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Magnetic Properties and Microstructure of Co-Cr Bulk Alloys

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ABSTRACT

The effect of Cr addition on the microstructural changes and magnetic properties of bulk Co-Cr alloys has been investigated. Several interrelated microstructural and magnetic changes result from Cr additions. While Cr additions enhance the fcc phase retention and subsequently the fault density, they cause a decrease in grain size for a given thermal treatment. The additions also decrease saturation magnetization monotonically. Furthermore, the intrinsic coercivity is fairly insensitive to Cr additions for a given grain size and the coercivity increases with decreasing grain size.

1. Introduction

Since the introduction of perpendicular recording in 1977¹, systematic efforts have been directed towards providing a unique combination of suitable recording media and head designs

such that the potential of this technology is realized. Among magnetic materials with vertical anisotropy studied, it is generally ascertained that Co-Cr thin films have intrinsically superior properties and are the primary choice for ultra-high density thin film recording media². The goal is to record each individual bit of information on the individual vertically aligned column separated by voids or Cr segregation in the Co-Cr film. Each column has a diameter of about 500Å, amounting to a linear recording density of 500 KFRPI (flux reversal per inch).

While the microstructural details and magnetic properties of Co-Cr thin films are being extensively investigated³, little information is available for the bulk Co-Cr alloys. Since thin films are made through bulk alloy targets by various energizing methods such as sputtering or electron beam evaporation, it is worthwhile to consider that in Co-Cr phase diagrams,^{4,5} phase equilibria in the cobalt-rich region (<20 at. % Cr) below 700°C have not been firmly established, presumably due to slow diffusion. Also a series of intermediate phases related to the ϵ -Co phase have also been reported to exist⁵. Most recently, Bolzoni *et. al.*⁶, have attempted to determine the hcp $\longleftrightarrow$ fcc transformation temperatures on heating and on cooling. They observed a hysteresis, characteristic of martensitic transformations, which increases with Cr content. As these authors have asserted, this region of the phase diagram remains uncertain and further investigations need to be carried out. These authors have also studied the influence of Cr content on the Curie temperature, saturation magnetization, and anisotropy field⁶. These parameters were all found to decrease with Cr content and alloys were non-magnetic for Cr contents in excess of 27 at. % Cr. In view of these efforts, the objectives of this work are to correlate the microstructural evolution and magnetic properties, in particular the coercivity of these alloys, as a function of Cr content and of grain size.

2. Experimental procedures

2.1 Material fabrication and heat treatments

Cobalt chromium alloys of the following nominal compositions of 1.44, 5.0, 10.0, 15.2 and 20.0 atomic % Cr. were fabricated by vacuum melting and cast into cylinders that were 1 cm in diameter and 1 cm long. These as-cast samples were then subjected to various heat treatments.

2.1.1 Homogenization Treatments

As cast samples of the five different chromium compositions were encapsulated in quartz tubes backfilled with helium and homogenized for 70 hours at 1100° C. The samples were then water quenched. These samples will henceforth be referred to as T1 samples.

2.1.2 Grain Refining Treatment

Samples were rolled at 900° C in four passes from a thickness of 1 cm to 1 mm. Cracking, especially for the higher Cr samples required an intermediate reheat of 1 minute at 900° C between the first and second passes.

The hot rolled samples were subsequently hot rolled at 870° and 770° C, respectively for 15 minutes and then water quenched. These are referred to as T2 and T3 samples respectively.

2.1.3 Equilibrium Annealing Treatment

As cast samples of 10.0 and 15.0 at% Cr were homogenized at 1000° C for 50 hours, quenched and then aged at 620° C for 700 hours. The samples were encapsulated in quartz and back filled with helium prior to any heat treatment. This annealing treatment, designated as T4 was done in order to obtain equilibrium microstructures and to compare these to the microstructures observed in the as-quenched samples (T1). These two compositions were chosen since they appear to lie well within the single phase field in the Co-rich region of the Co-Cr phase diagram.

