Direct writing of CoFe alloy nanostructures by focused electron beam induced deposition from a heteronuclear precursor

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Abstract

Recently, focused electron beam-induced deposition has been employed to prepare functional magnetic nanostructures with potential in nanomagnetic logic and sensing applications by using homonuclear precursor gases like Fe(CO)\textsubscript{5} or Co\textsubscript{2}(CO)\textsubscript{8}. Here we show that an extension towards the fabrication of bi-metallic compounds is possible by using a single-source heteronuclear precursor gas. We have grown CoFe alloy magnetic nanostructures from the HFeCo\textsubscript{3}(CO)\textsubscript{12} metal carbonyl precursor. The compositional analysis indicates that the samples contain about 80 at\% of metal and 10 at\% of carbon and oxygen. Four-probe magnetotransport measurements are carried out on nanowires of various sizes down to a width of 50 nm, for which a room temperature resistivity of 43 \(\mu\Omega\) cm is found. Micro-Hall magnetometry reveals that 50 nm \(\times\) 250 nm nanobars of the material are ferromagnetic up to the highest measured temperature of 250 K. Finally, the transmission electron microscopy (TEM) microstructural investigation shows that the deposits consist of a bcc Co-Fe phase mixed with a FeCo\textsubscript{2}O\textsubscript{4} spinel oxide phase with nanograins of about 5 nm diameter.

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(Some figures may appear in colour only in the online journal)

1. Introduction

In recent years, the synthesis of magnetic compounds at the nanoscale has generated a large scientific interest due to potential applications in spintronics, thermoelectrics, nanoelectronics, and information technology [1, 2]. Among the various nanofabrication techniques, focused electron beam-induced deposition (FEBID) is increasingly used in research and prototyping applications due to its direct writing and high-resolution capabilities [3, 4]. In FEBID the molecules of a precursor gas injected in a scanning electron microscope (SEM) decompose by interaction with the electron beam, forming a sample during the rastering process. Currently, FEBID relies on homonuclear precursors to fabricate nanostructures. Among the possibly relevant organometallic precursors, most attractive are those which result in virtually complete dissociation under electron impact. In this case, full desorption of the organic ligands under suitable FEBID process parameters can in principle be realized, resulting in clean metallic deposits. Additional design goals for the ideal FEBID precursor are sufficient thermal stability and suitable vapor pressure, i.e., in the range of 10\(^{-2}\)–50 mbar, at about room...
temperature. The precursors Fe(CO)$_8$ and Co$_2$(CO)$_8$ have been shown to yield high metal content deposits, and functional magnetic iron and cobalt nanostructures with a lateral size below 30 nm have been grown [5–7]. However, great care has to be taken in utilizing these precursors, since both are inclined to spontaneous dissociation on activated surfaces, as has been shown in several studies [8, 9]. This indicates the complexity of the task in finding the delicate balance between high metal yield precursors which, at the same time, must be sufficiently stable against spontaneous dissociation. As a second limitation in FEBID, in spite of the many successes in fabrication, characterization, and application of nanostructures grown by this technique in various areas [4, 6, 10, 11], the number of available magnetic materials is still very limited. Mixing precursor gases for the fabrication of binary alloys was proposed about a decade ago [12] and more recently realized by employing dual- or multichannel precursor gas injection systems [13–15]. However, the exact and reproducible control of the deposits’ composition remains a challenge and, more importantly, in the few examples studied so far the overall metal content in the deposits has been found to be below about 60% [15].

Here we show that magnetic alloy nano-structures can be fabricated reproducibly by FEBID with well-defined elemental composition and high metal content by using a heteronuclear precursor. We use the carbonyl HFeCo$_3$(CO)$_{12}$ to fabricate FeCo alloy-phase nanostructures with metal content up to 84% and lateral size down to 50 nm. Magnetotransport measurements and Hall magnetometry reveal the ferromagnetic nature of the deposit up to the highest measured temperature of 250 K. By transmission electron microscopy (TEM) we identify the metallic phase, verify the Co$_3$Fe elemental composition, and characterize a residual oxide phase.

2. Experimental section

2.1. Precursor

2.1.1. Synthesis of HFeCo$_3$(CO)$_{12}$. The synthesis was carried out in a slightly modified procedure as compared to that described by Chini et al [16]. The synthesis and handling of the precursor requires Schlenk techniques to prevent oxidation. The solvents were degassed before use. All chemicals were purchased from Sigma Aldrich.

