Decomposition of High Temperature $\gamma_{\text{Mn}}$ Phase during Continuous Cooling and Resultant Damping Behavior in Mn$_{74.8}$Cu$_{19.2}$Ni$_{4.0}$Fe$_{2.0}$ and Mn$_{72.4}$Cu$_{20.0}$Ni$_{5.6}$Fe$_{2.0}$ Alloys

Fuxing Yin*, Yoshiaki Ohsawa*, Akira Sato* and Kohji Kawahara**

*National Research Institute for Metals, Tsukuba, Ibaraki 305-0047, Japan
**B. B. Materia Co. Ltd., Chiba, Chiba 267-0066, Japan

The temperature dependent changes of logarithmic decrement $\delta$ and Young's modulus $E$ were measured for continuously cooled Mn$_{74.8}$Cu$_{19.2}$Ni$_{4.0}$Fe$_{2.0}$ and Mn$_{72.4}$Cu$_{20.0}$Ni$_{5.6}$Fe$_{2.0}$ (at%) alloys, after solid solution holding at 1173 K. Microstructure of these alloys, cooled to room temperature in different rates, was characterized by X-ray diffraction and electron microscopy. Ni apparently inhibited the decomposition of high temperature $\gamma_{\text{Mn}}$ phase during cooling, so lower fcc-fct transformation temperatures ($T_c$) were observed in Mn$_{72.4}$Cu$_{20.0}$Ni$_{5.6}$Fe$_{2.0}$ alloy. Besides, several fct phases with different lattice parameters or $T_c$ temperatures, i.e. an appearance of compositional heterogeneity in the Mn-enriched fraction, were also manifested in the alloy with Ni of 5.6 at%. Large amounts of (011) twinning plates were observed as a typical feature in the microstructure of Mn$_{74.8}$Cu$_{19.2}$Ni$_{4.0}$Fe$_{2.0}$ alloy. However, tetragonal distortion caused mis-fitting at the twinning boundaries resulted in a rapid reduction of logarithmic decrement in the vicinity of 273 K. Meanwhile, a larger amount of precipitated Cu-enriched phase confined the formation and growth of twinning plates in the fct phases, and hence reduced logarithmic decrements. In contrast, a twinned microstructure in Mn$_{72.4}$Cu$_{20.0}$Ni$_{5.6}$Fe$_{2.0}$ alloy sustained the high damping capacity in a wide temperature range near 273 K, irrespective of the cooling rates.

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I. Introduction

The high damping capacity of MnCu based alloys has been attributed to the mobility of internal boundaries, such as the (110) twin boundaries and phase boundaries, which appear when alloys are aged within the metastable miscibility gap of $\gamma_{\text{Mn}}$ solution$^{[10]}$. The formation of (110) twinning microstructure, usually found at room temperature in quenched MnCu alloys with Mn content of >82 at%, seems to be induced by the antiferro-magnetic transition in $\gamma_{\text{Mn}}$ phase at the $T_N$ temperature. As a result, many researchers have tried to relate the changes of damping capacity with temperature or strain amplitude to the magnetostriuctive motion of domain boundaries$^{[3]})$. In alloys with Mn content of <82 at%, aging treatment seems necessary to obtain a Mn-enriched matrix phase, in which antiferro-magnetic transition and the fcc-fct lattice distortion may occur at higher temperatures above room temperature$^{[4]}$. However, the distribution, composition and relative amount of the multiphase, decomposed from the original $\gamma_{\text{Mn}}$ phase in MnCu alloys, as well as the phase decomposition mechanism, have never been defined clearly.

