JOURNAL OF MATERIALS SCIENCE 14 (1979) 1890–1892

Permanent magnetic properties of Fe–Cr–Co–Mo alloy

S. SZYMURA
Institute of Physics, Technical University, Częstochowa, Poland

L. SOJKA
Institute of Ferrous Metallurgy, Gliwice, Poland

Changes in microstructure, hardness and magnetic properties of the Fe–Cr–Co–Mo alloy were examined under the influence of isothermal tempering and with optimum heat-treatment. At a temperature of 1350 K the alloy shows considerable increase in hardness which is connected with the separation of the phase σ. The temperature of 1350 K should be recognized as the lower limit of hot-working. The alloy obtains optimum magnetic properties after multistage ageing between 870 and 810 K with previous thermomagnetic treatment at 910 K. The greatest changes in magnetic properties take place in the first two stages of ageing (870 K, 2 h; and 850 K, 2 h), whereas in further stages only $(BH)_{max}$ increases. During tempering, in the 870 to 810 K range, the microstructure does not change. A slight overstepping of temperature to 910 K during thermomagnetic treatment causes precipitation of the σ phase, growth occurring at multistage ageing.

1. Introduction

Kaneko *et al.* [1] have suggested a new hard magnetic alloy containing 28 to 32% Cr, 22 to 26% Co, the balance being Fe. Superior magnetic properties of this alloy ($H_c > 500\,A\,cm^{-1}$, $B_r \sim 1.2\,T$, $(BH)_{max} \sim 40\,kJ\,m^{-3}$), similar to those of Alnico 5, are obtained during heat-treatment in a magnetic field at 900 to 910 K and multistage ageing within range 870 to 810 K. Through addition of Nb, Al, Si, Mo and Ti into the Fe–Cr–Co alloys it is possible to facilitate the heat-treatment conditions to produce optimum permanent magnetic properties [1–3].

The alloy acquires its magnetic properties by spinodal decomposition of the b c c α phase, stable above 1570 K, in two phases, α_1 and α_2, one being a ferromagnetic phase, rich in Fe, and the other a weakly magnetic phase, rich in Cr (e.g. [4]).

The great technical importance of the Fe–Cr–Co alloy, in addition to its favourable magnetic properties, is its good workability. This alloy, as opposed to the brittle and hard Alnico alloy, can be worked plastically in hot and cold condition; enabling one to produce strips and wires and to punch small components.

The aim of this work was to study (i) the structural changes of the Fe–Cr–Co–Mo alloy during heat-treatment within the temperature range of 1570 to 810 K, (ii) magnetic properties, and (iii) variations in hardness associated with these changes.

2. Material and investigation procedure

Investigations were conducted on the alloy, the chemical composition of which is given in Table I. In order to obtain full information on the alloy structure, several investigation methods were used, including optical microscopy, X-ray structure analysis, measurement of hardness and magnetic properties (specific coercive force ($_JH_c$), residual flux density (B_r) and product of maximum magnetic energy $(BH)_{max}$).

TABLE I Chemical composition of the alloy (wt.%)

Fe	Cr	Co	Mo	N	O	C
42.96	28.69	25.36	2.95	0.009	0.007	0.021

0022–2461/79/081890–03 $02.30/0

The investigations covered two series of samples which were subjected, prior to testing, to homogenizing heat-treatment at 1570 K for 40 min and quenching in water. The samples of the first series, upon isothermal holding for various times, were quenched into water from the temperature range 1570 to 870 K. The second series of samples were isothermally heat-treated for 10 min within the temperature range 950 to 850 K, with or without a magnetic field of 400 kA m^{-1}, then some samples were subjected to multistage ageing for 1 to 2 h at 870, 850, 830 and 810 K.

3. Experimental results and discussion

Studies of the microstructure and X-ray structure analysis of the samples upon quenching from the temperature range 1570 to 870 K have shown that only quenching from a temperature of 1570 K enables one to obtain a single-phase alloy, and quenching from lower temperatures leads to the development of two-phase or triple-phase structures, causing considerable changes in hardness (Fig. 1). The greatest increase in hardness (from 27 to 47 HRC) is associated with the occurrence in the alloy microstructure, beginning at 1350 K, of considerable quantities of σ phase (see Fig. 1). Hence, the temperature of 1350 K is to be considered as the lower limit of the hot-working range of the Fe–Cr–Co–Mo alloy. From the micrographs shown in Fig. 1 it is clearly evident that the dispersion of phases is dependent upon the temperature at which the permanent transformation occurs.

