Stable magnetostructural coupling with tunable magnetoresponsive effects in hexagonal ferromagnets

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The magnetostructural coupling between the structural and the magnetic transition has a crucial role in magnetoresponsive effects in a martensitic-transition system. A combination of various magnetoresponsive effects based on this coupling may facilitate the multifunctional applications of a host material. Here we demonstrate the feasibility of obtaining a stable magnetostructural coupling over a broad temperature window from 350 to 70 K, in combination with tunable magnetoresponsive effects, in MnNiGe:Fe alloys. The alloy exhibits a magnetic-field-induced martensitic transition from paramagnetic austenite to ferromagnetic martensite. The results indicate that stable magnetostructural coupling is accessible in hexagonal phase-transition systems to attain the magnetoresponsive effects with broad tunability.
The ferromagnetic martensitic transition (FMMT)\textsuperscript{1–3}, a coinciding crystallographic and magnetic transition mainly found in Fe-based and Heusler ferromagnetic alloys, is receiving increasing attention from both the magnetism and the material science community due to the massive variations of associated magnetoresponsive effects, such as magnetic-field-induced shape memory/strain effect\textsuperscript{4–7}, magnetoresistance\textsuperscript{8–10}, Hall effect\textsuperscript{11} and magnetocaloric effect\textsuperscript{11,12}. These effects are of interest for many potential technological applications like magnetic actuators\textsuperscript{13,14}, sensors\textsuperscript{15}, energy-harvesting devices\textsuperscript{16} and solid-state magnetic refrigeration\textsuperscript{17}. In these functionalities, the magnetostructural coupling between the structural and the magnetic transition has an essential role. Seeking a stable coupling in a broad temperature range is a scientific and technological challenge.

In the case of ferromagnetic phase transitions coupled with martensitic–like structural changes, it is the ferromagnetic ordering (for example, spontaneous magnetization) that triggers modest structural modifications due to the magnetoelastic coupling\textsuperscript{18}. These magnetoelastic transitions have been utilized in the intensive investigations of a large body of giant magnetocaloric materials\textsuperscript{19–27}. In contrast, in typical FMMTs, the change of structural symmetries of austenite and martensite is remarkable\textsuperscript{28–35}. The transition converts the different magnetic states (moment values and type of coupling) in between the two phases that have separate Curie (Néel) temperatures. Thus, this magnetostructural transition with large symmetry change and atomic displacement can bring about various magnetoresponsive effects.

Since the discovery of magnetic field-induced shape memory effect in the Ni–Mn–In Heusler alloy\textsuperscript{2}, attempts have been made to induce the magnetostructural transition by applying a magnetic field. To this end, a large magnetization difference $\Delta M$ between the austenite and the martensite phase is necessary to maximize the magnetic-energy change introduced by applying a magnetic field. In a given system, if the MT is used to convert the magnetic states from the paramagnetic (PM) to the ferromagnetic (FM) state, rather than from FM to FM\textsuperscript{28}, a large $\Delta M$ will be gained for the magnetic-field-induced MT. With this transition, also a decrease of the magnetic entropy is associated. For a martensitic-transition system, this PM–FM-type MT requires a primary condition that the Curie temperature of the martensite should be higher than that of the austenite. Such an MT is rarely observed in the case of Fe-based and Heusler alloys. Therefore, it is of interest to find an alloy system, which exhibits this particular magnetostructural transition, especially in a broad temperature range.

Recently, the magnetic-field-induced shape memory effect based on the magnetostructural coupling has also been found in another type of materials\textsuperscript{28}, the hexagonal ternary compounds with the Ni$_3$In structure\textsuperscript{29–31}. With the FMMTs in these materials, large magnetocaloric effects are also associated\textsuperscript{32–36}. In these compounds the magnetic-ordering (Curie or Néel) temperatures of the martensite are higher than those of the austenite\textsuperscript{32,33,36}. This satisfies the primary condition for potential PM–FM-type magnetostructural transitions. This large material pool provides a new platform for the desired magnetostructural transitions.

