where H_L and H_R are the left and right coercive fields,

respectively. H_{EB} increases up to a maximum bias value (H_{MEB}) and then decreases with increasing H_{FC} , forming a peak shape of the H_{EB} vs. H_{FC} curves. The H_{EB} increases with increasing H_{Max} when H_{FC} is very small, but it is obviously reduced by H_{Max} when H_{FC} is relatively large. Moreover, the shape of the H_{EB} vs. H_{FC} curve changes from a sharp peak to a broadened profile.

The dependence of EB on x , H_{FC} , and H_{Max} (Fig. 4a) can be understood as the follows. After the FC process, the random moments of FM clusters or spins become aligned along the positive H_{FC} direction. The unidirectional anisotropy can be created at the interface between FM clusters and the AFM matrix, because the AFM matrix exerts a pinning effect on the FM clusters at the interface through the interaction between them [1,11]. Evidently, the unidirectional anisotropy at the interface and its associated H_{EB} is determined by two factors: i) the volume fraction of the FM/AFM interface, which relies on the average size of the FM clusters and the distance between FM clusters; ii) the strength of the FM/AFM interface interaction.

The change of H_{EB} with H_{FC} (dot dash line of Fig. 4b) can be explained by Fig. 4b(1–3), which show the influence of H_{FC} on the magnetic state under the small H_{Max} . With increasing H_{FC} , the average size of the FM clusters grows, thus the volume fraction of the FM/AFM interface increases, leading to the enhancement of H_{EB} . However, H_{EB} cannot monotonously increase with H_{FC} , since the increase in H_{FC} not only causes cluster growth, but also reduces the distance between clusters and even makes them merge with each other, which reduces the volume fraction of FM/AFM interfaces and the corresponding H_{EB} . In the low- H_{FC} region, the average size of the FM clusters is small and the distance between them is large (Fig. 4b(1)), the growth of the volume fraction of the FM/

Figure 3 M - H curves of $Ni_{50}Mn_{34}In_{16-x}Fe_x$ ($x = 1, 3, 5$) MSMA. (a) M - H curves measured with $H_{FC} = 0.4$ T and $H_{Max} = 1$ T for $x = 1, 3, 5$. (b) M - H curves measured with $x = 3$ and $H_{Max} = 1$ T for $H_{FC} = 0.01, 0.4$, and 6 T. (c) M - H curves obtained with $x = 3$ and $H_{FC} = 0.4$ T for $H_{Max} = 1, 4$, and 6 T. A magnified view of the low-field region is depicted in the inset of (a) and (c).

Figure 4 (a) H_{EB} as a function of H_{FC} for $Ni_{50}Mn_{34}In_{16-x}Fe_x$ ($x = 1, 3, 5$) MSMAFs measured under $H_{Max} = 1, 4$, and 6 T. (b) Sketches for the evolution of magnetic state from (1) to (5) with varying H_{FC} and H_{Max} , which explains the dependence of EB on H_{FC} and H_{Max} . The yellow circles represent the FM clusters with different sizes and the blue background represents the AFM matrix.

AFM interface plays a dominant role in determining H_{EB} . However, the average cluster size becomes quite large and a considerable number of clusters merge (Fig. 4b(2 and 3)), when H_{FC} exceeds a critical value (H_p). For such a case, the decrease of volume fraction of the FM/AFM interface plays a dominant role in determining H_{EB} . Therefore, H_{EB} vs. H_{FC} shows an increase below the H_p but a decrease above it.

The changes of H_{EB} with H_{Max} can be interpreted by Fig. 4b(4 and 5), which show the effects of H_{Max} on the magnetic state. In the very small H_{FC} region ($< H_{Cross}$), the FM clusters are small and the distance between them is large (Fig. 4b(1)). Most of them grow but do not touch with each other when H_{Max} is applied (Fig. 4b(4)). Thus, the volume fraction of the FM/AFM interface and corresponding H_{EB} become larger. However, in the larger H_{FC} region ($> H_{Cross}$), a considerable number of FM clusters grow up and become close. The corresponding magnetic state under large H_{FC} and small H_{Max} is schematically displayed in Fig. 4b(2). The FM clusters merge

by applying large H_{Max} (Fig. 4b(5)), which reduces the volume fraction of the FM/AFM interface and the corresponding H_{EB} .

The dependence of H_{MEB} on x can be explained by Fig. 2d–f, which displays the change of magnetic state with x . For a certain composition x , its largest bias field H_{MEB} corresponds to the optimized volume fraction of the FM/AFM interface obtained by tuning H_{FC} and H_{Max} . However, the strength of the FM/AFM interaction at the interface is quite different for different compositions. With increasing x , the AFM interaction in the matrix becomes much stronger, which exerts a stronger pinning effect on FM clusters. This promotes strong interfacial unidirectional anisotropy, resulting in a large H_{MEB} .