The residual flux density (B_r) and the specific coercive force ($_JH_c$) versus the isothermal heat-treatment for 40 min at 830 to 950 K, with or without a magnetic field, are shown in Fig. 2. The curves showing the relationship between the specific coercive force and the residual flux density and tempering temperature of 910 K have a maximum value (see Fig. 2), the magnitude of which is higher for tempering in a magnetic field in comparison with tempering without a magnetic field (at $H_A = 0$ $_JH_c = 205$ A m^{-1} and at $H_A =$

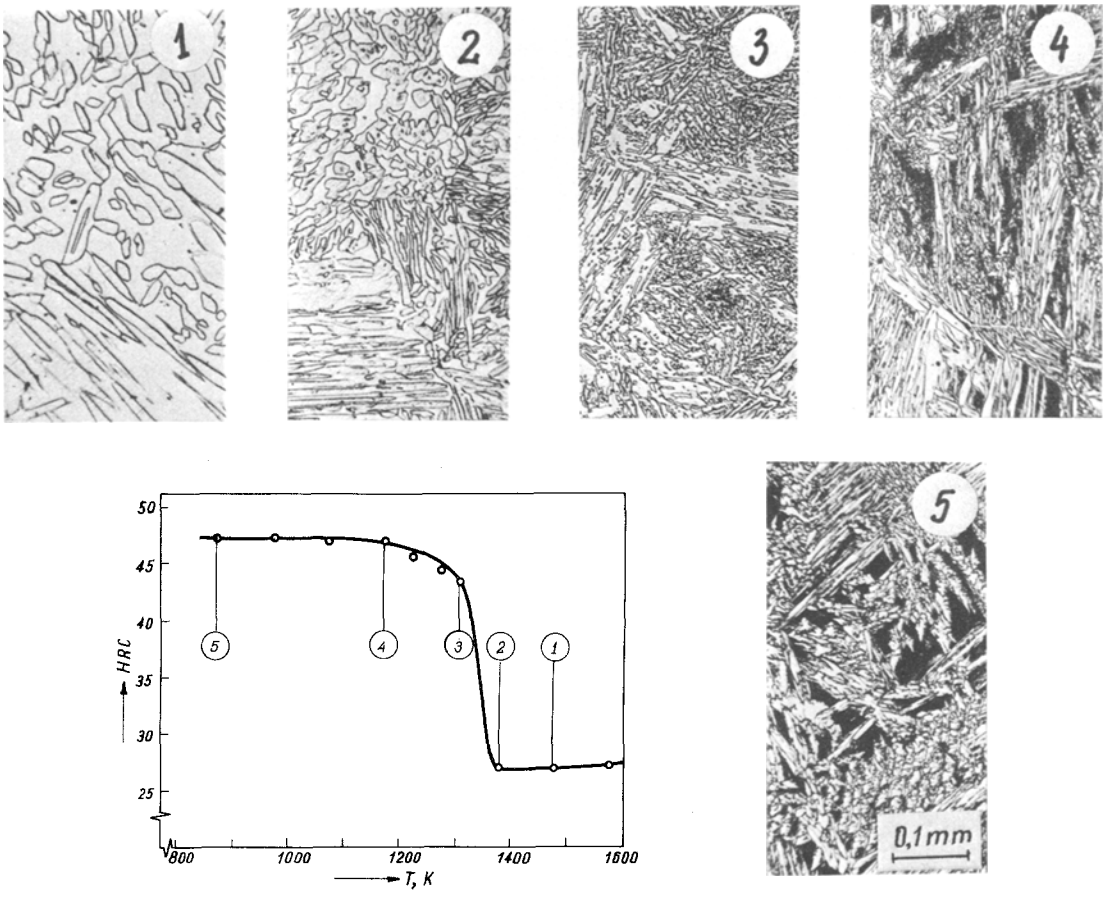

Figure 1 Microstructure and hardness of the Fe–Cr–Co–Mo alloy depending on hardening temperature.

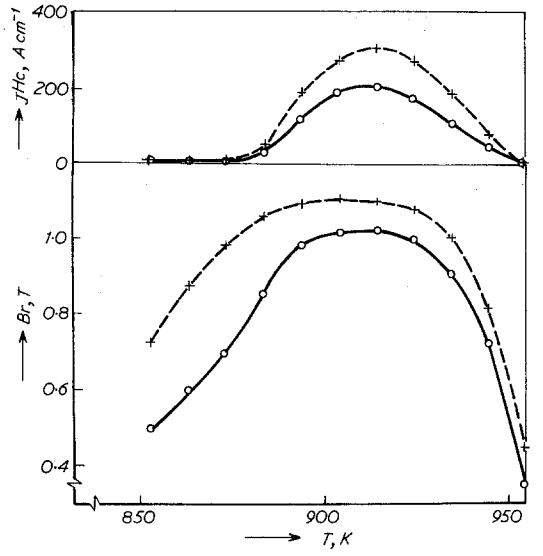

Figure 2 Magnetic properties versus temperature of isothermal heat-treatment: —— without a magnetic field, – – – in a magnetic field.