In this study we realize a PM–FM-type magnetostructural transition in hexagonal phase-transition materials in a broad temperature window by suitable chemical substitution of Fe in MnNiGe. A stable magnetostructural coupling can be achieved by simultaneous manipulation of the phase stability and the magnetic structure. On the basis of valence-electron localization function (ELF) calculations and magnetic configuration analysis, the increased phase stability and the conversion of magnetic structure are, respectively, attributed to the strengthened covalent bonding and Fe–6Mn local atomic configurations introduced by the Fe substitution. The MnNiGe:Fe materials exhibit magnetic-field-induced martensitic transitions and giant magnetocaloric effects with broad tunability.

### Results

**Design scheme.** To obtain the desired magnetostructural transition, we consider the hexagonal material MnNiGe as our starting system. Stoichiometric MnNiGe undergoes an MT at a quite high temperature $T_\text{C} = 470$ K from the ordered Ni$_3$In-type hexagonal structure ($P6/mmm$, 194) to the TiNiSi-type orthorhombic structure ($Pnma$, 62) (refs \textsuperscript{37–39}; Fig. 1). Because this transition occurs in the PM state, the expected magnetostructural coupling cannot be established. Upon cooling, the martensite phase shows a magnetic transition from the PM state to the antiferromagnetic (AFM) state at a Néel temperature $T_{\text{N}}^\text{M} = 346$ K (ref. \textsuperscript{37}). The magnetic moments of 2.8 $\mu_B$ which are only localized on the Mn atoms, form an AFM spiral structure\textsuperscript{37,38} so that the magnetization is very low. Besides, on the basis of data collected for near-stoichiometric MnNiGe systems (Supplementary Table S1), it can be estimated that the Curie temperature ($T_C^\text{M}$) of the high-temperature austenite of stoichiometric MnNiGe lies around 205 K, which is about 140 K lower than $T_N^\text{M}$ of the low-temperature martensite (Fig. 1a). Therefore, MnNiGe shows a potential possibility for the expected PM–FM-type magnetostructural transitions.

For modifying MnNiGe into a desired material, two important changes still have to be introduced in the system. In the first place, the martensite-transition temperature $T_\text{C}$ should be lowered in a controllable fashion to a temperature within the temperature interval, as indicated by dotted blue arrow in Fig. 1a, to establish the magnetostructural coupling. The second necessary modification is that the AFM transition in the martensite phase should be converted into a FM transition, that is, $T_N^\text{M}$ should become $T_C^\text{M}$. This modification is indicated in Fig. 1a by the dotted red arrow and line. Achievement of these two modifications turns out to be crucial for opening a temperature window between $T_C^\text{M}$ and $T_N^\text{M}$ in which the PM–FM-type magnetostructural transition with an appreciable value of $\Delta M$ can be realized.

To achieve this, it seems promising to substitute in MnNiGe the magnetic element Fe for the non-magnetic Ni or the magnetic Mn. This is promising because, there is no MT occurring in the isostructural compounds MnFeGe and FeNiGe so that the austenite structure is maintained down to 4.2 K (ref. \textsuperscript{29}). At the same time, the
magnetic Fe may alter the spirally AFM coupling of Mn moments in alloyed MnNiGe. In this sense, alloying these Fe-containing isostructures with MnNiGe may give rise to a more stable austenite (that is, with lower \( T_t \)) and a higher magnetization of martensite (that is, FM instead of AFM martensite). In the present investigation, we have partly substituted Fe for Ni and Fe for Mn in MnNiGe to create the quasi-ternary systems MnNi\(_1-x\)Fe\(_x\)Ge and Mn\(_1-x\)Fe\(_x\)NiGe.

