Damping Capacity of Graphite-Dispersed Composite Steel *

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Composite steels consisting of an iron-graphite sintered core and steel were prepared by powder metallurgy and their damping capacities were examined. The ferritic grain size in the sintered core decreases with increasing carbon (graphite) content, resulting in a decrease in internal friction due to magnetic domain wall movement. Conversely, internal friction due to the deformation of graphite increases with carbon content. The internal friction of graphite-dispersed composite steel reaches a maximum at a carbon content of approximately 5 mass%. When carbon content exceeds 12 mass%, delamination between the steel part and the sintered core occurs easily and a good composite steel cannot be obtained. Experimental results suggest that a high damping capacity can be provided by appropriate control of the microstructure of the steel part and the ferritic grain size and carbon content of the sintered core.

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1. Introduction

Recently, social demands for reducing the vibration and noise of machinery and structures have become stronger, and thus commercial machinery and structures require high-damping materials. Among steels, high-damping materials are divided into two classes: 1) high-damping laminate steels whose damping capacity is increased many times over by incorporation of sheet steel into the viscoelastic damping layer, such as by binding or sandwiching; and 2) high-damping alloys whose damping capacity is increased by alloying. Since high-damping laminate steel contains resin, the temperature for application is limited and many problems arise in welding, forming, and recycling. On the other hand, ferromagnetic types of Fe-Cr¹⁾ and Fe-Al²⁾ system alloys are well known as high-damping alloys. In order to realize easy movement of the magnetic domain wall, ferritic grains are usually coarsened. However, this coarsening of grains leads to reduced strength and toughness. Also, damping capacities are reduced considerably by welding and forming. Several reports³⁻⁷⁾ address reversible martensite-type Fe-Mn system alloys. However, these alloys have poor formability and are not utilized widely.

Takahashi *et al.*⁸⁾ examined the relationship between the damping capacity of mild-steel sheet and heat treatment and found that removing nearly all interstitial solution atoms (such as C and N) and coarsening the ferritic grain size increases the damping capacity to the level exhibited by high-Cr steel. However, this method reduces strength and toughness. Furthermore, the damping capacity of mild steel prepared by this method is markedly reduced by even a small amount of cold working.

Flaky graphite cast iron exhibits a high damping capacity due to the presence of easily-deformed graphite, and has had long and extensive use as a damping material. Graphite is absent in steels, except in special cases such as graphitic steels.⁹⁻¹²⁾ If graphite can be dispersed in the center of a steel,

the damping capacity may be increased without deterioration of the characteristics of the steel. However, there has been no report regarding graphite-dispersed composite steel. Since such composite steel contains neither special alloying elements nor resin, the environmental load over the course of production, use, and recycling would be small. Therefore, realization of such steel is desirable from the viewpoint of providing environment-conscious materials (ecomaterials).

In the present study, graphite-dispersed composite steels were prepared by powder metallurgy and their damping capacities were examined.

2. Experimental Procedure

Powder mixtures of iron (99.5 mass% purity, 5 μm particle size) and graphite (99.9 mass% purity, < 78 μm particle size) were inserted into a mild-steel (SS400) tube, degassed at 873 K in a vacuum, and sealed by welding. The resultant capsule was hot-rolled in order to prepare a composite steel consisting of a mild-steel part and an iron-graphite sintered core. When hot-rolling is performed at a temperature higher than the eutectoid transformation temperature (A_1 : 1000 K), carbon atoms dissolve into austenite during the rolling process, and many hard carbide particles precipitate during the subsequent cooling process. In order to avoid this phase transformation, hot-rolling was performed at 953 K, which is just below A_1 . Carbon (graphite) content was varied among 0, 2, 4, 6, 8, 10, and 12 mass%. The obtained composite steels were labeled Fe, Fe-2C, Fe-4C, Fe-6C, Fe-8C, Fe-10C, and Fe-12C steels, respectively. From these graphite-dispersed composite steels, 1 mm $\times$ 10 mm $\times$ 150 mm specimens for measuring damping capacity were prepared by machining and grinding. The specimens were annealed at 923 K in a vacuum in order to eliminate strain due to grinding and homogenize the hardness and ferritic grain size of the mild-steel part.

Damping capacity was evaluated by measuring internal friction by means of the vibration method. Figure 1 is a schematic diagram of the damping test. Specimens supported

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at sympathetic nodes were vibrated sympathetically by an electromagnet. After vibration was stopped, the free damping amplitude in the center of each specimen was measured by laser light. In order to simplify the comparison of damping capacities between specimens, a constant initial amplitude was employed. Damping tests were conducted at room temperature in air. Figure 2 shows an example of a free damping curve. Internal friction Q^{-1} for various strain amplitudes was obtained from the following equation:

$$Q^{-1} = 1/4\pi \cdot \ln(A_n/A'_n) \quad (1)$$

where A_n represents the amplitude for a given strain amplitude, and A'_n represents the amplitude after four cycles from the set point (see Fig. 2).