2.2 Microstructural Characterization

For optical metallography, specimens were mechanically ground following standard procedures. The ground surfaces were subsequently electropolished using a solution of 640 ml of ethyl alcohol, 140 ml of butoxyl alcohol and 220 ml of perchloric acid. Electropolishing conditions of 50 - 55 Volts were used. The samples were then etched in a solution of 60 ml HCl, 15 ml HNO₃, 15 ml CH₃COOH and 15 ml H₂O.

The etched specimens were examined with a Zeiss Axiomat optical microscope. Grain size measurements were performed using the intercept method and the calculated grain size for each sample was an average of ten intercepts.

The phase content of the bulk alloys was characterized by x-ray diffraction in a Siemens Diffractometer using Cu K α radiation on electropolished samples. Quantitative estimates of the amount of hcp phase in the alloys were obtained from equation (1) derived by Stage and Guillard⁷ for the calculation of weight % of hcp phase in pure cobalt.

$$\text{Wt. \% hcp} = \frac{I_{(10\ \bar{1}1)\text{hcp}}}{I_{(10\ \bar{1}1)\text{hcp}} + 1.5 I_{(200)\text{fcc}}} \times 100 \quad (1)$$

where I is the integrated intensity of the indicated peak. Equation (1) takes into consideration both the structural and multiplicity factors. The calculated weight percent of the hcp phase for the alloys is an estimate and not an accurate quantitative measurement because equation 1, which has been derived for a polycrystalline powder sample, is not adequately corrected for grain size effects.

For transmission electron microscopy (TEM), thin slices of approximately 0.4 mm in thickness were cut using a Buehler diamond saw and then mechanically punched out. The samples were then electropolished. Voltages used increased proportionally with the Cr content and ranged from 15 V for the low Cr compositions to 30 V for the highest Cr samples.

Conventional electron microscopy techniques of bright field and dark field imaging and diffraction were used to characterize the microstructure. The samples were examined in the Philips 301 and Philips 400 electron microscopes operated at 100 kV. Chemical analysis was done using the technique of energy dispersive x-ray spectroscopy (EDXS) on the Philips 400 microscope using a nominal probe size of 100Å.

2.3 Magnetic characterization

2.3.1 Magnetic properties

Magnetic hysteresis loops were measured using a Princeton Applied Research vibrating sample magnetometer (VSM). Samples for magnetic measurements were in the form of disks of 3 mm diameter, 75 - 100 μm in thickness. The hysteresis loops were measured in the plane of the sample with a saturating field of 12 KOe.

2.3.2 Domain observations

Both the Bitter powder method⁸ and transmission Lorentz microscopy were employed to study the magnetic domain structures. The former consists of placing a drop of colloidal suspension containing ferromagnetic particles of Co, 100 $\text{\AA}$ in diameter on the polished surface free from stress and covered with a thin cover glass to spread the suspension evenly. An external magnetic field of up to 1.5 kG was applied using a Sm-Co permanent magnet in order to observe domain wall motion.

The Philips EM400 electron microscope was used to image magnetic domains. The weak in-plane component of the objective lens field was used to move the domain walls. Upon tilting, the in-plane component of the field at the sample changes. Hence, observations of domain wall motion and its interaction with microstructural features were possible. Unfortunately, quantification of fields at the sample was not possible.

3. Results

3.1 Microstructure

Figure 1 (a-f) is the typical etched microstructure of the T1, T2, and T3 samples for both low (1.44 at.% Cr) (a,c,e) and high (20.0 at.% Cr) (b,d,f), respectively. The average grain size of the T1 sample was found to range between 500-600 μm , whereas, the average grain size of the T2 and T3 samples varies with Cr content. Figure 2 is the plot of the grain size as a function of Cr content for T2 samples. For both T2 and T3 heat treatments, the average grain size was observed to decrease with Cr content; however, the trend is more pronounced for T2 samples. As can be seen in Figure 1, the fault density increases with Cr content in the alloys. However, quantification of the relative amounts of fcc and hcp phases by optical metallography was not possible because of the fact that both phases can exist within a single grain and they are both faulted (as confirmed by TEM studies to be discussed later).