Sample preparation and characterization. The samples were prepared by arc melting and homogenization annealing. The structure of samples was determined with X-ray diffraction (XRD) and no impurity phase was found. Details of the methods are given in the Methods section. With increasing Fe content, the transformation temperature from Ni\(_3\)In-type hexagonal austenite to TiNiSi-type orthorhombic martensite is gradually lowered from higher temperatures to below the room temperature (Fig. 2a). The XRD data show that the \( c_0 \) (\( a_0 \)) axis of the austenite phase decreases (increases) upon Fe substitution (Fig. 2b). Temperature-dependent XRD reveals that the MT begins at 240 K in Mn\(_{0.84}\)Fe\(_{0.16}\)NiGe (Fig. 2c). An increase of 2.68% in unit-cell volume is found at the transition (Fig. 2d; Supplementary Table S2). This volume expansion is large and opposite to the usual contraction of about −1% at martensitic structural transitions. This indicates that the crystalline structure and the atomic surrounding undergo a pronouced reconstruction during the structural transition, as shown in Fig. 1b.

Structural and magnetic phase diagrams. To determine the crystallographic and magnetic structures, low- and high-field \( M(T) \) measurements and differential thermal analysis (DTA) were used (Supplementary Figs. S1 and S2; Table 1). On the basis of the experimental data, the MnNi\(_{1-x}\)Fe\(_x\)Ge and Mn\(_1-x\)Fe\(_x\)NiGe phase diagrams are proposed as shown in Fig. 3. In both systems, the Fe substitution makes the MT fall within the temperature range of ferromagnetic order of martensites.

Upon substitution of Fe for Ni (MnNi\(_{1-x}\)Fe\(_x\)Ge, Fig. 3a), \( T_{C_M}^N \) becomes \( T_{C_M}^M \) at about 300 K. Upon further increase of Fe content, \( T_t \) continuously decreases until \( T_{C_A}^M \) is reached. It can also be seen that, upon substitution, both \( T_{N}^M \) (\( T_{C}^M \)) and \( T_{C}^A \) basically remain constant, which offers an accessible temperature window of about 90 K between them. Within this window, the system undergoes an MT coupled with a magnetic transition from the PM to the FM state. Below \( T_{C_A}^M \), the magnetostructural transition decouples as the MT rapidly vanishes. In the case of substitution of Fe for Mn (Mn\(_{1-x}\)Fe\(_x\)NiGe, Fig. 3b), a quite low level of Fe substitution (about \( x = 0.08 \)) already lowers \( T_t \) to meet \( T_{N}^M \), and to introduce ferromagnetism at a relatively high temperature of 350 K. In the range 0.08 ≤ \( x ≤ 0.26 \), the FM martensite phase has a high magnetization in 5 T. The Fe substitution efficiently converts AFM martensite into FM one while, surprisingly, it still drives the FM austenite parent phase into a weak-magnetic spin-glass-like state (Fig. 4; Supplementary Fig. S3). Along with the eventual vanishing of the MT at the freezing temperature (\( T_{f_g} \) of
the spin-glass-like state, the significant consequence is obtained: the lowest temperature of the window has moved down to about 70 K and a quite broad temperature interval up to 280 K is generated for the stable magnetostructural coupling.

It should be further pointed out that, the temperature hysteresis of the MT for both systems is significantly reduced from about 50 K to below 10 K by the Fe substitution (Fig. 3c and d; Table 1), which implies a decreasing thermodynamic driving force for the martensite nucleation. For a first-order MT, the hysteresis of 10 K is very small⁴, which is beneficial for the temperature sensibility of magnetoresponsive smart applications based on martensitic transitions.

Thermomagnetic behaviour. To clarify the PM–FM MTs in the broad window, we measured the high-field thermomagnetic properties of the typical samples of Mn1−xFexNiGe system, as shown in Figure 4 (thermomagnetic properties of MnNi1−xFexGe system are shown in Supplementary Fig. S4). In accordance with Fig. 3, Tt decreases with increasing Fe content. For x=0.08, PM–FM jumps of the magnetization, with large ΔM up to 60 A m² kg⁻¹ in a field of 5 T, are observed. This signifies that the introduction of Fe has led to a great change of the magnetic exchange interaction in the martensite phase, changing the spiral AFM structure into a FM state. Upon cooling, for each composition the FM martensite phase nucleates and grows in the PM austenite matrix. Upon heating, the reversible nature of the MT can be seen. Here, it should be emphasized again that Tt is the martensitic-transition temperature, not the Curie temperature. The Curie (Néel) temperatures of both phases have the values at the respective window boundaries, which are shown in Fig. 3b. When the transition occurs, the austenite is still in PM state, whereas the martensite is already in its FM state. It is the crystallographic structural transition between a PM phase and a FM phase that gives rise to the abrupt magnetization change. For x=0.26, it can be seen that the MT becomes incomplete and the spin-glass-like behaviour shown by the irreversible zero-field-cooling/field-cooling (ZFC–FC) curves is in accordance with the phase diagram in Figure 3b. Within this broad temperature window, a stable magnetostructural coupling is obtained from above room temperature (350 K) to liquid-nitrogen temperature (70 K).

