Structure and Magnetic Properties of Fe nanoparticles embedded in a Cr matrix

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Abstract

The structure of 2 nm diameter (340 atoms) Fe nanoparticles embedded in a Cr matrix was determined using X-ray Absorption Fine Structure (EXAFS) and the magnetic properties studied by Superconducting Quantum Interference Device (SQUID) magnetometry. The thin films were produced by the co-deposition of pre-formed gas-phase Fe clusters synthesised by a gas aggregation source with an atomic vapour of Cr produced by an MBE source. The behaviour was studied as a function of Fe nanoparticle volume fraction in the range 5 – 20% and was compared to previous results on ferromagnetic nanoparticles in antiferromagnetic matrices. EXAFS showed that the atomic structure in the Cr-embedded Fe nanoparticles is the same as the bulk bcc structure. Whereas alloying between the nanoparticles and matrix material has previously been shown to be very pronounced for Co nanoparticles in antiferromagnetic Mn, it was found that any alloying between Fe nanoparticles and Cr matrix material is limited. For dilute samples of Fe nanoparticles in Cr the measured saturation magnetisation (M_S) was $1\mu_B/\text{Fe atom}$, which is significantly less than the bulk M_S value of $2.22\mu_B/\text{Fe atom}$ indicating that the surface of Fe nanoparticles is either antiferromagnetic or non-magnetic. An increase in the volume fraction produces an increase in the value of M_S and at a volume fraction of 20%, M_S exceeds the value of bulk Fe showing that some Cr spins provide a ferromagnetic contribution. After field cooling below 30K, all films show Exchange Bias (EB) and an increase of coercivity, which are both much larger for the most concentrated sample. The Cr spins at the surface of the Fe particles play a key role in determining the overall magnetic behaviour.

Keywords: nanoparticles, exchange bias, FM/AFM interfaces, non-magnetic.

1. Introduction

Nanosized magnetic particles have attracted intense attention due to their novel properties, which can be controlled, and their utility in a range of applications from biomedicines [1] to magnetic storage devices [2]. Fundamentally in such small particles the novel behaviour arises from the high proportion of under co-ordinated surface atoms and quantum size effects. Embedding the nanoparticles in different materials of choice (host matrices) increases the diversity of applications further [3]. Antiferromagnetic (AF) matrices are a particularly interesting choice for embedding ferromagnetic (F) nanoparticles as normally the exchange interaction at the F/AF interface produces a shift in the hysteresis loop in field cooled samples along the applied field direction, known as Exchange Bias (EB). In the presence of EB there is usually a significant enhancement in the coercivity [4]. The general explanation of EB is given in terms magnetic moments that are pinned in a single direction at the interface by the AF. So rotating the F material by an external field produces a shift in the magnetic loop given by the difference in exchange between parallel and antiparallel spin alignments at the interface (for reviews see [5,6]). In a typical system however, only a small fraction of the spins are pinned [7] and this proportion and thus the value of the EB depend strongly on the microstructure at the interface. It is therefore of great interest to study the phenomenon in F/AF interfaces at the surfaces of nanoparticles of controlled size.

The exchange bias phenomenon was observed about fifty years ago in partially oxidized Co particles [8]. Although there has been some research in the field of exchange bias in nanoparticles over the last few decades, the bulk of the research has focused on thin film systems due to their applicability in spin valves in magnetic recording and sensor devices [9]. This was due to (a) the possibility of an increased number of F/AF interfaces by utilising multilayers; (b) the greater control of the interface that thin films allow, (e.g. grain size, orientation, crystalline quality, roughness, spin structure or interface layers) and finally; (c) the fundamental role of EB in spin valve and tunnelling devices. Recently the study of EB effects in fine particle system is recovering its importance, since it is known that the F-AF exchange interaction can be used to enhance the hard magnetic properties of permanent magnets by increasing the coercivity [10-12] or to overcome the superparamagnetic limit in F nanoparticles [13]. In addition controlling the anisotropy will enable the optimisation of other performance parameters, for example the heat they produce in response to an alternating magnetic field, which could lead to the synthesis of improved nanoparticles for hyperthermia treatment of tumours [14]. It is well known that a range of novel properties arise in both the F and AF materials as the size is reduced (e.g. increased importance of surface effects, changes in the magnetization reversal modes or superparamagnetism [3]), so it is important to investigate how size reduction affects F-AF coupled systems.