The phase analysis can be evaluated through x-ray diffraction studies. The (200) fcc peak was used to monitor the presence of the fcc phase and the relative intensities of the (200) fcc and $(10\bar{1}1)$ hcp peaks through equation (1) show a systematic increase in the amount of fcc phase retained (Figure 3) as a function of Cr content in the as quenched alloys. Although Figure 3 is only meant to provide an estimate of retained fcc phase which increase with Cr content, no fcc phase was found in the samples of 1.44 at. % Cr to the detectibility of x-ray diffraction.

Figure 4 is a bright field electron micrograph of a T1 sample of 15 at.% Cr composition. It is representative of the general microstructure for Cr contents of 5 at. % or greater. In alloys 5-20 at. % Cr, two phases are present. Within a single grain, the hcp phase appears as highly faulted bands embedded in the matrix fcc phase.

The faults in the hcp phase are observed to lie on (0001) basal planes, and those in the fcc phase are observed to lie on all four {111} planes, as expected.

In order to determine the nature of the stacking faults in the fcc phase, bright field/dark field (BF/DF) imaging of the faults was carried out. Figure 5 shows one such BF/DF image in a [001] zone axis. The bright field was imaged using the 020 reflection and the centred dark field was imaged using the $0\bar{2}0$ reflection. All four {111} variants of the faults lie inclined to the foil surface with an angle of 55° and they are responsible for the four satellite spots observed about each diffracted spots. The top and bottom of the fault are labelled T and B respectively. It can be seen that fringe contrast reversal occurs at the top of the foil in Figure 5. Using the method of Hashimoto *et.al.*⁹ the faults were therefore determined to be intrinsic, as expected because they are likely to be shear transformation, rather than growth faults.

Attempts were made to determine the nature (intrinsic or extrinsic) of the faults in the hcp phase by observing the contrast in the outer fringes when the faults are imaged with reflections (hkil) such that $(h - k) = -1$ ¹⁰. The difference in contrast observed when l is even and when l is odd determines the nature of the fault. However, the fault density was such that the fringes from the faults overlapped and analysis was not possible.

The observed orientation relation between the fcc and hcp phases for the alloys over the composition range of (5 at% - 20 at% Cr) is :

$$\begin{aligned} (0001)_{\text{hcp}} // (111)_{\text{fcc}} \\ [11\bar{2}0]_{\text{hcp}} // [1\bar{1}0]_{\text{fcc}} \end{aligned}$$

The relation is known as the Shoji-Nishiyama relationship (S-N relation)¹¹ and it is the

same as that observed for pure Co. It indicates that the close packed planes and directions in the two phases are parallel.

In samples of 1.44 at% Cr, no fcc phase was detected, so that the regions of hcp no longer appear as bands. However within a grain, several regions of hcp phase were observed to share a common $[1\ \bar{2}10]$ zone and the interface between the two regions was faceted. The facets of the interface coincided with the $[10\ \bar{1}0]$ directions of each hcp region.

In the annealed T4 samples, only the hcp phase was detected. Although the fault density was considerably decreased by the aging treatment, it was never observed to be zero.

3.2 Chemical analysis

In T1 samples the composition of the fcc and hcp phases were found to be uniform for compositions of 5 at% Cr and higher. This is in agreement with the nature of the martensitic transformation which maintains that parent and transformed phases are of the same composition.

In T4 samples of Co-15 at% Cr, no second phases were observed. Chromium segregation to the grain boundaries was investigated; however no segregation was detected. Table 1 shows a typical EDXS analysis from grain interiors and from grain boundaries. The spatial resolution of this technique is $\sim 200\text{\AA}$. (The Cr content given by the EDXS analysis is higher than that quoted as the Cr content of the bulk alloys). Fluorescence of Cr ($Z=24$) by Co ($Z=27$) may account for the higher Cr content found by the thin film analysis which does not take into account fluorescence. Moreover, fluorescence of Cr from the microscope column can also lead to higher Cr counts. However, EDXS was used to compare the relative Cr content from different regions

within a sample and not to establish the absolute Cr content.