3.01 g (8.8 mmol) Co$_2$(CO)$_8$ were mixed with 1.02 g (5.2 mmol) Fe(CO)$_8$ in a round-bottom flask. Subsequently, 9.5 ml acetone were added dropwise, and the mixture was stirred at 40°C for 2 h and 12 h at 60°C. Afterwards, the volatile components were removed under reduced pressure (10$^{-2}$ mbar; 25°C) from the initially dark reddish-brown solution and collected in a cooling trap (liquid N$_2$). In the next step, 20 ml purified water were added to the dark solid, and the liquid phase was filtered in a flask containing 30 ml HCl (37%). The yield was increased by adding 20 ml water to the remaining solid and filtering this solution in the same HCl-containing flask. A dark purple solid was formed immediately in the acidic solution, which was stirred for 2 h before the liquid was removed by filtration. The solid was dried in a desiccator over P$_2$O$_5$ for 12 h and recrystallized from toluene to obtain highly crystalline HFeCo$_3$(CO)$_{12}$.

2.1.2. Characterization. The HFeCo$_3$(CO)$_{12}$ was analyzed by infrared (IR) spectroscopy, powder x-ray diffraction (XRD), and single-crystal diffraction, which was used to provide a reference for the powder XRD characterization. The crystallographic information file for the results of the single-crystal structure determination is not included or deposited in a database due to the uncertainty of assigning Fe and Co to specific positions (available from the authors upon request). However, the three equivalent positions should be occupied by Co atoms, as shown in figure 1. Strong disorder is observed for the crystals with tetrahedra tilted by 180° and 50% population density of the specific positions. The simulated powder XRD (Diamond 3) using the single-crystal data is shown as a reference for the experimentally obtained pattern (figure S1) using Cu Kα. Texture effects due to large anisotropic crystallites have not been considered. The strongly anisotropic growth of crystals (needle morphology) is the reason for the weak reflexes at 13 and 14 degrees. However, the positions of the most prominent reflexes are the same. IR spectra do not show any CH- or OH-related signals. Attenuated total reflection measurements show strong absorption bands at 2004, 1966, 1863, and 1104 cm$^{-1}$ for this compound.

2.2. Fabrication

The samples were grown by using a dual beam SEM/FIB (scanning electron microscope/focused ion beam) microscope (FEI, Nova NanoLab 600) equipped with a Schottky...
electron emitter. The HFeCo$_3$(CO)$_{12}$ precursor was heated at 64°C for half an hour before use. The SEM basis pressure was 4.1·10$^{-6}$ mbar, which increased to 4.2·10$^{-6}$ mbar during deposition. The precursor was injected in the SEM via a capillary with a 0.5 mm inner diameter and tilting angle of 50° in close proximity to the focus of the electron beam on the surface substrate. The distance capillary–surface substrate was about 100 μm. The samples were grown on Si(p-doped)/SiO$_2$(10 nm)/Si3N4(100 nm) substrates and contacted with Au(60 nm)/Cr(20 nm) electrodes prepared by ultraviolet photolithography. Three Co-Fe nanowires were grown with different electrical electron beam parameters. Sample A, a small nanowire with ~50 nm width, 10 μm height, and 300 nm length, was prepared with a beam energy, beam current, pitch, and dwell time of 5 keV, 30 pA, 10 μm, and 100 μs, respectively, in the high-resolution deposition mode. Samples B and C, two nanowires with ~700 nm width, 90 μm height, and 5 μm length, were prepared with a beam energy, beam current, pitch, and dwell time of 5 keV, 2.9 nA, 20 nm, and 1 μs, respectively. The growth rate of the nanowires was about 18 nm min$^{-1}$. This good growth rate is compatible with a high electron-induced dissociation cross section and a small sticking coefficient. After growth, sample B was treated by electron irradiation to improve the electrical properties of the material [21]. The irradiation was carried out with 0.12 μC μm$^{-2}$ by using the same electron beam parameters as employed during the deposition. Furthermore, in order to investigate the magnetic behavior of the material, a nanobar with 50 nm width, 250 nm length, and 30 nm height was prepared by using the same deposition parameters used for sample A. The geometrical characterization of the deposits was carried out by using a commercial atomic force microscope (AFM) (Nanosurf, easyscan2) in the dynamic force mode (tapping mode). Finally, the composition of the deposits was investigated by energy dispersive x-ray (EDX) analysis.