Both SONOSTON and INCRAMUTE alloys are commercially developed MnCu based high damping alloys, with Al content of 2–4% (mass%). In order to improve the workability of the alloys, Kawahara et al. have developed a Ni added high damping Mn$_{72}$Cu$_{20}$Ni$_{5}$Fe$_{2}$ (at%) alloy, which shows both higher mechanical properties and a large damping capacity in a slowly cooled condition after solid solution treatment at 1173 K$^{[5]}$. In studying the aging process of MnCuNi ternary alloys, Sugimoto et al. proposed a Cu-cluster formation model and suggested that higher Ni content in the alloys suppress the enrichment of Mn in the matrix phase during the aging treatment$^{[6]}$. Nevertheless, neither direct evidence was provided to describe the formation of Cu-clusters during aging process, nor the effects of Ni on the phase transition on resultant microstructure was clarified for the Mn-enriched $\gamma_{\text{Mn}}$ phase.

The $T_N$ temperature of the Mn–Cu alloys is often assumed to be coincident with the tetragonal distortion (fcc-fct) temperature, $T_c$. However, Hicks et al. reported that the fcc $\gamma_{\text{Mn}}$ phase Mn–Ni alloys with Ni of >15 at% transforms into fct structure of $d/a > 1$, at a much lower temperature below the $T_N$ temperature$^{[7,8]}$. Obviously the effects of Ni on the magnetic structure and the fcc-fct transformation mode of $\gamma_{\text{Mn}}$ phase are different from those of Cu. Even the segregation of Ni to 15 at% seems difficult in MnCu based alloys, the influences of Ni on the decomposition of high temperature $\gamma_{\text{Mn}}$ phase and the microstructure in the Mn-enriched $\gamma_{\text{Mn}}$ phase are inevitable in MnCuNiFe alloys. The purpose of this work is to clarify the effects of Ni content and cooling rate on the microstructure at room temperature, as well as the temperature dependent damping behavior in two MnCuNiFe alloys, which are treated by continuous cooling after solid solution holding.
II. Experimental Procedure

Alloys with the compositions of Mn$_{72.4}$Cu$_{19.2}$Ni$_{5.0}$Fe$_{2.0}$ and Mn$_{72.4}$Cu$_{20.0}$Ni$_{5.0}$Fe$_{2.0}$ (at%), were prepared by induction melting pure metals in argon atmosphere. The ingots of 1 kg weight were forged and rolled to thin plates of 1 mm in thickness, after annealing at 1173 K for 1 h. Sample are finally filled in evacuated quartz tubes, and solid solution hold at 1173 K for 1 h, followed by water quenching or continuous cooling to room temperature in 10, 50 and 100 h, respectively.

An ULVAC IFT-1500 internal friction instrument was used to determine the temperature-dependent changes of logarithmic decrement $\delta$ and Young’s modulus $E$. Electrostatic vibration is excited on the cantilever resonant-bar samples in the dimension of $1 \times 10 \times 100 \text{ mm}^3$, and the resonant vibration frequency and a nominal surface strain were found to be 170 Hz and $6 \times 10^{-6}$, respectively. The logarithmic decrement $\delta$ was calculated by counting the vibration cycles $N$ for the displacement amplitude of the free side to decrease from $A_0$ to $A_n$, which was initially as $\exp(0.3\pi)$. On the other hand, Young’s modulus $E$ is obtained from the measured average damping frequency of vibration waves and known sample dimension. The temperature-dependent changes of $\delta$ and $E$ were simultaneously recorded during heating samples at a rate of 0.5 K/min from 233 to 473 K, in a quartz tube evacuated to about 2.6 Pa.

Microstructural features for the samples, heat treated in various conditions, were characterized with a RIGAKU Rint-2500 X-ray diffractometer operated at 40 kV and 250 mA, and a JEOL TEM2000FX transmission electron microscope operated at 200 kV. The foil samples were prepared by jet thinning 3 mm discs ground to 80 $\mu$m thickness initially, in a solution of 15 g chromic acid in 4 mL water and 100 mL glacial acetic acid at 40 V.