The grain size of T2 and T3 samples which were deformed and recrystallized at 870°C and 770°C for 15 minutes, respectively decreased with Cr content. Since Cr segregation to the grain boundaries can influence the recrystallization and grain growth behavior, the role of Cr segregation was investigated in samples of 5 at% Cr and 20 at% Cr. Typical analyses of spectra were collected from grain interiors and grain boundaries (Table 2). It can be seen that the Cr content does not differ by more than 1% suggesting that the Cr distribution is uniform.

3.3 Magnetic Properties

The saturation magnetization of all the specimens decreases linearly with Cr content which is in agreement with previously reported data⁷. Figure 6 shows a plot of the intrinsic coercivity of T1 samples as a function of Cr content. The intrinsic coercivity was found to be low and nearly invariant (approx. 30 Oe.) with Cr content.

The intrinsic coercivity as a function of grain size was investigated for all the alloys T1, T2 and T3. For each Cr composition, three grain sizes over a range of 500 μm to 20 μm were studied. Figure 7 is a plot of the intrinsic coercivity as a function of log average grain size for compositions of 1.44 at% Cr and 20 at% Cr. In all compositions, a small increase in coercivity was observed with decreasing grain size; however, the absolute values of the coercivity remain fairly low (< 50 Oe). The increase in coercivity with decreasing grain size was most prominent for samples of 1.44 at% Cr. The increase in coercivity as a function of grain size was less pronounced with increasing Cr content.

3.4 Magnetic domains

Figure 8 is a Bitter pattern observed on the basal plane (i.e., c-axis normal to the surface) of T1 (1.44 at% Cr) under an external applied normal field of about 1kOe. The observed magnetic structure consists of a wavy pattern of oppositely vertically magnetized regions. Near the surface, spike domains have formed and they are arranged in a hexagonal pattern. The same patterns were observed in samples up to 15 at% Cr. Domain walls also were observed to move readily under the application of external fields. This indicates high virgin permeability.

Interpretation becomes considerably difficult for polycrystals where the surface of individual grains make random angles with the easy axis of magnetization, resulting in complex patterns. Also, faulting in quenched samples further complicates the interpretation of the pattern as shown in Figure 9. The ferromagnetic particles were observed to decorate the intersection of faults with the surface. This can be explained by interpreting the faults as microscopic grooves, in the otherwise flat electropolished surface. At the site of the grooves, free poles develop and a field gradient exists to attract colloidal particles. This observation in conjunction with the low resolution of the technique does not provide information concerning the interaction of the domain walls and the faults.

Figure 10 a,b are typical Lorentz electron micrographs from a T1 sample of 5 at.% Cr. Since the easy axis of magnetization for the hcp phase is the [0001] direction, the domain walls will intersect the (0001) faults in the hcp phase at 90° as shown in Figure 10a. The orientation of the foil in Figure 10a is $[2\ \bar{4}23]$ corresponding to the c-axis being tilted 35° from the plane of the sample. Domains of opposite magnetization are labelled on the figure. The stripe domains with

Bloch-type walls observed in these alloys are similar to those observed for pure Co¹². These 180° Bloch walls where the magnetization vector rotates about the axis normal to the wall were observed in the thick regions (>700Å) of the samples. Near the edge of the sample, cross tie walls consisting of alternative Neel segments separated by Bloch lines were observed as shown in Figure 10b.