2.3. Electrical and magnetotransport measurements

Four-point electrical transport measurements were carried out in the temperature range 2–300 K in a variable-temperature insert mounted in a $^3$He cryostat equipped with a 12 T superconducting solenoid. Standard measurements were performed using a Keithley Sourcemeter 2400 and an Agilent 34420A nanovoltmeter. For measuring the smallest nanowire, we used a four-wire set-up employing a SR830 lock-in amplifier, a SR560 pre-amplifier, and an Eaton RT-20A tunable voltage divider.

2.4. Hall magnetometry

Six adjacent Hall-crosses with areas of $1 \times 1 \text{ μm}^2$ were defined by electron beam lithography, followed by wet chemical etching. (The electronically active area is slightly reduced due to edge depletion.) The micro-Hall sensor was fabricated from a two-dimensional electron system (2DES) based on an AlGaAs/GaAs heterostructure, which lies about 140 nm below the wafer surface [17, 18]. The sensitive 2DES of the Hall structure is covered by a thin top gate (5 nm Cr/40 nm Au) and electronically contacted by annealed planar AuGe/Ni contacts. The CoFe sample was deposited slightly off-center on one cross to maximize the generated Hall voltage $V_H = I/n_e \cdot \langle B_z \rangle$ during magnetic reversal (see figure 7), where $n$ and $I$ denote carrier density and applied current, respectively, and $\langle B_z \rangle$ the magnetic stray field emanating from the end of the sample averaged over the active area of the Hall cross. (For a sketch of the experimental setup, see figure 8.) After deposition the sensor was transferred to a low-temperature cryogenic system equipped with a superconducting solenoid. Above 75 K a variable temperature insert (VTI) cryostat was used. The stray field of the sample, which in first approximation is directly proportional to its magnetization, is detected by measuring the Hall voltage generated in the sensor plane using a standard low-frequency lock-in technique. The samples’ magnetization is switched by an external magnetic field parallel to the sensor plane, which ideally has no perpendicular component and therefore creates no Hall signal. A small misalignment ($\sim 1^\circ$) can be corrected for, and the sensitivity can be further improved by eliminating any background signal by subtracting an empty reference cross (the sensor structure consists of an array of six identical Hall crosses).

2.5. TEM

The TEM lamellae were prepared with a NOVA 200 Nanolab dual-beam system from FEI. TEM investigations were carried out on a Tecnai F20 from FEI with a Schottky Field Emitter at 200 keV. Images were performed with a post-column energy filter (Gatan Imaging Filter, GIF) using an energy slit of 10 eV. The images were recorded zero-loss filtered (i.e., elastically scattered electrons only) on a 2K charge coupled device (CCD). For the image recording and processing (Fourier transformation) the software DigitalMicrograph from Gatan was used.

3. Results

3.1. Deposit composition

The material was characterized by EDX, carried out on deposits of 2 × 2 μm$^2$ lateral size and thickness in the range of 100 nm. Figure 2 depicts a typical EDX spectrum with about 80 at% [Co]+[Fe] and 10 at% [O] and [C]. In figure 3 we plot the concentrations of the elements as a function of the electron beam power. (Figure S2 in the supporting data depicts the concentrations versus the electron beam current.) In the range of the electron beam parameters considered, the global metal content of the deposits, i.e., [Co]+[Fe], is about 80 at%, with a peak value of 84 at%.

After having performed the compositional analysis on the square deposits, we fabricated two kinds of functional magnetic nanostructures: magnetic nanowires, which were characterized by magnetotransport measurements and TEM, and
nanobars, which were used to characterize the material by micro-Hall magnetometry.

