THE INFLUENCE OF HEAT TREATMENT ON THE MAGNETIC PROPERTIES AND THE MICROSTRUCTURE OF Fe-Ni-Al ALLOYS

by K. J. de VOS

Abstract

It is generally accepted that, for obtaining optimum magnetic properties, Fe-Ni-Al alloys have to be cooled continuously from the homogeneous α region. In the article it is shown that in alloys, lying along the Fe-NiAl line, with less than approximately 45 at. % Fe, the coercivity is appreciably higher after quenching followed by tempering. Electron micrographs demonstrate that the differences in magnetic properties, arising from the two heat treatments, are connected with morphological changes in the α + α' structure.

1. Introduction

Some years ago the author 1) investigated the magnetic properties of a range of alloys of the Fe-Ni-Al system, most of which lay along the Fe-NiAl line; the Fe content was varied between 10 and 80 at. % (fig. 1). The optimum coercivities which could be obtained by continuously cooling from the homogeneous α region are given in fig. 2 (curve a). In terms of shape anisotropy the variation of coercivity with composition could be very well related qualitatively to the changes in morphology in the α + α' microstructure.

As a development from this, it was recently found that the coercivity of these alloys after quenching, followed by an optimum tempering between 600 and 500 °C, varies as shown in fig. 2 (curve b). Efforts were made to relate the observed differences in coercivity to variations in the microstructure of the alloys. The replicating technique is based upon

Fig. 1. Section of the Fe-Ni-Al phase diagram at 750 °C; • alloys under investigation.
selective oxidation of the $\alpha$ and $\alpha'$ phases during electropolishing in a solution of chromium trioxide and glacial acetic acid. The details of this method have been previously described by the author in this periodical 2).

2. Discussion of the experimental results

From fig. 2 it is seen that the curves have a similar shape after the two heat treatments. For both conditions very low coercivity is given with low (10 at. %) and high (80 at. %) Fe content. Only the maximum values differ, in that for the tempering treatment the maximum value occurs at 33 at.% Fe (as compared with 50 at. % for the continuously cooled alloy), this maximum having a higher value (910 Oe as compared with 700 Oe).

From the figure it can further be deduced that Fe-Ni-Al permanent-magnet alloys can roughly be divided into two groups in accordance with their magnetic behaviour. In alloys with more than about 45 at. % Fe, the highest coercive force is obtained after a controlled cooling; on the other hand, in alloys with an Fe content below about 45 at. %, $H_C$ is appreciably higher after quenching followed by tempering at an appropriate temperature.

In 1942 Dannöhl 3) discussed the difference in magnetic properties of some commercial Fe-Ni-Al alloys after a controlled cooling and quenching followed by tempering. On the basis of his modified Fe-Ni-Al phase diagram Dannöhl stated that if the quenched $\alpha$ phase is tempered below 650 °C, only the $\alpha'$ phase will precipitate. During tempering between 650 and 900 °C the $\alpha'$ phase preci
pitates first, the \( \gamma \) phase afterwards. He further stated that during cooling from the high-temperature homogeneous \( \alpha \) region a simultaneous precipitation of the \( \alpha' \) phase and the \( \gamma \) phase will take place. He claimed that this multiple precipitation process ("Mehrfachauflädtung") would produce a denser pattern of high stresses, resulting in a higher coercivity than that obtained by the precipitation of \( \alpha' \) alone, or subsequent precipitation of \( \alpha' \) and \( \gamma \). However, from work undertaken by Bradley \(^4\) it has become obvious that the conceptions of Dannöhl were not correct. In addition to this his experiments were restricted to alloys with about 50 to 65 at.\% Fe, from which he was not able to establish that the magnetic behaviour of alloys with less than about 45 at.\% Fe is contrary to that of the high-iron alloys.

With a view to explaining this phenomenon, electron micrographs have been taken of typical examples of each group, namely of the alloy with 50 at.\% and 36 at.\% Fe, respectively. The \( \alpha + \alpha' \) structure of the alloy with 50 at.\% Fe after continuous cooling at a rate of 7 °C/s, being the normal industrial heat treatment, has been shown in fig. 3. In this case the coercive force amounts to 700 Oe (fig. 2, curve a). It can be seen that the greyish white \( \alpha \) phase forms a continuous matrix, while the \( \alpha' \) phase is precipitated as isolated particles. The shape of these \( \alpha' \) particles is not uniform. Probably they are partly globular.

![Fig. 3. Electron micrograph of an alloy containing 50 at.% Fe, 25 at.% Ni and 25 at.% Al, after optimum cooling.](image-url)
(about 200 Å in diameter), but many neighbouring globules coalesce to form elongated particles. As can be seen in fig. 2 (curve b), quenching of this alloy from the homogeneous \( \alpha \) state followed by tempering at 700 °C results in a lower \( tH_C \) value of approximately 400 Oe. Figure 4 reveals the structure of the alloy after this heat treatment. It differs from the earlier structure in that the Ni-Al phase has lost its continuity while the Fe phase has become continuous. A similar structure arises from heat-treating the quenched alloy at 750 °C (fig. 5). It is quite probable that the lower value of the latter specimens is due to the inversion of the character of the microstructure.

Figure 6 shows the structure of the alloy containing 36 at. % Fe after cooling at a rate of 3 °C/s. Difficulty has been encountered in ascertaining the shape of the \( \alpha' \) precipitate, as the dimensions are very small (less than about 100 Å). On overaging the specimen for 10 min at 800 °C the precipitated \( \alpha' \) particles grow, and the predominant spherical shape becomes visible (fig. 7). Because of the ultrafine and equiaxed nature of the particles, the coercive force can be expected to be rather low (\( \approx 350 \) Oe) after cooling at the above-stated rate.

A completely different structure occurs after quenching followed by tempering during 6 hours at 700 °C (fig. 8). After this treatment the \( \alpha' \) particles are no longer predominantly spherical but for the greater part they show a pronounced elongation along two orthogonal axes which are probably \( \langle 100 \rangle \) directions.

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Fig. 4. Electron micrograph of an alloy containing 50 at. % Fe, 25 at. % Ni and 25 at. % Al after quenching followed by tempering for 4 hours at 700 °C.
Fig. 5. Electron micrograph of an alloy containing 50 at. % Fe, 25 at. % Ni and 25 at. % Al, after quenching followed by tempering for 4 hours at 750 °C.

Fig. 6. Electron micrograph of an alloy containing 36 at. % Fe, 32 at. % Ni and 32 at. % Al, after optimum cooling.
Fig. 7. Electron micrograph of the same specimen as shown in fig. 6, overaged for 10 min. at 800 °C.

Fig. 8. Electron micrograph of an alloy containing 36 at. % Fe, 32 at. % Ni and 32 at. % Al after quenching followed by tempering for 6 hours at 700 °C.
Moreover, the particle size is much greater than given in fig. 6. That is why the coercive force of the material in this condition is appreciably higher (800 Oe).

3. Conclusion

The above results show that the differences in magnetic properties of Fe-Ni-Al alloys arising from variations in heat treatment (controlled cooling or quenching followed by tempering) can qualitatively be interpreted on the grounds of the differences in morphology of the $\alpha + \alpha'$ structure. These are closely connected with the character of the precipitation process and the relative amounts of $\alpha$ and $\alpha'$, as will be described in a later publication. In any case the idea of a third (fcc) phase, as proposed by Dannöhl, must be firmly rejected.

REFERENCES