In this paper we present structural measurements, performed using EXAFS, of 2 nm diameter Fe (F) nanoparticles dispersed at low volume filling fractions (VFF) in Cr (AF) matrices, along with two conventional FeCr alloy films produced by Molecular Beam Epitaxy (MBE) for the purposes of comparison. We also present magnetisation measurements over a range of temperatures for 5%, 10% and 20% VFF of Fe nanoparticles in Cr matrices.

2. Details of Sample Preparation and Magnetic Measurements

Fe nanoparticles were formed in the gas phase using a UHV-compatible gas aggregation source described in detail in [15] and shown schematically in Fig. 1(a), which illustrates the method of preparing embedded nanoparticle samples. The source produces Fe particles with a log-normal distribution of diameters with a most probable diameter of ≈ 2 nm. A quadrupole filter allows in situ monitoring of particle sizes and the mass spectrum of nanoparticles for the samples produced in this work is shown in Fig. 1(b). The data (circles) has been fitted to a log-normal curve (line) and indicates a median diameter of 1.97 ± 0.05 nm (340 atoms).

Fig. 1(a) Schematic of gas aggregation source used to produced Fe nanoparticles and the method for depositing embedded nanoparticle films. **(b)** Mass sepctrum of gas phase Fe nanoparticles giving a median diameter of 1.97 ± 0.05 nm.

Two basic types of sample were produced that were suitable either for magnetometry or structural measurements using EXAFS. For the magnetometry measurements the films were prepared by co-depositing the nanoparticles and Cr vapour produced by an MBE source onto poly-ether-ether-ketone (PEEK) substrates of area $11 \text{ mm} \times 20 \text{ mm}$ and covered by a $200 \text{ \AA}$ thick buffer layer of Ag. The total equivalent thickness of the Fe part was about $300 \text{ \AA}$ in each sample. A capping layer of Ag with a thickness of $200 \text{ \AA}$ was deposited onto each sample to protect it against oxidation after removal from the ultra-high vacuum deposition chamber. Deposition rates at the substrates for Fe nanoparticles and the Cr matrix were measured by a quartz crystal thickness monitor to control the volume filling fraction (VFF) of the clusters in the thin films. Three magnetometry samples with approximately 5% , 10% and 20% VFF concentration were prepared by choosing specific deposition rates for the Fe and Cr.

For EXAFS measurements the embedded nanoparticle samples were grown by co-depositing Fe nanoparticles and the Cr matrix (as described above) onto Si substrates. In addition pure Fe and Fe-Cr alloy films were produced using MBE as a check for known bulk interatomic distances.

The atomic structure of the Fe nanoparticles embedded in Cr matrices and FeCr MBE films was investigated by means of Fe K edge EXAFS experiments. The measurements were done on beamline BM29[16] at the ESRF. A double crystal Si(111) monochromator was used and the intensity of the incident x-rays was calibrated using an ionization chamber containing He. The Fe edge absorption spectra were measured in fluorescence using a nine-element monolithic Ge detector. These were background-subtracted and normalised using the program PySpline[17] to provide the EXAFS spectra $\chi(k)$, where k is the photoelectron wavevector. Analysis of these, by fitting the experimental $\chi(k)$ to calculated EXAFS functions, then yielded structural parameters – interatomic distances r_i , mean square variations in interatomic distance σ_i^2 (Debye-Waller factors) and coordinations N_i . An additional parameter E_F , which is a correction to the Fe absorption edge energy, was allowed to freely vary in the fitting procedure. The analysis was performed with the aid of the EXCURV98 program [18], which uses fast curved wave theory [19] to calculate $\chi(k)$. Atomic scattering potentials and phase shifts are calculated within the program using Hedin-Lundqvist potentials, which account for the amplitude reduction effects in EXAFS [20]. All errors are quoted to ± 2 standard deviations.

Magnetic measurements were carried out using a commercial SQUID magnetometer ($H_{\max} = 55$ KOe; $1.8 \text{ K} < T < 400 \text{ K}$).