3.5 Interaction of domain walls with microstructural features

Figure 11a,b,c shows a series of interaction of domain walls with structural features from a region populated with thin bands of fcc and hcp phases in a T3 sample of 5 at. % Cr. The domain wall is continuous across the interface of the two phases. The domain walls in the hcp phase always intersect the faults at 90° since the crystallography of the faults dictates that the faults lie on (0001) planes and the magnetization lies along the easy axis of magnetization, the [0001] direction. Hence, motion of domain walls in the hcp phase was observed to be unrestricted whereas the fcc phase which has four equivalent easy directions of magnetization; i.e., <111> has lower anisotropy, and domain walls in the fcc region are more flexible and need not necessarily run normal to the faults. Pinning of domain wall by the fcc faults is not strong even when the faults lie parallel to the domain wall. Figure 11b shows the domain wall at position 2 after tilting of the sample. Below the arrows, the wall has moved smoothly in the hcp phase, unrestricted by the faults which run normal to the wall. The wall, however, drags behind in the fcc regions. The fcc regions are easily recognized as those regions which show several variants of the {111} fault planes. Domain wall interaction with the faults in the fcc region was observed to vary depending on the angle between the faults and the domain wall. When the domain wall intersects the fault at 90°, no interaction was observed and when the domain wall is parallel to the faults, weak pinning was observed.

This can be explained as follows. Assume that the wall energy is a function of the wall position and that the energy of the wall is lowered near the stacking fault. Displacement of the wall that is normal to the stacking fault will not change the energy of the wall as compared to the original position. Hence, the wall moves easily into the new position to minimize the overall magnetic energy corresponding to that particular value of the in-plane component of the magnetic field of the objective lens. However, when the wall is parallel to the stacking fault, as it can be in the case of the fcc phase, then displacement of the wall from its "pinned" position at the fault, into the unfaulted region can lead to an increase in the wall energy. This means that the wall will have a tendency to "stick" to the fault interface, as is qualitatively observed in Figure 11. The angular dependence of the interaction between the faults and the domain wall has also been observed in ferrites¹³

Figure 12a shows a portion of a domain wall pinned at the fault at position 1. With a succession of two abrupt jumps, the wall has moved to position 2 (Figure 12b) and upon reversal of the field, the wall jumps back to position 1. This reversible wall movement inside the grains indicates, qualitatively, that the strength of pinning by the faults in the fcc phase is not very strong.

Figure 12a also shows a domain wall intersecting a grain boundary at position 3 between grain 1 (G1) and grain 2 (G2). The wall intersects the grain boundary at near 90° . As for the case of stacking faults, grain boundary interaction with domain walls when the angle is close to 90° is very weak. This is shown in figure 12b where the domain wall has moved across the grain to position 4. The interaction of the grain boundary with domain walls is stronger when the domain wall and the grain boundary are parallel. Figure 13 shows a domain wall that has moved through grain 1 (G1) towards grain 2 (G2). Domain wall motion was observed to be effectively

impeded at the grain boundary between G1 and G2. (The arrows indicate the region of the wall which runs parallel to the grain boundary.)

4. Discussion

In terms of microstructural features, the addition of Cr was found to increase the fcc phase content and the density of fcc stacking faults in the range of compositions studied.

From the literature, it has been reported that the addition of C, W and Fe increases monotonically the fcc phase content of Co-C, Co-W and Co-Fe alloys^{14,15}. Furthermore, C was found to be more effective in stabilizing the fcc phase and the rationale for this is on the basis of expansions in the unit cell per atom additions. The expansion was much greater for interstitial C atoms than it was for substitutional W. Such an expansion in the unit cell makes the fcc-to-hcp transformation energetically unfavorable and leads to the retention of the fcc phase. However, considering the atomic radii of Co and Cr (1.2530 Å and 1.2490 Å, respectively) additions of substitutional Cr to Co are unlikely to cause large changes in the unit cell dimensions. Hence changes in phase content cannot be attributed to changes in unit cell volume resulting from Cr additions. Within the resolution limits of x-ray diffraction the results proved to be inconclusive due to the low intensities at the high angles required for accurate lattice parameter determinations and also due to the overlap of fcc and hcp peaks.

The fcc-to -hcp phase transformation is martensitic in nature. The role of Cr then would appear to affect fcc phase retention through its influence on the M_s temperature. By lowering the M_s temperature the driving force for transformation is reduced, causing the retention of the high temperature fcc phase. Indeed, it has been found that beyond 5 at.% Cr, the transformation

temperature of the fcc $\longleftrightarrow$ hcp transformation decreases with Cr content¹⁶. In the range of 0-5 at.%Cr, there exists an inconsistency with the present work which may have to be investigated in greater detail.