3. Results and Discussion

Figure 3 shows the optical microstructure of a transverse section of Fe-6C steel. The center depicts the iron-graphite (black) sintered core, and the upper and lower sides (white parts) depict the mild-steel part. As shown in Fig. 3, the thickness of the sintered core is about 300 μm . The graphite deformed in the rolling direction is distributed uniformly. When the sintered core was etched, no cementite was found. The amount of graphite increases with carbon content up to 10 mass%. However, the microstructure is not significantly changed from that shown in Fig. 3. In the case of the specimen with 12 mass%, the graphite particles connected with each other at the boundary between the mild-steel part and the

sintered core, making the composite steel prone to delamination. Therefore, good composite steel could not be obtained. Thus, when powder metallurgy is applied to prepare graphite-dispersed composite steel such as in the present study, the limitation on the carbon content for obtaining good composite steel may be about 10 mass%.

As described previously, the damping capacity depends on the ferritic grain size. In order to homogenize the hardness and ferritic grain size of the mild-steel part, the specimens were annealed in a vacuum after grinding. As a result, the average surface hardness (load: 4.9 N) for the six kinds of steels from Fe to Fe-10C was found to be HV 107 (maximum $HV = 109$ and minimum $HV = 105$). Figure 4 shows the hardness distribution (load: 1.96 N) with respect to depth from the surface for the longitudinal section of Fe-6C steel. The hardness of the surface is the same as that of the interior. From this, the ferritic grain size of the mild-steel part can be considered to be almost uniform from the surface to the interior. As mentioned below, the ferritic grain size of the sintered core varied with carbon content.

Figure 5 shows the free damping waveform of Fe-6C and Fe-10C steels. The damping waveform changes with carbon content. Based on these free damping waves, internal friction Q^{-1} for various strain amplitudes was obtained. Figure 6 shows the relationship between Q^{-1} and the strain amplitude. For all composite steels, internal friction linearly increases with strain amplitude up to about 1.0×10^{-4} and thereafter shows small increases. However, the internal friction grad-

Fig. 1 Schematic diagram of damping test.

Fig. 2 Schematic illustration of amplitudes A_n and A'_n required in order to evaluate internal friction.

Fig. 3 Optical microstructure of the transverse section of Fe-6C steel.

Fig. 4 Change in Vickers hardness with the depth from the surface for the longitudinal section of Fe-6C steel.

Fig. 5 Damping wave form of Fe-6C and Fe-10C steels.

Fig. 6 Internal friction of composite steels as a function of strain amplitude.

ually decreases when strain amplitude exceeds 1.5×10^{-4} . The proportion of strain amplitude to internal friction at a relatively small strain amplitude suggests that internal friction is caused mainly by magnetic-mechanical static hysteresis due to nonreversible movement of the magnetic domain wall.⁶⁾ As shown in Fig. 6, when strain amplitude exceeds 1.0×10^{-4} , there is no linear relationship. This is because the energy loss due to nonreversible movement of the magnetic domain wall reaches a critical value at this strain amplitude. As shown in Fig. 6, the value of Q^{-1} varies with carbon content. The ferritic grain size of each mild-steel part is similar. Therefore, this change may be caused by microstructural differences between the sintered cores.

Figure 7 shows the change in internal friction with carbon content. For each strain amplitude, the value of Q^{-1} increases rapidly when the carbon content exceeds 2 mass%, reaches a maximum value at about 5 mass%, and decreases with further increases in carbon content.

Fig. 7 Change in internal friction with carbon content in composite steels.

In order to clarify the mechanism of these changes, the microstructure of a longitudinal section of the sintered core was examined with an optical microscope. Figure 8 shows some examples of the results. Figure 9 shows the change in ferritic grain size of the sintered core with carbon content. As is evident from Figs. 8 and 9, the ferritic grain size decreases with increasing carbon content. This is because the growth of ferritic grains is controlled by the dispersed graphite. Thus, as carbon content increases, ferritic grain size decreases and the amount of graphite increases.