3. Results and Discussion

3.1. EXAFS measurements for FeCr MBE Films

Fe K edge EXAFS spectra (k^3 -weighted) and their Fourier transforms for the two FeCr alloy reference films produced by MBE are presented in Fig. 2, along with the corresponding data for an Fe MBE film. Analysis of the Fe MBE data shows a fit entirely consistent with the bcc structure as expected for bulk Fe. During the fitting procedure, N_i were held fixed at values consistent with the bcc structure while r_i and σ_i^2 were allowed to vary freely. It was possible to fit five significant shells including direct multiple scattering effects between the first and fifth shell, as expected for a bcc structure. Table 1 gives the fit parameters obtained from the data along with the coordinations and interatomic distances in bulk bcc Fe.

Fig. 2 Fe K edge EXAFS spectra χ , weighted by k^3 and associated Fourier transform for FeCr MBE films with two different compositions and an Fe MBE film. The full lines represent the data and Fourier transform of the data and the dashed lines represent the fit to the data and the Fourier transform of the fit.

As can be seen in Fig.2 the Fe edge EXAFS for the FeCr alloy films looks qualitatively very similar to the data for the Fe film, strongly suggesting a b.c.c. structure in these samples. Hence a similar fitting procedure was adopted i.e. N_i were held fixed at values consistent with the bcc structure. The data analysis shows that the structure in both alloy films is indeed bcc; the fit parameters obtained for the FeCr MBE films are shown in Table 1. In the Fe dilute FeCr MBE sample (~ 0.09 at Fe %) a slight stretch is apparent in the interatomic distances relative to those measured in the Fe MBE film. This is reasonable as Fe can be considered isolated at such a low concentration and the bcc Cr structure is stretched relative to bcc Fe – the lattice constants for bulk Fe and Cr are $2.87\text{\AA}$ & $2.91\text{\AA}$ respectively. Note that using Fe scatterers significantly worsens the fit for this sample. The

interatomic distances in the $\text{Fe}_{0.49}\text{Cr}_{0.51}$ alloy film, are much nearer to those of Fe, although perhaps slightly stretched. Using Cr scatterers in the fit increases r_i by $\sim 0.02\text{\AA}$. Doing this improves the fit slightly, although not significantly. Assuming that the bonding in the b.c.c. FeCr MBE alloy films is randomly ordered, Fe atoms in a $\text{Fe}_{0.5}\text{Cr}_{0.5}$ alloy should be coordinated equally by Fe and Cr atoms.

3.2. EXAFS measurements for Fe clusters in a Cr matrix

The Fe K edge EXAFS spectrum (k^3 -weighted) and associated Fourier transform for a sample that contains 5.3% VFF Fe nanoparticles embedded in a Cr matrix is shown in Fig. 3.

Initial observation of the Fe K edge EXAFS for the Cr-embedded Fe nanoparticles reveals that the data looks very similar to that measured for the Fe MBE sample. This is reinforced by the data analysis, which yields a five-shell fit consistent with the bcc structure. As for the MBE films discussed in Section 3.1, N_i were held fixed at b.c.c. values while r_i and σ_i^2 were allowed to vary freely. The fit values obtained are given in Table 1, which show that the values for r_i are consistent with the bcc structure in Cr-embedded Fe nanoparticles.

Table 1. Structural parameters r_i and $2\sigma_i^2$ (interatomic distances, Debye-Waller factors respectively) obtained from the fits to Fe K edge EXAFS for Fe nanoparticles in Cr matrix, FeCr MBE films and an Fe MBE film. Also included are the coordination numbers and interatomic distances in the bulk bcc Fe.