Table 1 | Key parameters of MnNiGe:Fe systems.

| Alloying case          | X   | Tc (cooling)* | Tc (heating)* | ΔT | TcE(Tp)‡ | TNM‡ | M²M‡ | M²A‡ | ΔM# | Bcr** | Bs†† |
|------------------------|-----|---------------|---------------|----|----------|------|------|------|------|-------|------|
|                        | K   | A m² kg⁻¹     | Tesla        |    |          |      |      |      |      |       |      |
| MnNi1−xFexGe           | 0   | 470‡‡         | -             | 346‡‡ | 39‡‡    | -    | -    | -    | -    | >5‡‡  |       |
|                        | 0.10| 400           | 16            | 324 | 52           | -    | -    | -    | -    | 1.20   | >5   |
|                        | 0.15| 352           | 14            | 306 | 72           | -    | -    | -    | -    | 0.60   | >5   |
|                        | 0.20| 300           | 12            | 291 | 87           | -    | -    | -    | -    | 0.45   | >5   |
|                        | 0.27| 230           | 10            | 95  | 49           | -    | -    | -    | -    | 0.20   | >5   |
|                        | 0.30| 189           | 12            | 98  | 40           | 0.05 | 5.0  |      |      |        |      |
|                        | 0.33| -             | -             | -   | -            | -    | -    | -    | -    | -      |      |
|                        | 0.36| -             | -             | -   | -            | -    | -    | -    | -    | -      |      |
|                        | 0.40| -             | -             | -   | -            | -    | -    | -    | -    | -      |      |
|                        | 0.50| -             | -             | -   | -            | -    | -    | -    | -    | -      |      |
| Mn1−xFexNiGe           | 0.03| 429           | 43            | 349 | 52           | -    | -    | -    | -    | 1.13   | >5   |
|                        | 0.05| 389           | 20            | 347 | 65           | -    | -    | -    | -    | 0.86   | >5   |
|                        | 0.08| 336           | 21            | 341 | 75           | 36   | 0.57 | 5.0  |      |        |      |
|                        | 0.11| 298           | 7             | 78  | 55           | 0.48 | 4.0  |      |      |        |      |
|                        | 0.13| 277           | 8             | 80  | 57           | 0.45 | 3.0  |      |      |        |      |
|                        | 0.16| 241           | 8             | 78  | 59           | 0.33 | 2.3  |      |      |        |      |
|                        | 0.18| 207           | 7             | 79  | 59           | 0.24 | 1.8  |      |      |        |      |
|                        | 0.22| 153           | 10            | 76  | 52           | 0.05 | 0.9  |      |      |        |      |
|                        | 0.24| 127           | 8             | 75  | 54           | 0    | 0.5  |      |      |        |      |
|                        | 0.26| 74            | 10            | 25.7| -            | -    | -    | -    | -    | -      |      |
|                        | 0.28| -             | -             | -   | -            | -    | -    | -    | -    | -      |      |
|                        | 0.30| -             | -             | -   | -            | -    | -    | -    | -    | -      |      |
|                        | 0.35| -             | -             | -   | -            | -    | -    | -    | -    | -      |      |
|                        | 0.40| -             | -             | -   | -            | -    | -    | -    | -    | -      |      |
|                        | 0.50| -             | -             | -   | -            | -    | -    | -    | -    | -      |      |

* Martensitic transition temperature.
† Temperature hysteresis of martensitic transition.
‡ Curie temperature of austenites. Tt is frozen temperature of spin-glass-like state of Mn1−xFexNiGe system (x=0.26).
§ Néel temperature of martensites. Determined as average values measured on cooling and heating.
¶ Martensitic magnetization. Measured at 5 K in a field of 5 T.
‖ Austenitic magnetization. Measured at 5 K in a field of 5 T.
# Magnetization difference between austenite and martensite across martensitic transition.
** Critical field of metamagnetic transition in martensite phase.
†† Saturation field of martensite.
‡‡ From ref. 37.