III. Results and Discussion

1. Microstructure and damping properties of the water quenched MnCuNiFe alloys

According to the Cu–Mn binary phase diagram$^{(9)}$, solid solution holding at 1173 K retains 5–40 at% Cu in a single $\gamma_{\text{Mn}}$ phase region. Figure 1 shows the X-ray diffraction pattern for the Mn$_{72.4}$Cu$_{20.0}$Ni$_{5.0}$Fe$_{2.0}$ alloy, water quenched to room temperature after holding at 1173 K for 1 h. A single fcc phase is observed in the water quenched microstructure, and the lattice constant is found to be 0.3746 nm. Figure 2 compares the 200 diffraction peak of the alloy with that of Mn$_{72.4}$Cu$_{19.2}$Ni$_{5.0}$Fe$_{2.0}$ alloy in the same quenched condition. It is obviously indicated that decreased Ni content by 1.6 at% contracts the lattice constant of $\gamma_{\text{Mn}}$ to 0.3741 nm. Cowlam et al. established an empirical formula between the lattice constant of fcc $\gamma_{\text{Mn}}$ phase and Cu content in MnCu binary alloys$^{(10)}$,

$$a_{\text{MnCu}}(\text{nm}) = 0.37236 + 0.02182x_{\text{Cu}}$$ (1)

where $x_{\text{Cu}}$ is the atomic content of Cu with $0 < x_{\text{Cu}} < 0.4$. The effect of Ni in a small content ($<10$ at%) on the lattice constant of MnCu based $\gamma_{\text{Mn}}$ phase is feasible supposed to be linear also, while the effect of Fe is considered to be the same as that of Mn. Therefore, using the composition and lattice constants of the two alloys, as well as the constant of 0.37236 in eq. (1), a formula for the lattice constant of $\gamma_{\text{Mn}}$ phase in MnCuNiFe alloys is therefore deduced as:

$$a_{\text{MnCuNiFe}}(\text{nm}) = 0.37236 + 0.0057x_{\text{Cu}} + 0.0169x_{\text{Ni}}$$ (2)

where $x_{\text{Cu}}$ and $x_{\text{Ni}}$ are the atomic ratios of Cu and Ni in the $\gamma_{\text{Mn}}$ phase of the quenched MnCuNiFe alloys. It is obviously demonstrated that the expanding effect of Cu on the unit cell volume of fcc $\gamma_{\text{Mn}}$ phase is extensively decreased with the addition of Ni.

Figure 3 shows the changes of Young’s modulus $E$ and logarithmic decrement with temperature for the water quenched Mn$_{72.4}$Cu$_{19.2}$Ni$_{5.0}$Fe$_{2.0}$ and Mn$_{72.4}$Cu$_{20.0}$Ni$_{5.0}$Fe$_{2.0}$ alloys. The minimum of elastic modulus appears clearly at 253 and 237 K respectively for the two alloys, which are then defined as the fcc-fct transformation temperatures ($T_c$) of the water quenched alloys. For binary MnCu alloys, a linear relations between the $T_c$ temperature and atomic content of Cu was established by Sugimoto$^{(11)}$:

$$T_{\text{MnCu}}(K) = 418 - 1227.4x_{\text{Cu}}$$ (3)

where $x_{\text{Cu}}$ is the atomic content of Cu ($0 < x < 0.4$) in MnCu $\gamma_{\text{Mn}}$ phase. Similarly, the $T_c$ temperature of the $\gamma_{\text{Mn}}$
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\[ T_t = 418 - 1130.8x_{Cu} - 345.9x_{Ni} \]  \hspace{1cm} (4)

With formulae (2) and (4), one can find that the effects of Ni on the unit cell volume and \( T_t \) temperature of \( \gamma_{Mn} \) phase in MnCuNiFe alloys are the similar to those of Cu, but to different extents.