Sample	Shell 1	Shell 2	Shell 3	Shell 4	Shell 5
bcc Fe	$r_1 = 2.49 \text{ \AA}$	$r_2 = 2.87 \text{ \AA}$	$r_3 = 4.06 \text{ \AA}$	$r_4 = 4.76 \text{ \AA}$	$r_5 = 4.97 \text{ \AA}$
	$N_1 = 8$	$N_2 = 6$	$N_3 = 12$	$N_4 = 24$	$N_5 = 8$
Fe MBE film	$r_1 = 2.47 \pm 0.01 \text{ \AA}$	$r_2 = 2.85 \pm 0.01 \text{ \AA}$	$r_3 = 4.06 \pm 0.02 \text{ \AA}$	$r_4 = 4.74 \pm 0.01 \text{ \AA}$	$r_5 = 4.98 \pm 0.02 \text{ \AA}$
	$2\sigma_1^2 = 0.009 \pm 0.001 \text{ \AA}^2$	$2\sigma_2^2 = 0.013 \pm 0.002 \text{ \AA}^2$	$2\sigma_3^2 = 0.016 \pm 0.003 \text{ \AA}^2$	$2\sigma_4^2 = 0.015 \pm 0.002 \text{ \AA}^2$	$2\sigma_5^2 = 0.015 \pm 0.002 \text{ \AA}^2$
$E_F = 3.3 \pm 0.6 \text{ eV}$					
5.3% VFF Fe Nanoparticles in Cr Matrix	$r_1 = 2.49 \pm 0.01 \text{ \AA}$	$r_2 = 2.85 \pm 0.01 \text{ \AA}$	$r_3 = 4.08 \pm 0.02 \text{ \AA}$	$r_4 = 4.78 \pm 0.02 \text{ \AA}$	$r_5 = 5.00 \pm 0.02 \text{ \AA}$
	$2\sigma_1^2 = 0.014 \pm 0.001 \text{ \AA}^2$	$2\sigma_2^2 = 0.019 \pm 0.002 \text{ \AA}^2$	$2\sigma_3^2 = 0.024 \pm 0.004 \text{ \AA}^2$	$2\sigma_4^2 = 0.019 \pm 0.002 \text{ \AA}^2$	$2\sigma_5^2 = 0.025 \pm 0.004 \text{ \AA}^2$
$E_F = 2.1 \pm 0.5 \text{ eV}$					
Fe _{0.09} Cr _{0.91} (Cr scatters in the fit)	$r_1 = 2.50 \pm 0.01 \text{ \AA}$	$r_2 = 2.86 \pm 0.01 \text{ \AA}$	$r_3 = 4.09 \pm 0.02 \text{ \AA}$	$r_4 = 4.76 \pm 0.01 \text{ \AA}$	$r_5 = 5.02 \pm 0.02 \text{ \AA}$
	$2\sigma_1^2 = 0.008 \pm 0.001 \text{ \AA}^2$	$2\sigma_2^2 = 0.009 \pm 0.002 \text{ \AA}^2$	$2\sigma_3^2 = 0.015 \pm 0.003 \text{ \AA}^2$	$2\sigma_4^2 = 0.012 \pm 0.002 \text{ \AA}^2$	$2\sigma_5^2 = 0.012 \pm 0.002 \text{ \AA}^2$
$E_F = 3.5 \pm 0.6 \text{ eV}$					
Fe _{0.49} Cr _{0.51} (Fe scatters in the fit)	$r_1 = 2.47 \pm 0.01 \text{ \AA}$	$r_2 = 2.85 \pm 0.01 \text{ \AA}$	$r_3 = 4.07 \pm 0.02 \text{ \AA}$	$r_4 = 4.74 \pm 0.01 \text{ \AA}$	$r_5 = 5.00 \pm 0.02 \text{ \AA}$
	$2\sigma_1^2 = 0.010 \pm 0.001 \text{ \AA}^2$	$2\sigma_2^2 = 0.012 \pm 0.002 \text{ \AA}^2$	$2\sigma_3^2 = 0.017 \pm 0.003 \text{ \AA}^2$	$2\sigma_4^2 = 0.013 \pm 0.002 \text{ \AA}^2$	$2\sigma_5^2 = 0.013 \pm 0.002 \text{ \AA}^2$
$E_F = 3.3 \pm 0.6 \text{ eV}$					

It can be seen from the fit parameters in the above table that there is a slight stretch in the interatomic distances for the embedded clusters compared to the Fe MBE film. This small change can be due to the fact that Cr has a slightly stretched bcc structure in comparison to Fe. However the possibility of alloying at the Fe/Cr interface also needs to be considered. The formation of the σ -FeCr structure, which has a complicated tetragonal structure (30 atoms in the unit cell) can be ruled out as this would produce low amplitude EXAFS oscillations due to many overlapping/closely spaced shells. The formation of α -FeCr (bcc alloy) remains a possibility. However we note that switching from Fe scatterers to Cr scatterers produces a significant worsening of the fit. This can be compared to the Fe_{0.09}Cr_{0.91} alloy film (where switching from Cr to Fe scatterers significantly worsened the fit) and the Fe_{0.49}Cr_{0.51} alloy film (where switching between Cr and Fe scatterers did not significantly improve the fit quality). This indicates that although it can't be excluded any alloying at the Fe/Cr interface is limited.