Magnetic field-induced properties across the transitions. In what follows, we study typical magnetoresponsive properties for both systems. First, we present the magnetic-field-induced MT effect. Fig. 5a shows the magnetization curves of MnNi₀.⁷₇Fe₀.₂₃Ge at various temperatures within the temperature window. Above 276 K, the austenite shows a PM behaviour. Between 274 and 258 K, the continuous metamagnetic behaviour at each temperature in high field (marked by arrows) reveals a field-induced MT effect, indicating that the FM martensite phase is induced by an applied field in the PM austenite matrix. This behaviour corresponds to an upward shift of about 11 K of the martensitic starting transition.
**Figure 3 | Structural and magnetic phase diagrams and phase-transition hysteresis.** PM, FM and AFM indicate the PM austenite, the ferromagnetic martensite and the AFM martensite, respectively. (a,b) Structural and magnetic phase diagrams of MnNi$_{1-x}$Fe$_x$Ge (a) and Mn$_{1-x}$Fe$_x$NiGe (b). The red circles denote the martensitic-transition temperature $T_t$ and the red arrows indicate the decreasing trend of $T_t$. In the range 0.20  $\leq$ 0.30 for MnNi$_{1-x}$Fe$_x$Ge (a) and 0.08  $\leq$ 0.26 for Mn$_{1-x}$Fe$_x$NiGe (b), the systems undergo a Ni$_3$In-type to TiNi$_3$-type MT at $T_t$. Above $x$ = 0.30 (a) and $x$ = 0.26 (b), the systems are single-phase Ni$_2$In-type austenites. The solid diamonds correspond to $T_{CT}$ ($T_{CM}$) of the martensite (a,b) and the open diamonds to $T_{CT}$ ($T_{CM}$) of the austenite. In the range 0.20  $\leq$ 0.30 (a) and 0.08  $\leq$ 0.26 (b), the temperature windows are limited by $T_{CT}^M$  $-$  $T_{CT}^A$ (a) and by $T_{CM}^A$  $-$  $T_{CT}^B$ (b), respectively. In (a), the FM state of the martensite returns back to an AFM state upon cooling. The yellow stars show the critical temperature ($T_{CT}^A$) between the FM and AFM states in a field of 5 T (more details in Supplementary Fig. S5). In (b), the austenitic-phase zone with $x$  $\geq$ 0.26 enters into a spin-glass-like state below about 70 K (more details in Supplementary Fig. S3). (c,d) Temperature hysteresis of the first-order martensitic transition for MnNi$_{1-x}$Fe$_x$Ge (c) and Mn$_{1-x}$Fe$_x$NiGe (d). Here, the hysteresis is defined as the interval of peak values of the $dM/dT$ and/or DTA curves upon cooling and upon heating (Supplementary Figs S1, S2; Table 1). The error bars of the hysteresis in (c,d) are given by repeating the measurements.

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The possibility to obtain large $\Delta S_m$ values for the MnNi$_{1-x}$Fe$_x$Ge system in an interval of nearly 100 K. In the Mn$_{1-x}$Fe$_x$NiGe system, even more appreciable $\Delta S_m$ values are observed in an even more extended temperature window ranging from 350 to 70 K. As an example, a low substitution level of $x$ = 0.18 gives rise to a giant $\Delta S_m$ value of $\Delta S_m$ = 31 J kg$^{-1}$ K$^{-1}$ for $\Delta B$ = 5 T. These larger $\Delta S_m$ values are attributed to the more ferromagnetically ordered martensite and thus a lower magnetic-entropy state after the transition. In accordance with the field-induced MT effect, the $\Delta S_m$ peak position also shows a field dependence and shifts to higher temperatures with increasing magnetic field (dashed grey lines in Fig. 5c and d). Another feature of the MnNi$_{1-x}$Fe$_x$Ge systems is that the magnetic and martensitic transitions have the same sign of the enthalpy change (more details in Supplementary Figs S1 and S2), as the crystallographic and magnetic symmetries are both lowered on cooling. This prevents opposite heat processes that counteract the calorific effects, which is very different from the common FM–AFM(PM) MTs in Fe-based and Heusler magnetic-shape memory alloys.