Beside the difference in the \( T_t \) temperature, the relative decreases of Young's modulus \( E \) in the temperature range of 473 K - \( T_t \) are also significantly varied in the two alloys. However, in binary MnCu alloys with different Cu content, the \( T_t \) temperature changes sensitively with Cu content, but the relative reduction of elastic modulus \( E \) is nearly unchanged\(^{11}\). Apparently, Ni has possibly decreased the saturation magnetotrictic constant of MnCu \( \gamma_{Mn} \) phase, which influences the softening of \( \gamma_{Mn} \) lattice before transformation. Also shown in Fig. 3 is a small logarithmic decrement peak at 237 K for Mn\(_{74.8}\)Cu\(_{19.2}\)Ni\(_{5.6}\)Fe\(_{2.0}\) alloy, accompanying the rapid changes of \( E \) in the temperatures above the \( T_t \) temperature. Because this damping peak occurs approximately at the \( T_N \) point on the higher temperature side of \( T_t \), it may originate from the appearing or disappearing process of magnetic domain boundaries in the \( \gamma_{Mn} \) phase. While in the water quenched Mn\(_{72.4}\)Cu\(_{20.0}\)Ni\(_{5.6}\)Fe\(_{2.0}\) alloy, two damping peaks with much smaller magnitudes are observed at 293 and 261 K, respectively.

2. Phase decomposition induced multi-phase microstructure in Mn\(_{74.4}\)Cu\(_{19.2}\)Ni\(_{5.6}\)Fe\(_{2.0}\) and Mn\(_{72.4}\)Cu\(_{20.0}\)Ni\(_{5.6}\)Fe\(_{2.0}\) alloys during slow cooling after solid solution holding at 1173 K

When the alloys are cooled continuously to room temperature in 10 or 100 h from 1173 K, the relative reduction of Young’s modulus in the temperature range from 473 to 233 K increases significantly in both alloys, in contrast to those in water quenched condition. As shown in Fig. 4, decreasing of Young's modulus \( E \) starts in the vicinity of 453 K irrespective of the cooling rate and alloy composition. However, Mn\(_{72.4}\)Cu\(_{20.0}\)Ni\(_{5.6}\)Fe\(_{2.0}\) alloy is observed to have a larger relative reduction of Young’s modulus \( E \) in both 10 and 100 h cooled samples, in contrary to the water-quenched specimens. It is also noted that long cooling time obviously increases the \( T_t \) temperature, but in the same time broadens the minimum peaks of Young’s modulus \( E \).

Figure 5 summarizes the changes of the \( T_t \) temperature and the Vickers hardness of the two alloys, cooled to room temperature in different rates from 1173 K. Both the \( T_t \) temperature and hardness increase linearly with the logarithm of cooling time. It has been clarified that the increase of the \( T_t \) temperature of MnCu alloys is induced by the enrichment of Mn in the matrix \( \gamma_{Mn} \) phase, and that consequently Cu-enriched phase should appear in the slowly cooled alloys\(^{12}\). There is no doubt that the phase decomposition process of the high temperature \( \gamma_{Mn} \) phase, usually studied for aging treatment, proceeds also during continuous cooling. With longer cooling time, the enrichment of Mn in the matrix \( \gamma_{Mn} \) phase is intensified, simultaneously more Cu-enriched phase precipitates, and adequately results in the hardened microstructure. Therefore, a multi-phase microstructure is expected to occur in the slowly cooled MnCuNiFe alloys.

Figure 6 shows the 111 and 200 X-ray diffraction profiles for Mn\(_{74.4}\)Cu\(_{19.2}\)Ni\(_{5.6}\)Fe\(_{2.0}\) and Mn\(_{72.4}\)Cu\(_{20.0}\)Ni\(_{5.6}\)Fe\(_{2.0}\) alloys, cooled to room temperature in 10 and 100 h, respectively. Since the \( T_t \) temperatures have been manifested higher than room temperature for the Mn-enriched \( \gamma_{Mn} \) phases in the alloys, 200 diffraction is splitted into 200 and 002 peaks at room temperature as the result of fcc-fct phase transformation at the \( T_t \) temperature. A
sub-peak is observed on the low diffraction angle side of 111 diffraction, which is attributed to a phase with a larger unit cell, which is tentatively considered as the reflection of a retained fcc γ_{Mn} phase. According to the empirical relation for the lattice parameter of γ_{Mn} phase and Mn content in MnCu binary alloys, fitted by Cowlam et al., Cu-enriched γ_{Mn} phases with Cu content of >70 at% have much smaller lattice constants than the Mn-enriched γ_{Mn} phase, and should demonstrate a diffraction peak at the high angle edge of the main peak. Unfortunately, such diffraction peaks have not been detected in the diffraction experiment, probably because of the fine precipitation undetectable for normal X-ray diffraction.