3.2. Electrical properties

Transport measurements were carried out on a set of three nanowires. Sample A has a room temperature electrical resistivity of about 43 $\mu\Omega$ cm, (see figure 4). This value is close to the one obtained for cobalt nanowires fabricated by FEBID [19], about a factor of 2 smaller than the value of iron FEBID structures [20, 21] and about a factor of 7 and 4 larger than the values of bulk cobalt and iron [20, 21], respectively. The higher resistivity value of the nanowire compared to those of the bulk samples is expected to be due to the residual impurities of C and O and to the surface scattering as, e.g., described in the theory of Fuchs and Sondheimer [22]. The resistivity of sample A grows monotonically in the range 100–300 K, as expected for metallic samples, and shows a minimum at 87 K. A similar behavior is known from thin epitaxial Fe films, indicating a transition from three-dimensional (3D) to 2D behavior at lower temperature, and has been attributed to weak electron–localization and/or electron interaction effects [23, 24]. Like sample A, samples B and C show metallic behavior (see figure 5). Below 230 K the
resistivity of sample C is smaller than that of sample B. This may be expected, since the metal content of sample C is about 2 at% higher than that of sample B, i.e., 79 at% and 77 at%. Furthermore, both samples are more resistive than sample A, which is due to the different electron beam parameters used during deposition. Although metallic, samples B and C are not far away from a metal–insulator transition. This is also evident from the presence of a resistance minimum at low temperature, which does not relate to a size effect as for sample A, but to a change in transport regime from metallic towards, presumably, variable range hopping at low temperature. Note also that for sample C the minimum is at a lower temperature, as expected for more metallic samples.

3.3. Magnetotransport properties

In figure 6 we plot the Hall resistivity of samples A and C. The Hall resistivity is given by the sum of the ordinary and of the anomalous Hall effects, \( \rho_{xy} = \rho_{0xy} + \rho_{Axy} \), with \( \rho_{0xy} = R_0 \mu_0 H \) and \( \rho_{Axy} = R_A \mu_0 M \), \( \mu_0 H \) being the magnetic induction corresponding to the external field \( H \), \( R_0 \) and \( R_A \) the ordinary and the anomalous Hall coefficients, respectively, and \( M \) the magnetization of the sample in the field direction assuming a demagnetizing factor of the nanowires of \( N \approx 1 \) [26]. The ordinary Hall coefficient \( R_0 \) is determined by the slope of the Hall effect at high magnetic fields, while \( R_A \) is given by the extrapolation of this slope to zero field. In general, according to a unified theory for multiband ferromagnetic metals with dilute impurities [27], the anomalous Hall effect (AHE) shows a crossover from extrinsic to intrinsic behavior by lowering the longitudinal conductance \( \sigma_{xx} \). In particular, in the limit of highly conductive metals, i.e., for \( \sigma_{xx} > 10^6 \Omega^{-1} \text{cm}^{-1} \), the AHE is dominated by the extrinsic skew scattering. Here \( \sigma_{xy} \propto \sigma_{xx} \), \( \sigma_{xy} \) being the anomalous conductivity. In the intermediate metallic region, with \( \sigma_{xx} = 10^4 - 10^6 \Omega^{-1} \text{cm}^{-1} \), the anomalous conductivity is constant, i.e., \( \sigma_{xy} = \text{const.} \), and it is dominated by the contribution of the intrinsic Berry phase [27, 28]. Finally, for \( \sigma_{xx} < 10^4 \Omega^{-1} \text{cm}^{-1} \), in the dirty metal region, \( \sigma_{xy} \propto \sigma_{xx}^2 \). The anomalous and the longitudinal conductivities of our samples are extracted from figure 6, taking into account that \( \sigma_{xy} \approx \rho_{xy} / \rho_{xx}^2 \) and \( \sigma_{xx} \approx 1 / \rho_{xx} \), for \( \rho_{xx} \ll \rho_{xx} \). For sample A we find \( \sigma_{xx} = 2.3 \cdot 10^4 \Omega^{-1} \text{cm}^{-1} \) and \( \sigma_{xy} = 1.9 \cdot 10^2 \Omega^{-1} \text{cm}^{-1} \). These values, which show that sample A is in the intermediate metallic regime, are in the same range as those measured for Fe-based and Co-based FEBID nanowires [19, 21, 25]. For sample C we find \( \sigma_{xx} = 0.85 \cdot 10^4 \Omega^{-1} \text{cm}^{-1} \) and \( \sigma_{xy} = 12 \Omega^{-1} \text{cm}^{-1} \). Similar values are found for sample B. Therefore the transport properties of samples B and C are those of a dirty metal, similarly to non-purified Co-based FEBID nanostuctures and to Fe epitaxial thin films [24]. Finally, we notice that the value of the anomalous Hall resistivity of sample A is about a factor of 3 larger than the value of sample C (see figure 6). This increase, which is also found in Co-based FEBID deposits [6, 19], can be attributed to electron surface scattering due to the reduced size of the samples [29].