Supposing that graphite and ferritic grain size influence the internal friction of composite steel independently, Q^{-1} is given by the following equation:

$$Q^{-1} = Q_F^{-1} + Q_G^{-1} \quad (2)$$

where Q_F^{-1} is the internal friction due to the magnetic domain wall movement of ferrite and Q_G^{-1} is the internal friction due to the deformation of graphite. For the strain amplitude of 1.0×10^{-4} in Fig. 7, Q^{-1} can be divided into Q_F^{-1} and Q_G^{-1} , as shown in Fig. 10. As carbon content increases, the amount of the graphite dispersed into the matrix increases. The graphite arrests the growth of ferritic grains. Therefore, ferritic grain size decreases, resulting in a reduction in Q_F^{-1} . However, because the amount of easily deformed graphite increases, energy is absorbed easily and Q_G^{-1} increases. The maximum value of Q^{-1} at approximately 5 mass% is caused by the influence of both internal frictions, that due to ferritic grain size and that due to graphite. Supposing that ferritic grain size is unchanged and only the amount of graphite is changed, Q^{-1} of such a composite steel may exhibit a curve which is shifted upward to the Fe steel position parallel to the Q_G^{-1} curve. In this case the value of Q^{-1} at 10 mass% reaches 2.6×10^{-2} . This suggests that when the ferritic grain size and the amount of graphite are controlled properly, a composite steel of higher damping capacity can be attained.

The value of the internal friction of Fe-4C steel, which shows the highest damping capacity in the present study, was compared with that of several kinds of iron-based damping alloys. Figure 11 shows the comparison of Fe-4C steel with various kinds of heat-treated mild steels and cast iron.⁸⁾ The value of Q^{-1} of rimmed steel is very low. However, the value

Fig. 8 Optical microstructure of the longitudinal section of sintered region in composite steels. (a) Fe-2C, (b) Fe-6C and (c) Fe-10C.

of Q^{-1} of Ti-killed steel in which C or N is fixed as carbide or nitride is high. In Si-killed steel and Al-killed steel, the high Q^{-1} is attained by significantly reducing interstitial solution atoms by decarburization and coarsening the ferritic grain. The value of Q^{-1} markedly increases with such heat treatments, while strength and toughness decrease. The value of Q^{-1} of Fe-4C steel is similar to those of Si-killed steel and Al-killed steel and considerably higher than those of Ti-killed steel and cast iron.

Figure 12 shows a comparison of Fe-4C steel with ferromagnetic-type Fe-Cr and Fe-Al system alloys and a reversible martensite-type Fe-Mn system alloy.⁷⁾ In the case of the ferromagnetic-type alloy, since internal friction varies with the applied magnetic field, Fig. 12 shows two types of internal friction: with and without application of a magnetic field. The value of Q^{-1} of Fe-4C steel is much greater than that of a ferromagnetic-type alloy. On the other hand,

Fig. 9 Change in ferritic grain size of the sintered core with carbon content.

Fig. 10 Contributions of the damping mechanisms, Q_F^{-1} and Q_G^{-1} , to the damping capacity of composite steels, Q^{-1} .

Fig. 11 Comparison of damping capacity of Fe-4C steel with that of several mild steels and cast iron.

compared with the Fe-Mn system alloy, the Fe-4C steel has higher Q^{-1} at small strain amplitudes. However, when the strain amplitude exceeds 1.5×10^{-4} , the value of Q^{-1} of Fe-4C steel decreases. In contrast, the value of Q^{-1} of the Fe-Mn system alloy continues to increase at high strain amplitudes. Therefore, at high strain amplitudes, the value of Q^{-1} of the Fe-Mn system alloy is greater than that of Fe-4C

Fig. 12 Comparison of damping capacity of Fe-4C steel with that of Fe-Cr, Fe-Al and Fe-Mn alloys.

steel. Nevertheless, in the composite steels, Q^{-1} at high strain amplitude may be improved by controlling the microstructure of the mild-steel part.

4. Conclusions

Graphite-dispersed composite steels were prepared by powder metallurgy and their damping capacities were examined. The results obtained are as follows:

(1) Composite steel consisting of steel and an iron-graphite sintered core is successfully produced by hot-rolling of a steel tube containing iron and graphite powders at a temperature just below the eutectoid transformation temperature. However, when carbon (graphite) content exceeds 12 mass%,

delamination occurs easily at the boundary between the sintered core and the steel part. Therefore, good composite steel cannot be obtained.

(2) The ferritic grain size of the sintered core decreases with increasing carbon content. And thus internal friction due to the magnetic domain wall movement of ferrite decreases. However, internal friction due to the deformation of graphite increases. Therefore, composite steel prepared by powder metallurgy such as in the present study exhibits maximum internal friction at a carbon content of approximately 5 mass%.

(3) A composite steel showing promise as an advanced damping material can be attained by appropriate control of the microstructure of the steel part, and ferritic grain size and carbon content of the sintered core.

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