Based on the data in Fig. 4a, H_{MEB} of $Ni_{50}Mn_{34}In_{16-x}Fe_x$ ($x = 1, 3, 5$) is plotted as functions of x , H_{FC} , and H_{Max} in Fig. 5. These results clearly reveal that the optimized value of H_{MEB} is jointly determined by the three parameters (x , H_{FC} , and H_{Max}). For a fixed composition x , the combination of an increase in H_{FC} and a decrease in H_{Max} is

Figure 5 Three-dimensional contour for H_{MEB} (black spheres) as the function of x , H_{FC} , and H_{Max} .

beneficial to achieving large EB, because the maximum FM/AFM interface can be obtained in such a magnetic-field history. Furthermore, H_{MEB} shifts to higher values, when x changes from 1 to 5. The $x = 5$ sample with DSG/AFM state shows large H_{MEB} , and its H_{MEB} increases rapidly by tuning H_{FC} and H_{Max} . By contrast, the $x = 1$ sample with CSG/AFM shows much smaller H_{MEB} , and its H_{MEB} increases slowly for the same situation. This demonstrates that a DSG/AFM state with small FM clusters embedded in a strong AFM matrix is more capable of achieving large EB. The guidance for obtaining large EB is visually indicated by an arrow on the contour surface of Fig. 5, which can be described as the following two points: i) the composition with a magnetic ground state of DSG/AFM is preferred to obtain large EB; ii) tuning the magnetic-field history by enhancing H_{FC} and reducing H_{Max} is beneficial to achieving large EB. The maximum bias field of 0.53 T is obtained for $x = 5$ under $H_{\text{FC}} = 1.2$ T and $H_{\text{Max}} = 1$ T. The maximum H_{EB} of the $x = 5$ sample ($\text{Ni}_{2.36}\text{Mn}_{1.36}\text{In}_{0.44}\text{Fe}_{0.2}$) with respect to its optimized H_{FC} and H_{Max} is compared with those of some important MSMAs in Fig. 6. As indicated by a red pentagram, its maximum H_{EB} is much larger than those of other MSMAs.

CONCLUSIONS

In this study, the EB behaviours of $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16-x}\text{Fe}_x$ ($x = 1, 3, 5$) MSMAs have been systematically investigated by tuning the magnetic ground states (composition x) and magnetic-field histories (H_{FC} and H_{Max}). The following conclusions were obtained: (1) the ground state changes from the CSG/AFM to DSG/AFM as x increases from 1 to 5, because the substitution of Fe for In decreases the lattice parameter of the system, which leads to the en-

Figure 6 Plot of H_{EB} with respect to H_{FC} and H_{Max} for important MSMAs.

hancement of the AFM interaction and the size reduction of FM clusters; (2) through tuning the magnetic ground states (composition x) and magnetic-field histories (H_{FC} and H_{Max}), the volume fraction of FM/AFM interfaces and the strength of interface interaction can be optimized, which leads to strong unidirectional magnetic anisotropy at FM/AFM interfaces and large EB. A large bias field of 0.53 T is achieved for $x = 5$, $H_{\text{FC}} = 1.2$ T, and $H_{\text{Max}} = 1$ T; (3) two guidelines should be utilized to realize large EB: i) the composition with a magnetic ground state of DSG/AFM consisting of small FM clusters and strong AFM matrix is preferred; ii) tuning magnetic-field history by enhancing H_{FC} and reducing H_{Max} . These guidelines provide an effective way for producing large EB in the systems with inhomogeneous magnetic phases.

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Supplementary information Supporting data are available in the online version of the paper.

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基于磁性形状记忆合金磁基态和磁场历史调控的大交换偏置效应

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摘要 由于交换偏置效应对磁记录和自旋电子器件的重要意义, 在磁性形状记忆合金中获得大偏置场一直是人们长期努力的目标. 本文通过调控 $\text{Ni}_{50}\text{Mn}_{34}\text{In}_{16-x}\text{Fe}_x$ ($x = 1, 3, 5$) 磁性形状记忆合金的磁性基态(取决于合金成分)和磁场历史(取决于冷却过程中冷却场 H_{FC} 和等温磁化过程中所施加的最大测试场 H_{Max}), 对交换偏置行为进行了系统研究. 结果表明, 调控磁性基态和磁场历史可以获得最大铁磁簇与反铁磁基体界面体积分数以及它们之间的强相互作用, 从而诱发了界面上强单向各向异性, 产生了大交换偏置场. 本文提出了获得大交换偏置场需要遵循的两条原则. 首先, 选择稀磁自旋玻璃与强反铁磁基体的磁性基态成分可获得大交换偏置场; 其次, 通过增强 H_{FC} 和降低 H_{Max} 有利于实现大交换偏置. 本研究为设计具有大交换偏置场的非均匀磁性相化合物提供了一种有效的方法.