$400\,\mathrm{kA\,m^{-1}}$, ${}_JH_c = 310\,\mathrm{A\,cm^{-1}}$). The residual is practically constant within the temperature range 890 to 920K and rapidly decreases above and below this range (see Fig. 2). As might be expected, the residual flux density of samples which were tempered in a magnetic field is considerably higher than that of samples which were heat-treated without the effect of a magnetic field ($B_r = 1.03\,\mathrm{T}$ at $H_A = 0$, and $B_r = 1.1\,\mathrm{T}$ at $H_A \neq 0$). Variations of ${}_JH_c$ and B_r, as obtained by us, are similar to those given in [5].

The samples for which the best magnetic properties were obtained by tempering at 910K were subjected to multistage ageing within the temperature range 870 to 810K. The greatest changes in magnetic properties occurred in the first two stages of ageing (870K, 2h; 850K, 2h; see Fig. 3). Subsequent stages of ageing did not cause greater changes in ${}_JH_c$ and B_r, but they increased $(BH)_{max}$ (see Fig. 3).

Microstructure investigations have shown that multistage tempering does not influence its change. It should be noted that even a slight overstepping of temperature to 910K during thermomagnetic treatment causes a precipitation of the σ phase, which grows at multistage ageing.

The observed changes in magnetic properties resulting from the applied heat-treatments can be understood if one assumes a spinodal decomposition occurring during the tempering of the alloy. This decomposition has an anisotropic character

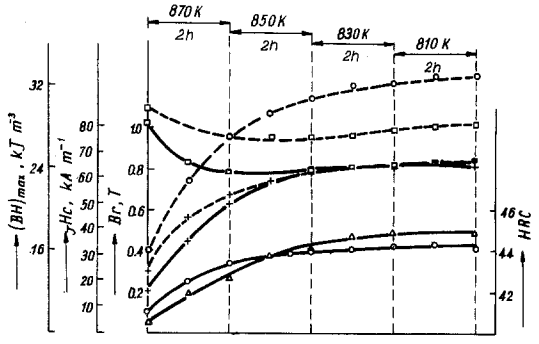

Figure 3 Effect of multistage annealing on B_r (□), H_c (+) and $(BH)_{max}$ (○) of the Fe–Cr–Co–Mg alloy: —— without a magnetic field, – – – in a magnetic field.

and proceeds along definite crystallographic directions and planes. In the Fe–Cr–Co–Mo alloy with regular structure, for which the elastic constants fulfil the condition: $2C_{44} - C_{11} + C_{12} > 0$, the spinodal decomposition of structure proceeds along the ⟨1 0 0⟩ direction, where the size of the precipitated phase particles depends on the tempering temperature [6]. In accordance with the results obtained here (see Fig. 2) it is to be concluded that the temperature of 910K and the size of the α_1 phase particles obtained at that temperature are optimum.

A high increase in specific coercive force and residual flux density which occur in the ageing temperature range 870 to 850K, are associated with the growing difference in saturation magnetization of the α_1 and α_2 phases, produced by changes in their chemical composition due to diffusion exchange of atoms – mainly Fe and Cr. Further stages of ageing have a minor effect on the interphase diffusion of elements, causing a small increase in B_r and a greater increase in $(BH)_{max}$, the value of ${}_JH_c$ being practically constant (see Fig. 3). The latter fact indicates that the size of the α_1 phase particles does not change during the ageing process.

References

1. K. KANEKO, M. HOMMA and K. NAKAMURA, AI Conf. no. 5 (1972) 1088.
2. K. KANEKO, M. HOMMA, K. NAKAMURA and M. MIURA, *IEEE Trans. Mag.* 8 (1972) 347.
3. K. KANEKO, M. HOMMA, T. FUKUDA and M. OKADA, *ibid* **11** (1975) 1440.
4. D. CHANDRA and L. H. SCHWARTZ, *Met. Trans.* **2** (1971) 511.
5. R. CREMER and I. PFEIFFER, *Physica* **80B** (1975) 164.
6. J. W. CHAN, *J. Appl. Phys.* **34** (1963) 3581.

Received 25 August and accepted 3 November 1978.