The role of Cr addition in refining the grain size lies in the competing contributions of the rate of nucleation, N vs. the rate of growth, G during the recrystallization process. When the N/G ratio is high, many nuclei will form prior to completion of the recrystallization process and a fine grain size will result. Another possible effect is the retardation of grain growth rather than the enhancement of nucleation rate that contributes to grain size refinement. Retardation could result from solute segregation and subsequent impediment to grain boundary motion; however, EDX spectroscopy did not reveal any solute segregation at the grain boundaries within the spatial resolution of the technique (200 Å). Therefore, given the same history for all the compositions and the lack of solute segregation, it would appear that Cr increases the nucleation rate, leading to a finer grain size.

As grain boundaries serve to stabilize the parent phase and thus hinder the martensitic transformation, the observed decrease in grain size would lead to increased fcc phase retention. This is consistent with the x-ray diffraction and optical microscopy results discussed above. Thus, it would appear that Cr additions affect the M_s temperature both directly and indirectly through their influence on the grain size. The contribution of the latter effect is likely to be less significant than the former given the relatively large grain sizes (30-70 μ m) used in this study.

The saturation magnetization of the alloys was observed to decrease almost linearly with Cr content. This is consistent with the results reported in the literature [Bolzoni et.al., 1983]. Bolzoni and co-workers explain the decrease in saturation magnetization as caused by Cr atoms

which carry a moment of 4.25 Bohr magnetons coupling antiferromagnetically with Co moments. Neutron diffraction data ¹⁷ supports the contention that the decrease in saturation magnetization is due to antiferromagnetic coupling of Co and Cr moments rather than by a dilution of Co moments by Cr.

The intrinsic coercivity for all the compositions studied was found to be low (~30 Oe.) and almost invariant with composition for a given heat treatment. The intrinsic coercivity, however, for a given composition showed a small increase with decreasing grain size. This effect decreased with increasing Cr content.

The observed dependence of the intrinsic coercivity on the grain size is in agreement with the results obtained by other researchers for SmCo_5 ¹⁸ and for ferrites¹⁹. Both Lorentz microscopy and the initial magnetization curve which shows a concave upward shape have provided direct and indirect evidence that the magnetization reversal proceeds predominately by domain wall displacement. The increase in intrinsic coercivity with decrease in grain size has two possible contributions, one affecting the nucleation of reverse domains and the other affecting the growth of reverse domains.

Nucleation of reverse domains are most likely at surface defects or at sites where the local anisotropy is lowered. It has been shown that the number of defects per unit surface area decreases rapidly with decreasing grain size²⁰. Thus, a smaller grain size implying a smaller surface area for a given grain would increase the coercivity by decreasing the nucleation probability. Growth of reverse domains can be effectively impeded by grain boundaries since grain boundaries constitute regions of local decrease in exchange energy and anisotropy and hence, lowered domain wall energy. Thus a smaller grain size would provide more grain boundary area

to pin domain walls and impede the growth of reverse domains. This would also contribute to an increase in the coercivity.

5. Conclusion

The major findings of this study of the microstructure and magnetic properties of bulk Co-Cr alloys are the following:

1. The addition of Cr to Co was found to stabilize the fcc phase. This is attributed to the lowering of the martensitic start temperature, M_s .
2. Chromium additions tend to decrease the average grain size of these binary alloys probably by enhancing the nucleation rate during recrystallization. No chromium segregation is detected in bulk Co-Cr alloys within the detectibility of EDXS.
3. The intrinsic coercivity for a given composition is shown to increase with decreasing grain size. This is due to the decrease of nucleation sites for reverse domains and to the increase in grain boundary area which act as pinning sites for domain wall motion.
4. Lorentz microscopy revealed that the stacking faults in the fcc phase and grain boundaries acted as pinning sites. The latter were observed to be more effective pinning sites than the former. The coercivity mechanism is predominately wall-displacement controlled.