Table 1 summarizes the characteristic features of the matrix phases in the slowly cooled MnCuNiFe alloys. As derived from the X-ray diffraction patterns of Fig. 6, a fct phase and a fcc phase co-exist in the matrix fraction.

Fig. 4  Temperature dependence of Young’s modulus $E$ in Mn$_{74.4}$Cu$_{19.2}$Ni$_{4.0}$Fe$_{2.0}$ and Mn$_{72.4}$Cu$_{20.0}$Ni$_{5.6}$Fe$_{2.0}$ alloys, cooled from 1173 K to room temperature in 10 and 100 h, respectively.

Fig. 5  The $T_c$ temperatures, detected by the minimum $E$ positions, and Vickers hardness $H_v$ in Mn$_{74.4}$Cu$_{19.2}$Ni$_{4.0}$Fe$_{2.0}$ and Mn$_{72.4}$Cu$_{20.0}$Ni$_{5.6}$Fe$_{2.0}$ alloys, increasing with the decrease of cooling rate from 1173 K to room temperature.

Fig. 6  Slowly scanned 111 and 200 X-ray diffraction peaks of the 10 and 100 h cooled Mn$_{74.4}$Cu$_{19.2}$Ni$_{4.0}$Fe$_{2.0}$ and Mn$_{72.4}$Cu$_{20.0}$Ni$_{5.6}$Fe$_{2.0}$ alloys, showing the multi-phase microstructure at room temperature.
of Mn_{44.8}Cu_{19.2}Ni_{4.0}Fe_{2.0} alloy, which have different lattice parameters in 10 and 100 h cooling condition. In contrast, three phases are found in the matrix fraction of Mn_{72.4}Cu_{20.0}Ni_{4.0}Fe_{2.0} alloy, all of which become fct structured in the 100 h cooling condition. In order to estimate the Cu, Ni contents of the matrix phases in the slowly cooled alloys, the dominant phases are indicated in Table 1, which are adopted to the Tt temperatures measured by the changes of Young's modulus E. Since formula (2) supposes a fcc γ_{m} phase that correlates the lattice constant to Cu, Ni contents linearly, the equivalent fcc lattice constant a_{0} of the fct phase at room temperature is evaluated by means of high temperature X-ray experiment on the changes of a with temperature above Tt. Using the formulae (2) and (4), Cu and Ni contents are calculated for the dominant phases in the matrix phases of slowly cooled MnCuNiFe alloys, and the results are listed in Table 1, together with the correspondent a_{0} data.

The temperature dependent changes of logarithmic decrement δ are shown in Fig. 7, for the two alloys cooled to room temperature in 10 and 100 h, respectively. It is found that rapid increases of δ with decreasing temperature are usually initiated at a temperature higher than the Tt temperature. The two-peak mode of δ in 10 h cooled Mn_{44.8}Cu_{19.2}Ni_{4.0}Fe_{2.0} alloy indicates that two possible internal interfaces, i.e. magnetic domain boundaries and fcc-fct transformation induced (011) twinning boundaries, are involved in the vibrating attenuation processes, attaining the maximum magnitudes at different temperatures. As mentioned for water quenched alloys, minor damping peaks at temperatures above Tt have also appeared in the slowly cooled condition. In the same 10 h cooling condition, different Cu, Ni contents in the matrix phases of the two alloys produce an obvious deviation in domain boundary damping peaks, and also the difference in twinning boundary damping peaks.