**Discussion**

In this study, stable magnetostructural coupling has been realized by appropriate material design and the associated magnetoresponsive effects have been presented. This magnetostructural coupling has been achieved by decreasing $T_f$ of MT of the alloyed MnNiGe and converting the AFM to the FM state in martensite phase by
Figure 4 | Magnetostuctural coupling in the established temperature window. The magnetization curves of Mn$_{1-x}$Fe$_x$NiGe have been measured at ZFC-FC in an applied magnetic field of 5 T. The grey curve represents the magnetization of the spiral AFM martensite of Fe-free MnNiGe. For this stoichiometric MnNiGe, $T_N$ is 352 K and $T_t$ (460 K) is outside the temperature window. The magnetostuctural coupling based on the PM–FM-type MT is obtained in a broad temperature window. Displayed curves correspond to Fe contents $x = 0.08$ (purple), $0.11$ (red), $0.16$ (green), $0.18$ (blue), $0.24$ (orange) and $0.26$ (pink). An incomplete MT and spin-glass-like behaviour are observed for the sample with $x = 0.26$ (more details in Supplementary Fig. S3).

Figure 5 | Magnetic-field-induced martensitic transitions and magnetic-entropy changes. (a,b) Magnetic isotherms of MnNi$_{0.77}$Fe$_{0.23}$Ge (a) and Mn$_{0.95}$Fe$_{0.05}$NiGe (b) at various temperatures in the temperature window. The measurement temperature decreases from 280 to 252 K (a) and from 214 to 192 K (b) with an interval of 2 K. The metamagnetic behaviour (marked by arrows) indicates the magnetic-field-induced martensitic transition. (c,d) Isothermal magnetic-entropy changes ($\Delta S_m$) for various field changes derived from the magnetic isotherms of MnNi$_{1-x}$Fe$_x$Ge (c) and Mn$_{1-x}$Fe$_x$NiGe (d). Displayed curves correspond to the field changes $\Delta B = 1$ T (black squares), 2 T (red circles), 3 T (blue triangles), 4 T (green pentagons) and 5 T (pink diamonds). The shift of the $\Delta S_m$ maximum with increasing field to higher temperatures (indicated by the dashed grey lines) is related to the magnetic-field-induced MT effect. $M_s$ and $M_f$ denote the starting and finishing temperatures of the MT, respectively.
occupy, some Ni–6Mn local atomic configurations naturally change into Fe–6Mn ones. That is, crystallographically, every introduced Fe atom is always surrounded by six nearest-neighbour Mn atoms. With magnetic moments of $0.5 < \mu < 1 \mu_B$ (ref. 44) for Fe atoms, this Fe–6Mn local configuration internally establishes FM coupling. The spirally AFM-coupled Mn moments within the configuration is thus changed into parallel alignment because of the Fe moments. We schematically illustrate this FM Fe-centered local configuration (pink zones) in comparison with the spiral AFM matrix in Fig. 7a. This mechanism is similar to the FM exchange interaction established between Mn and Co atoms by substituting Co ($\mu = 1 \mu_B$) for Ni in MnNiGe (ref. 45). With increasing Fe content, the number of FM configurations will increase and they will overlap and form larger FM zones, in this way promoting the AFM–FM conversion in the martensite phase.