3.3. Measurement of magnetic moments of Fe clusters in Cr matrix

In Fig. 4(a) the field dependence of magnetisation after ZFC at 5 K for Fe in Cr at three different volume fractions are shown and the magnetic moments per atom, obtained from the saturation magnetisation of the three samples are summarised in figure 4b.

The equivalent thickness of Fe, initially determined by a quartz crystal thickness monitor, was calibrated by measuring the magnetic signal from a film of pure Fe nanoparticles with no matrix and assuming the bulk Fe magnetic moment of $2.2\mu_B$ /atom. In principle, the small stretch observed by EXAFS in the structure of the embedded Fe clusters could lead to an enhanced atomic moment in the Fe as theoretically predicted [21]. However as shown in Fig. 4(b), a significantly lower atomic moment than bulk Fe is observed in the dilute sample and the value increases as the VFF increases. Recently a similar value of the Fe atomic magnetic moment ($\sim 1\mu_B$ /atom) was observed by XMCD measurements for Fe@Cr core-shell nanoparticles with a 2.7 nm diameter Fe core and a shell thickness of one or two monolayers of Cr embedded in Ag matrices [22]. In that case, the lowering of the Fe atomic moment was ascribed either to a disordered or an antiferromagnetic spin configuration at the interface, depending on the number of Cr monolayers.

A low value of the Fe atomic moment in embedded Fe nanoparticles was also reported by Baker et al [23] in 5.4% VFF (fcc) Fe clusters in Cu, although in that case they showed that the reduction was due to a change in crystal structure from bcc to fcc. Co clusters embedded in a Mn matrix, prepared using the same source as in the present study, also show a considerable decrease in their atomic moment per atom as compared to the bulk Co value [22]; this was ascribed to significant alloying between Co and Mn, resulting in an AF CoMn alloy shell surrounding a much reduced Co core. In the case of Fe clusters in a Cr matrix, the EXAFS measurements discussed above show that a decrease in the Fe moment at low volume fractions does not correlate with a change in crystal structure from the bcc phase; they also indicate that the degree of any alloying between Fe and Cr is slight (much less than in the F/AF Co/Mn system). In any case, b.c.c. FeCr alloys are ferromagnetic for Cr compositions between 0 and > 70 at.% [25,26]. From diffuse neutron scattering experiments on b.c.c. FeCr alloys [26], values for the atomic Fe and Cr moments of $2.05 \mu_B$ and $-0.05 \mu_B$ have been found in $\text{Fe}_{0.5}\text{Cr}_{0.5}$. So even if all the Fe atoms at the nanoparticle surface ($\sim 50\%$ of the Fe atoms in the nanoparticle) are alloyed with Cr to form $\sim \text{Fe}_{0.5}\text{Cr}_{0.5}$, this would not account for the measured Fe moment of $\sim 1\mu_B$ /atom. This implies that the change must be due to a reduced Fe magnetisation at the F/AF interface, which is either magnetically disordered or AF, causing a reduction of the net atomic moment per Fe atom of the entire nanoparticle. From the values of net atomic moment per Fe atom shown in Fig. 4(b), it appears that for isolated nanoparticles $\sim 50\%$ of the Fe atoms are non-magnetic with the remainder ferromagnetic. In a 2 nm nanoparticle 50% of the atoms are on the surface atomic layer so this reduction in magnetic moment is consistent with all the Fe atoms that are in contact with Cr not contributing to the magnetisation of the nanoparticles.

At the highest Fe volume fraction of 20%, which is close to the percolation threshold, the apparent magnetic moment per atom derived from just the equivalent thickness of Fe exceeds the bulk Fe value. We ascribe this to the morphology of the most concentrated sample consisting of regions of correlated Fe nanoparticles inducing in the neighbouring Cr ions some canting, with a net uncompensated moment in the same direction as the Fe moments. The observed exchange bias behaviour, discussed below, supports this interpretation. Indeed, the different degree of spin ordering at the Fe/Cr interface, assumed to be responsible for the different magnetic moment in the three samples, would explain the different behaviour of the Exchange Bias in the three samples, shown in Fig. 5, where the FC and ZFC hysteresis loops at 5K are plotted along with the variation in coercivity and EB with VFF.