![Fig. 7 Temperature dependent changes of logarithmic decrement δ in Mn_{44.8}Cu_{19.2}Ni_{4.0}Fe_{2.0} and Mn_{72.4}Cu_{20.0}Ni_{4.0}Fe_{2.0} alloys, cooled from 1173 K to room temperature in 10 and 100 h, respectively. Arrows in the figures indicate the correspondent Tt temperatures and the dashed curve shows a domain boundary related damping peak near 313 K.](image)
in the vicinity of 273 K. A flattened damping peak is found in the 10 h cooled Mn$_{24}$Cu$_{20}$Ni$_{8}$Fe$_{2}$ alloy, which is not only contributed by the neighboring of the two damping peaks, also related to the twinning features in the dominant fct1 and fct2 phases. The $T_t$ temperature of Mn$_{48}$Cu$_{19}$Ni$_{4}$Fe$_{2}$ alloy is increased to about 400 K in 100 h cooled condition, the clearly divided domain boundary and twinning boundary damping peaks become decreased obviously than those in 10 h cooled condition. In contrast, three fct phases with different $c/a$ parameters are formed in the matrix fraction of 100 h cooled Mn$_{24}$Cu$_{20}$Ni$_{8}$Fe$_{2}$ alloy. And corresponding to the two $T_t$ temperatures at 373 and 333 K indicated by the minimums of Young's modulus $E$, two domain boundary damping peaks are found above 313 K. Meanwhile, the twinning damping peak in 100 h cooled Mn$_{24}$ Cu$_{20}$Ni$_{8}$Fe$_{2}$ alloy is further broadened and sustained at a high level, irrespective of the largely deviated composition in the three fct phases. It is therefore summarized that the mixture of several phases with different composition and lattice parameters in the matrix fraction seems beneficial to acquire higher and stable damping capacity.

3. TEM analysis of the room temperature microstructure formed in the slowly cooled MnCuNiFe alloys

As described in Table 1, multi-phase room temperature microstructure forms easily in slowly cooled MnCuNiFe alloys, especially for the alloy with a higher Ni content. It is considered that the fct phases, which have $T_t$ temperatures higher than room temperature, occupy the dominant fraction in the Mn-enriched matrix of the MnCuNiFe alloys. Figure 8 shows the typical twinning microstructure in 10 h cooled Mn$_{48}$Cu$_{19}$Ni$_{4}$Fe$_{2}$ alloy. A lot of (011) twinning plates of about 100 nm in width seem to form separately in the matrix fraction. If those plates are just in the (011) twinning orientation to the matrix fraction, the reflection zones for the two parts should belong to the same $\{hkl\}$ group. However, electron diffraction patterns manifest a (031) zone normal of the plates, while the matrix fraction is tilted along (012)*. Obviously, those plates have rotated several degrees from the initial (021)* twinning orientation along [100] direction. Such orientation deviation of the twinning plates is induced by the increased tetragonal distortion of the fct phases with the decrease of temperature. As shown in Fig. 9, a coherent (011) plane is supposed to separate two twinned fractions at a temperature just below $T_t$. Accompanying the decreases of temperature, the $c/a$ becomes smaller than 1, and the initially parallel (011) planes respectively in the two fractions then rotate in opposite directions by an angle of $45 - \tan^{-1} (c/a)$. In order to sustain the coherent (011) interface, a relative re-orientation between the two fractions seems inevitable, such accommodation process will naturally relax some of the distortion strains in the matrix fraction, simultaneously destroy the present twinning relationship. While the temperature dependent behavior of mobile twinning boundary damping has been explained by a thermal activation process, destruction of the original twinning orien-

Fig. 8 Typical (011) twinning plates with boundary fringes, observed in 10 h cooled Mn$_{48}$Cu$_{19}$Ni$_{4}$Fe$_{2}$ alloy. Electron diffraction patterns of (012)* and (031)* zones suggest that the plates deviate from the complete twinning orientation.