The above-described FM coupling in Fe–6Mn configurations is explicitly confirmed by the magnetization behaviour of MnNi$_{1-x}$Fe$_x$Ge martensites at 5 K in fields of up to 5 T (Fig. 7b). The Fe-free sample shows AFM behaviour with a metamagnetic kink at a critical field $B_{c2}$ of 1.3 T, indicating a metamagnetic transition from a spiral AFM to a canted FM state, in accordance with the reported stoichiometric MnNiGe (ref. 37). In contrast, the Fe-substituted martensites show a large slope of the $M(B)$ curves in low fields that increases as a function of the Fe content. This suggests that an increasing FM component is generated in the system due to the existence of the Fe-centered Fe–6Mn configurations. Meanwhile, $B_{c2}$ rapidly decreases with increasing Fe content (Fig. 7b,d), which corresponds to an increasing ease for the applied field to change the spiral AFM structure to a forced parallel alignment. The larger the number of local Fe–6Mn configurations becomes, lower the $B_{c2}$ will be. For $x = 0.30$ (the highest Fe content in the martensite phase due to the vanishing of the MT for higher substitution), a FM ground state with a very low $B_{c2}$ of 0.05 T is found. Owing to both the Fe substitution and the applied field, the magnetization reaches values of up to about 100 A m$^{-2}$ kg$^{-1}$ in a field of 5 T, much higher than that of the Fe-free sample (Fig. 7b,d).

By substituting Fe for Mn in MnNiGe, the AFM–FM conversion is expedited (Fig. 7c). This is because the Fe atoms not only introduce FM coupling between Fe and Mn atoms, but also break up the consecutive AFM sublattices of the Mn moments. This rapidly makes the AFM order collapse. Thus, only a small Fe content of about $x = 0.08$ is sufficient to approach the maximal magnetization for Mn$_{1-x}$Fe$_x$NiGe (Fig. 7c,d). The samples more and more easily get magnetically saturated and show a rapidly decreasing saturation field (Fig. 7c,e). The complete FM ground state appears in the sample with $x = 0.22$ (Fig. 7c), versus $x = 0.30$ in MnNi$_{1-x}$Fe$_x$Ge (Fig. 7b). Therefore, we conclude that the AFM–FM conversion becomes more efficient when the substituted Fe atoms occupy Mn sites in Mn$_{1-x}$Fe$_x$NiGe. As a consequence, this FM state in martensite phase facilitates the magnetoresponsive effects presented in this study.

In previous studies$^{45,46}$, it has been reported that the large-size Ge and zero-moment Ni in MnNiGe can be replaced by small-size Si and magnetic Co, respectively. Actually, these replacements are also equal to alloying the isostructural Ni$_2$In-type MnNiSi and MnCoGe compound with the MnNiGe mother compound. MnNiSi and MnCoGe undergo MTs at high temperatures and their martensite phases are both ferromagnetic$^{39}$. Therefore, in insert martensite structure, MnNiSi and MnCoGe can reasonably change the AFM state of MnNiGe into a FM state. However, they fail to lower the temperature of MT from 470 K of MnNiGe to below $T_{cM}^{13}$ which would establish the coupling needed for the magnetostuctural transition. The substitution of Fe applied in this study thus shows the best results for both the decrease of $T_{cM}$ and the magnetic-state conversion of MnNiGe; that is, for the desired PM–FM magnetostuctural transition. Very recently, an interesting paper has been published on the pressure-tuned magnetostuctural transition in Cr-doped MnCoGe$^{47}$.

Figure 6 | ELF of MnNi$_{0.5}$Fe$_{0.5}$Ge austenite. (a, b) ELF isosurface basins (top view) with an isovalue of $2.35 \times 10^{-4}$ for Ni–Ge (a) and Fe–Ge (b) layers. The Ge1 and Ge2 atoms are encapsulated by separate isosurfaces. More surfaces appear around Fe than around Ni. (c) ELF contour map in the (110) plane. The scale bar from blue to red corresponds to increasing electron localization (namely, increasing covalent bonding). Compared with Ni–Ge1 and Ni–Ge2, there are higher values at Fe–Ge2 and Fe–Ge1. The distance between Mn2 and Mn3 atoms at different sides of the Fe–Ge layer is reduced compared with that between Mn1 and Mn2 atoms at different sides of the Ni–Ge layer.
To summarize, we have found that a stable PM–FM magnetic transition and sensitive magnetic switching from the PM to the FM state. The presented design scheme may be of importance in exploring multifunctional magnetoresponse materials among new and known magnetic martensitic-transition systems.