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Table I

EDXS analysis from sample T4 of 15 at. % Cr

Grain boundary	at. % Cr	at. % Co
	20.11 ± 0.29	79.89 ± 0.65
	19.95 ± 0.25	80.05 ± 0.49
	19.82 ± 0.21	80.18 ± 0.49
	19.59 ± 0.21	80.41 ± 0.49
	19.88 ± 0.25	80.32 ± 0.49
	19.56 ± 0.28	80.44 ± 0.49
Grain interior	19.88 ± 0.25	80.32 ± 0.49
	19.51 ± 0.21	80.49 ± 0.49
	19.54 ± 0.21	80.46 ± 0.49

Table II

EDXS analysis of sample T2

Sample T2: 20 at. % Cr

Grain boundary	at. % Cr	at. % Co
	24.45 ± 0.40	75.55 ± 0.78
	24.48 ± 0.44	75.52 ± 0.78
	24.77 ± 0.45	75.23 ± 0.77
	24.78 ± 0.50	75.22 ± 0.93
	24.90 ± 0.50	75.10 ± 0.93
Grain interior	at. % Cr	at. % Co
	24.31 ± 0.35	75.69 ± 0.62
	24.94 ± 0.32	75.06 ± 0.62
	24.98 ± 0.45	75.02 ± 0.77
	24.42 ± 0.35	75.58 ± 0.62
	25.16 ± 0.55	74.84 ± 1.08
	24.20 ± 0.40	75.80 ± 0.80

Sample T2: 5 at. % Cr

Grain Boundary	at. % Cr	at. % Co
	7.55 ± 0.30	92.45 ± 0.93
	7.44 ± 0.26	92.56 ± 0.93
	7.38 ± 0.29	92.62 ± 0.93
	7.46 ± 0.31	92.54 ± 0.93
	7.08 ± 0.21	92.92 ± 0.94
Grain Interior	at. % Cr	at. % Co
	7.33 ± 0.25	92.68 ± 0.75
	7.28 ± 0.26	92.72 ± 0.75
	7.41 ± 0.22	92.59 ± 0.75

Figure captions

- Figure 1. Etched microstructure of the T1, T2, and T3 samples of 1.44 at.% 5 Cr (a,c,e) and of 20 at.% Cr (b,d,f).
- Figure 2. Plot of grain size as a function of Cr content for T2 samples.
- Figure 3. Plot of the ratio of fcc phase to hcp phase content.
- Figure 4. Bright field (BF) electron micrograph of a T1 sample of 15 at. % Cr composition.
- Figure 5. Bright field/dark field (BF/DF) image of the faults in the fcc phase. The BF and DF micrographs were imaged using the 020 and $0\bar{2}0$ reflections, respectively.
- Figure 6. Plot of intrinsic coercivity of T1 samples as a function of Cr content.
- Figure 7. Plot of intrinsic coercivity as a function of log (average grain size).
- Figure 8. Bitter pattern observed on the basal plane of a T1 sample of 1.44 at. % Cr.
- Figure 9. Bitter pattern observed on a highly faulted T1 sample.
- Figure 10. Lorentz electron micrographs from a T1 sample of 5 at.% Cr showing (a) stripe domains with Bloch-type domain walls and (b) cross tie walls.
- Figure 11. Series of Lorentz electron micrographs showing the interaction between domain walls with faults .
- Figure 12. Series of Lorentz electron micrographs showing the interaction between domain walls and microstructural features (see text).
- Figure 13. Lorentz electron micrograph showing the interaction between a domain wall and a grain boundary.

Figure 1

XBB870-10270

XBL 876-2861

Figure 2

XBL 878-3686

Figure 3

XBB876-5087

Figure 4

XBB876-5093

Figure 5

XBL 876-3000

Figure 6

XBL 878-3688

Figure 7

XBB873-2151

Figure 8

Figure 9

XBB860-8649

XBB876-5094

Figure 10a

Figure 10b

XBB877-5643

XBB877-5644

Figure 11

XBB877-5645

Figure 12

XBB877-6396

Figure 13

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