Fig. 9 A sketch model of the mis-fitting of (011) twinning plane. The rotation angle becomes greater with larger tetragonal distortion in the fct phase.
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Figure 10 presents a dark field micrograph of the microstructure in 100 h cooled Mn$_{32.4}$Cu$_{20.0}$Ni$_{15.4}$Fe$_{2.0}$ alloy in (100)* orientation. The elliptic spreading of the (011) and (111) reflection spots is corresponding to the relative rotation of the (011) twinning plane, due to the lower c/a of the matrix fct phase. It is also observed that the (011) and (001) twinning plates, of only about 20 nm in width, in the Mn-enriched fct phase demonstrate a perpendicularly intersecting configuration. Because both the fcc Mn-enriched phase ascertained by X-ray diffraction and the expected Cu-enriched $\gamma_{\text{Mn}}$ phase have the similar crystal structure and lattice constant to the matrix fct phase, characterization of these phases seems impossible by ordinary TEM method. Meanwhile, another set of diffraction spots, that occur at the inner of the matrix pattern, indicate that MnO phase is formed during preparing of TEM foils. It is considered that the large amount of phase decomposition products retained in the microstructure not only interfere the movement of domain boundaries but also limit the growth of twinning plates on one twinning plane. As a result, the intersected thin twinning plates are induced along (011) and (001) planes, respectively, and therefore the constrained mobility of the twinning boundaries produces a lower damping capacity.

When Ni content is increased to 5.6 at% in Mn$_{72.4}$Cu$_{20.0}$Ni$_{15.4}$Fe$_{2.0}$ alloy, the phase decomposition of high temperature $\gamma_{\text{Mn}}$ phase is obviously inhibited, consequently the Cu content in the fct matrix phases is still remained at a level of 15 at%. In the same time, compositional heterogeneity in matrix fraction is clearly manifested by both X-ray diffraction results and the minimums of Young’s modulus $E$. Figure 11(a) shows a tweed microstructure, commonly found in 10 h cooled Mn$_{32.4}$Cu$_{20.0}$Ni$_{15.4}$Fe$_{2.0}$ alloy. Vitek and Warlimont explained that the appearance of tweed microstructure in aged Mn–30Cu (at%), and the absence in Mn–20Cu (at%) alloy, was induced by the different proportions of Cu-enriched phase in the microstructure. However, it is difficult to expect that more Cu-enriched phase has formed during 10 h cooling in Mn$_{72.4}$Cu$_{20.0}$Ni$_{15.4}$Fe$_{2.0}$ alloy than that in Mn$_{44.8}$Cu$_{19.2}$Ni$_{15.4}$Fe$_{2.0}$ alloy, because of the higher Cu contents in the fct matrix phase of Mn$_{32.4}$Cu$_{20.0}$Ni$_{15.4}$Fe$_{2.0}$ alloy. As is confirmed by Fig. 10, a great amount of Cu-enriched phase exists in the 100 h cooled Mn$_{14.4}$Cu$_{9.2}$Ni$_{15.4}$Fe$_{2.0}$ alloy, but only a slightly mottled morphology is found in the dark field micrograph. Therefore, the tweed microstructure observed in Mn$_{32.4}$Cu$_{20.0}$Ni$_{15.4}$Fe$_{2.0}$ alloy, instead of the twinning plates in Mn$_{44.8}$Cu$_{19.2}$Ni$_{15.4}$Fe$_{2.0}$ alloy, should be produced by the fct phases with different Ni contents. Because the twinning microstructure has scarcely been observed in the Mn$_{32.4}$Cu$_{20.0}$Ni$_{15.4}$Fe$_{2.0}$ alloy, the stable damping behavior in the vicinity of 273 K is characteristic of the regular compositional modulation in the fct matrix fraction in a scale of about 15 nm.