**Methods**

**Sample preparation.** Polycrystalline ingots were prepared by arc melting high-purity metals in argon atmosphere. The ingots were melted four times and turned over in between to guarantee good alloying. The ingots were subsequently homogenized by annealing in evacuated quartz tubes under argon at 1,123 K for 5 days and slowly cooled at 1 K min\(^{-1}\) to room temperature to avoid stress in samples.

**Structural and thermal analysis.** The samples for powder XRD were made by fine grinding. The room-temperature powder XRD measurements were performed using a Rigaku XRD D/Max 2,000 diffractometer with Cu-Kα radiation. Temperature-dependent XRD measurements were performed from 285 to 98 K with a cooling rate of 2.5 K min\(^{-1}\) using a Bruker XRD D8-Advance diffractometer.
with Cu-Ka radiation. At each temperature, a waiting time of 30 min was included before the measurement. DTA with heating and cooling rates of 2.5 K min\(^{-1}\) was used to determine the martensitic-transition characteristic temperatures.

**Magnetic measurements.** Magnetization measurements were carried out on powder samples using a superconducting quantum interference device (SQUID, Quantum Design MPMS XL-7). Low-field \(M(T)\) measurements were performed to study the temperature-dependent magnetic behaviour of samples and to determine the martensitic-transition characteristic temperatures. These experiments were combined with DTA. ZFC–FC thermomagnetization in a field of 0.01 T and frequency-dependent magnetic susceptibility were measured with frequencies \(f=1, 97, 496, 997\) and 1,488 Hz in an AC magnetic field of 4 Oe after ZFC from 300 K.

To accurately derive the magnetic-entropy changes (\(\Delta S_m\)) at magnetostuctural transitions with thermal hysteresis, the so-called loop process method was adopted to get the isothermal magnetization curves. The isothermal \(M(B)\) curves were measured in fields of up to 5 T upon cooling with a temperature interval of 2 K. Before each isothermal magnetization, the samples were all the way heated up to the complete PM austenite region (100 K away from the magnetostuctural transitions) with heating rate of 5 K min\(^{-1}\) to eliminate the history-dependent magnetic states and then cooled back to the targeted measurement temperature points. All these temperature loops were performed in zero field. To avoid the overmuch temperature-induced FM martensite phase during approaching each targeted temperature point, the cooling rate was set as 2 K s\(^{-1}\). Besides, a waiting time of 300 s was compelled before the measurements to guarantee a highly stable temperature. The \(\Delta S_m\) was then derived from the resulting magnetization curves using the Maxwell relation:

\[
\Delta S_m(T,H) = S_m(T,H) - S_m(T,0) = \int_0^H \frac{\partial M}{\partial T}\, dH
\]

**ELF calculations.** ELF calculations were carried out using the pseudopotential method with plane-wave-basis set based on the density-functional theory. The electronic exchange correlation energy was treated under the local spin density approximation. Plane-wave cutoff energy of 770 eV and 126 (13×13×12) k points in the irreducible Brillouin zone were used for a good convergence of the total energy. The absolute total-energy difference tolerance for the self-consistent field approximation electronic exchange correlation energy was treated under the local spin density approximation.

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Author contributions

Experiments were conceived by E.K.L. and W.H.W. and carried out by E.K.L., L.F., W.Z. and G.J.L. with assistance of J.L.C. Results were analysed and interpreted by E.K.L., W.H.W., L.F., W.Z., G.J.L., G.H.W., H.W.Z. and F.R.deB. ELF calculations were performed by E.K.L., C.B.J. and H.B.X. The manuscript was written by E.K.L., G.H.W., W.H.W. and F.R.deB. W.H.W., H.W.Z. and G.H.W. are responsible for project direction, planning and infrastructure.

Additional information

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