Magnetoresistance measurements were carried out on samples B and C in the perpendicular geometry, i.e., with the magnetic field perpendicular to the sample surface, as for the Hall effect measurements. The low signal-to-noise ratio hindered measurements on sample A. Negative

![Figure 6](image-url)
magnetoresistance with values smaller than one $\%$ were found (see the inset of figure 6). These values are about one order of magnitude smaller than those found in Co-based FEBID nanowires [19], which attribute to the higher resistivity of our samples. For the same reason the magnetoresistance of sample B is about 30$\%$ smaller than the one of sample C.

3.4. Hall magnetometry
The magnetic properties of a single CoFe alloy nanobar element with 50 nm $\times$ 250 nm planar size and 30 nm height have been determined by micro-Hall magnetometry. At low temperatures a distinct hysteresis loop is observed during magnetization reversal, proving the ferromagnetic character of the sample (see figure 7). The hysteresis has a characteristic wasp-waist shape with distinct step-like switching. While different mechanisms are known to cause such a characteristic constricted hysteresis shape, two fundamentally different explanations can be considered in the present case: the reversal via several metastable magnetic states, involving the formation and the annihilation of magnetic vortices; or the reversal of a mixed system, consisting of a soft and a hard magnetic phase [30, 31]. In the measured temperature range from 250 K down to about 300 mK, the shape of the hysteresis remains essentially the same (see inset of figure 7). As a general trend, in particular below 15 K, the coercivity increases with decreasing temperature.

3.5. Microstructural characterization
The microstructure of the alloy deposits was investigated by means of high-resolution TEM measurements on samples B and C. TEM lamellae were prepared after covering the samples with a Pt-C protecting layer by FEBID. In figure 9(a) we show a cross-sectional TEM micrograph of sample B. Dark metal nanocrystals with typical diameters of about 5 nm are homogeneously distributed in the sample. The corresponding fast Fourier transformation (FFT) (see inset) shows well-defined rings. In figure 9(c) we depict the radial intensity obtained by azimuthal integration of the FFT. Two groups of peaks are visible: one compatible with the (111), (220), and (311) reflections of the spinel Co$_2$FeO$_4$, the other with the (220) and (110) reflections of CoFe. Note that the CoFe peaks are close to the (002) and (102) peaks of hcp-Co. However, the (100) and (101) hcp-Co peaks are absent (see the positions 4.58 1/1nm and 6.72 1/1nm in figure 9(c)). Therefore, although we cannot exclude the presence of hcp-Co, the systematic absence of the (100) and (101) reflections indicates clearly that the sample contains only Co$_2$FeO$_4$ and CoFe nanocrystals. Differently from sample B, which after deposition was irradiated with an electron dose of about 0.12 $\mu$C$\mu$m$^{-2}$ to decrease the electrical resistivity, sample C was irradiated only during the deposition of the additional Pt-C protecting layer. Therefore, while the entire thickness of sample B was subjected to the electron irradiation, only the upper part of sample C was irradiated during the growth of the protecting layer. In figure 9(b) the TEM micrograph of sample C is shown. The nanocrystals are localized in the upper part of the picture, close to the Pt-C covering layer (see figure 9(d), while the lower part shows an amorphous region of the sample. As a consequence, the corresponding FFT image does not have well-defined rings (see inset).

4. Discussion
Currently, FEBID employs homonuclear precursor gases to prepare nanostructures, which are made of metallic nanograins of one single species embedded in an insulating matrix. High metal content materials, which are required in many
applications, have been obtained mainly by means of a few homonuclear carbonyl precursors like Co$_2$(CO)$_8$, Fe(CO)$_5$, and Fe$_2$(CO)$_9$[7, 19, 20, 32, 33]. In the present work, we propose to use heteronuclear precursors to design high metal content FEBID compounds. This route, known from the preparation of thin film alloys in chemical vapor deposition [34], allows the direct growth of magnetic nanostructures with the same stoichiometry as the heteronuclear precursor. Here we use the HFeCo$_3$(CO)$_{12}$ carbonyl to grow CoFe magnetic nanostructures with a metal content above 80 at%, in a wide range of the electron beam power used. It is remarkable that these results are obtained in non-optimal vacuum conditions, i.e., with a base pressure of 4·10$^{-6}$ mbar. For comparison, we note that Co- and Fe-based deposits grown in non-optimal vacuum conditions have a metal atomic concentration about 15–20% smaller than those found in deposits grown in optimal vacuum conditions, i.e., with a base pressure between 1–2·10$^{-6}$ mbar [19, 20, 25].