The Exchange Bias in the three samples is shown in Fig. 5, where the FC and ZFC hysteresis loops at 5K are plotted along with the variation in coercivity and EB with VFF.

Fig. 5 (a) – (c) Hysteresis loops measured in ZFC (full symbols) and in FC (empty symbols) for samples at different VFF. (d) Variation of coercivity in ZFC (full symbols) and FC (empty symbols) and EB as a function of VFF. (e) Variation of EB as a function of temperature for the three samples.

The shift of the FC hysteresis loop, revealing the existence of the unidirectional anisotropy induced by the F/AF exchange interaction, was observed up to about 30K for all samples (Fig. 5(e)) though the values, obtained from $H_{eb} = (H_c^+ + H_c^-)/2$, are very low. They increase with increasing VFF from $H_{eb}=30(2)$ Oe (5%) to $H_{eb}=50(3)$ Oe (10%) and $H_{eb}=150(7)$ Oe (20%) at 5 K (Fig. 5(d))

Often, the exchange interaction is also responsible for an increase in coercivity after FC, as reported for many EB systems [5,6]. This is due to the presence of a fraction of AF spins, with lower anisotropy, that are dragged by the F moments during their reversal, resulting in an additional source of anisotropy and then leading to an increase of coercivity. In our system this effect is observed only in the most concentrated sample (20% VFF), whose FC coercivity is 147% higher than the ZFC coercivity. In addition, this sample has an EB (H_{eb}), which is a significant fraction of the coercivity, being H_{eb}/H_c (FC)=0.20. On the other hand, both the dilute samples (5% and 10% VFF) show a very similar behaviour, with a small increase of coercivity after FC and a weak exchange bias effect ($H_{eb}/H_c=10\%$). The observed behaviour is markedly different to that observed

in Co nanoparticles embedded in a Mn matrix at different values of VFF[27] where the efficiency of the exchange bias effect was higher for the lowest concentration sample. The combined EXAFS and XMCD study revealed, in the Co@Mn case, the occurrence of a significant alloying at the nanoparticles surface, with the formation of an antiferromagnetic CoMn shell around the Co core[23]. As a consequence, the EB shift was attributed to the exchange interaction between the F Co core and the CoMn AF shell.

In the present system, where no significant alloying was detected by the EXAFS analysis, the marked increase of the atomic moment in the concentrated sample suggests that the long-range interactions among the Fe nanoparticles through the matrix induce in the Cr spins at the interface (characterized by a lower anisotropy) some canting and partial alignment to the F core spins, with the consequent increase in the magnetic moment above the bulk Fe value. Thus, due to the strong exchange coupling between Fe core spins and Cr spins at the interface, most of the latter play the same role as the Fe spins, being pinned by the AF matrix, in the observed shift of the hysteresis cycle below 30 K after field cooling. This picture is supported by the observed hardening of the system; indeed, the nanoparticles magnetization reversal is hindered by the torque exerted by the Cr matrix on the Cr spins at the nanoparticles surface producing the observed increase of coercivity. In fact, a gradual change in the Cr spin alignment should occur moving from one Fe nanoparticle to another and additional investigations will be carried out in order to elucidate the peculiar role of the Cr spins in the concentrated sample.

4. Summary

Extended X-ray Absorption Fine Structure measurements were carried out on Fe clusters in a Cr matrix and for two different FeCr MBE thin alloy films. No comprehensive structural change was observed in any sample, since both Fe & Cr have a bcc structure. Slightly larger interatomic distances found in the Fe clusters embedded in Cr matrices were assigned to the stretch in the respective bulk materials. The large-scale formation of the σ -FeCr alloy was ruled out because of the amplitude of the EXAFS oscillations. At low volume fractions a decrease in the magnetic moment per Fe atom relative to the bulk value was found and was believed to arise from a non-magnetic shell at the surface of each cluster as a result of the exchange interaction between the Fe and Cr. With an increase in volume fraction to 20%, the measured moment per atom was found to be higher than the bulk value indicating that some canting is induced in Cr atoms, bearing a net moment. The system shows a very small EB at the lowest Fe nanoparticle concentration. EB increases with increasing Fe VFF but it remains weak with respect to similar systems of F nanoparticles embedded in AF matrices. It is proposed that the magnetization reversal EB in the highest concentration sample involves the canted Cr spins surrounding the Fe particles.

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