Figure 11(b) shows the coarsened tweed microstructure observed in 100 h cooled Mn$_{32.4}$Cu$_{20.0}$Ni$_{15.4}$Fe$_{2.0}$ alloy. The three fct phases seem to have nearly the same proportions so that the correspondent (002) diffraction peaks present the approaching intensities. Moreover, the electron diffraction spots beside the (022) spots are analyzed to be the (022) reflections of MnO phase. Because several Mn-
enriched phases have formed during the slow cooling, a certain amount of Cu-enriched phase should exist in the room temperature microstructure. Even the coherency strains between the fct phases as that shown in Fig. 11(a) have disappeared in 100 h cooled microstructure, the damping behavior in the temperatures near 273 K has rarely changed, as compared with the that in 10 h cooling condition. In the same time, the double domain boundary damping peaks, relating to the fct phases with different $T_{1}$ temperatures, spread the high damping state to higher temperatures. It is clarified that the resulted matrix fraction is constituted by small fct structured regions with different Cu, Ni contents, but with similar lattice parameters. The growth of (011) twinning plates is constrained in the regions, and little frictional forces against the movement of the twinning boundaries exist during the vibrating attenuation process in these fine regions. A similar mechanism was proposed for the aged Cu based INCRAMUTE damping alloy, where higher damping capacity was attributed to the (011) twins appeared in Mn-enriched regions of about 20 nm large$^{(2)}$.

IV. Conclusions

Logarithmic decrement $\delta$ and Young’s modulus $E$ are measured with increasing temperature in Mn$_{72.4}$Cu$_{19.2}$Ni$_{2.0}$Fe$_{2.0}$ and Mn$_{72.4}$Cu$_{20.0}$Ni$_{3.6}$Fe$_{2.0}$ (at%) alloys, while the alloy are continuous cooled from 1173 K to room temperature in 10 and 100 h, respectively. The variation of logarithmic decrement with testing temperature, i.e. the overlapping of the (011) twinning boundary damping peak and the magnetic domain boundary damping peaks, seems sensitive to both the alloy composition and the cooling rate. The decomposition process of high temperature $\gamma_{Mn}$ phase during slow cooling, which are affected by the alloy composition and cooling rate, determines the final constituents of microstructure, and therefore the temperature-dependent damping behavior.

(1) Referring to the lattice constants and $T_{1}$ temperature data measured in water quenched alloys, the lattice constant of Mn-enriched $\gamma_{Mn}$ phase and correspondent $T_{1}$ temperature are linearly related to the Cu, Ni contents in the phase. The slowly cooled MnCuNiFe alloys principally contain the fct structured Mn-enriched matrix phases, a fcc phase retained from the high temperature $\gamma_{Mn}$ phase, and a Cu-enriched phase at room temperature.

(2) Higher Ni content in Mn$_{72.4}$Cu$_{20.0}$Ni$_{3.6}$Fe$_{2.0}$ alloy is clearly demonstrated to inhibit the phase decomposition of high temperature $\gamma_{Mn}$ phase during slow cooling, and as a result, lower $T_{1}$ temperatures, i.e. lower Mn content, is observed in the Mn-enriched matrix phases. On the other hand, the higher Ni content tends to induce a compositional heterogeneity in the matrix fraction, and several fct fractions with different lattice parameters and $T_{1}$ temperatures are characterized in the microstructure of the 10 h cooled Mn$_{72.4}$Cu$_{20.0}$Ni$_{3.6}$Fe$_{2.0}$ alloy.

(3) The position and magnitude of the magnetic domain boundary damping peaks are characteristic of the individual fct phases in the matrix fraction, depending on the mobility of the domain boundaries in each phase. (011) twinning plates are the typically found in the 10 h cooled Mn$_{72.4}$Cu$_{19.2}$Ni$_{4.0}$Fe$_{2.0}$ alloy. The increasing tetragonal distortion with decreasing temperature may cause a mis-fitting at (011) twinning boundaries, and therefore result in the rapid reduction of $\delta$ below 273 K. More Cu-enriched phase is precipitated during 100 h cooling, so that the formation and growth of (011) twinning plates seems confined in the single fct phase and the damping capacity is decreased. On the other hand, a tweed microstructure is demonstrated by the fct phases in both 10 and 100 h cooled Mn$_{72.4}$Cu$_{20.0}$Ni$_{3.6}$Fe$_{2.0}$ alloy. Even no twinning plates have been observed in the alloy, a high but stable damping behavior is demonstrated under both cooling conditions.

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