As a consequence of the high metal content, the electrical resistivity of the nanowires shows metallic behavior. For sample A, the residual resistivity ratio (RRR) between room temperature and 2 K is about 1.2, which is slightly smaller than the values obtained for highest-metal-content [19] or post-growth-purified [25] FEBID cobalt nanowires. Coherently with this analysis the Hall effect measurements show that sample A is a metal with intermediate characteristics. In samples B and C the RRR is 1.06 and 1.03, respectively. These samples, which are less conductive than sample A, fall into the dirty metal regime. Here we want to underline that the optimization of the metallic behavior of the deposits was out of the scope of the present study. However, an improvement of the metallic characteristics might be straightforward in better vacuum conditions or by applying post-growth purification treatments.

We now turn to the microstructural investigations. After deposition the samples were covered with a Pt-C FEBID protecting layer in order to prevent oxidation and to avoid damages during the TEM lamella preparation. In sample C, the covering process induces a partial crystallization, mainly located in the upper layers, where the electron irradiation is

Figure 9. High resolution TEM micrographs of samples B (panel (a)) and C (panel (b)). In the inset the corresponding FFT is plotted. (c) Radial intensity of the FFT of sample B. The peaks belong to the Co$_2$FeO$_4$ spinel oxide and to the bcc CoFe phases. (d) Zoom-out of the micrograph of sample C. The Pt-C cover layer, the CoFe sample, and the Si$_3$N$_4$ substrate are visible (see text for details).
more effective (see figure 9(d)). The amorphous-to-crystalline transformation is more evident in sample B, which was specifically irradiated after deposition to increase the conductivity. From the TEM investigation of sample B (see figure 9(a)), and, although less clearly, of sample C (see figure 9(b)), it is evident that the material investigated contains Co-Fe and Co2FeO4 nanocrystals. Furthermore, from the TEM and the EDX analyses, we find that the ratio [Co]/[Fe] is equal to 3 and that the elements’ concentrations are about 60 at% for Co, 20 at% for Fe, 10 at% for O, and 10 at% for C. By making the assumption that Co2FeO4 binds all the oxygen atoms available, it follows that 7.5 at% of the metal present in the material belongs to the spinel oxide phase, while the remaining 72.5 at% forms the bcc Co-Fe crystal phase. Therefore one concludes that the Co-Fe nanocrystal phase present in the material is the dominant one.

In order to interpret the micro-Hall magnetometry data and, in particular, the characteristic wasp-waist shape of the hysteresis loop measured at low magnetic fields (see figure 7), two different scenarios should be considered. First, the behavior of the system during reversal may relate to the intrinsic nature of the material, i.e., to the presence of the CoFe2O4 and Co-Fe phases. Indeed, it is known from the literature that two-phase magnetic systems can generate a constriction in the hysteresis loop, as found in our experiment [31, 35]. However, even if two phases coexist, according to the previous considerations, the Co2FeO4 spinel oxide phase is not the dominant one in our deposits. Therefore, we would not expect a double jump in the hysteresis, but only one coercivity. Apart from this, we remark that the nanobar for magnetic measurements was not irradiated after growth, as in sample A, and, therefore, is most likely in the amorphous phase. In this case a double coercivity in the hysteresis is not expected. Second, the reversal may take place via metastable states involving the formation and annihilation of magnetic vortices, as found in Co and Fe circular disk-like structures [30]. In particular, it is known that switching events between distinct micromagnetic configurations strongly affect the stray field emanating from the sample, which results in step-like changes in the hysteresis loop measured by micro-Hall magnetometry [30]. Although we cannot exclude a contribution to the shape of the hysteresis due to the formation of a crystalline phase, we consider the reversal via metastable micromagnetic configurations the most likely explanation for the interpretation of our data.

5. Conclusions

In this work, we have fabricated CoFe alloy magnetic nanostructures by means of FEBID from the HFeCo3(CO)12 heteronuclear carbonyl precursor. A complete characterization of the composition, the microstructure, and the electrical and magnetic properties of the deposits has been carried out. The bi-metallic and ferromagnetic nature of the nanocomposite and the high metal content obtained, together with the high resolution and direct-writing capabilities offered by the FEBID technique, make these functional magnetic nanostructures attractive for prototype applications in spintronics and information technology. Furthermore, our results indicate that heteronuclear carbonyl precursors with multiplet-metal bonding might form a very promising precursor class for future development of FEBID towards a direct-writing technique for multi-component metallic nanostructures.

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