Ordered arrays of ferromagnetic, compositionally graded \( \text{Cu}_1 - x \text{Ni}_x \) alloy nanopillars prepared by template-assisted electrodeposition†

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Periodic arrays of compositionally graded Cu–Ni alloy nanopillars (100 or 200 nm in diameter and 450 nm in height) have been fabricated by means of potentiostatic electrodeposition into patterned Au/Ti/Si(111) substrates using a single electrolytic solution. The pillars do not bend after the template removal but remain straight and perfectly attached to the substrate. The average composition of the alloy nanopillars can be tuned between 34 and 70 at% of Ni by varying the applied potential from a \([\text{Ni}(\text{II})]/[\text{Cu}(\text{II})] = 95.2\) electrolytic solution. The nanopillars are Ni-rich at the bottom with a compositional gradient about \(\pm 20-25\%\) above/below the average composition. The magnetic characterization reveals that all the nanopillars are ferromagnetic with coercivity values around 100–150 Oe (in-plane) and 200–500 Oe (out-of-plane), which are larger than those for continuous films of similar average composition. Atom probe tomography of the Cu-rich Cu–Ni nanopillars indicates the segregation of Ni, which would explain their unexpected magnetic character. The graded-anisotropy ferromagnetic Cu–Ni pillars could be potentially applied in several fields including micro/nano-electromechanical systems (MEMS/NEMS), magnetic recording media or spintronics.

1 Introduction

Nanocrystalline (nc) metallic films are known to benefit from enhanced physical properties (e.g. mechanical, magnetic and optical) as compared to their coarse-grained counterparts. \(^1\) One of the key applications of nc-metallic films is in the micro/nanoelectromechanical systems (MEMS/NEMS) field. \(^2\) To successfully implement nc-metallic materials in MEMS/NEMS devices, the miniaturization (i.e. nanostructuring) of a continuous nc thin film is required. Such miniaturization is usually performed by electron beam lithography (EBL) combined with sputtering, thermal evaporation or e-beam evaporation. \(^3\) However, when using EBL combined with lift-off in narrow pores, sputtering only allows thicknesses below a few tens of nm. \(^4\) Another technique often used to fabricate metallic MEMS/NEMS is focused ion beam (FIB) lithography. \(^5\) However, this technique also suffers from some drawbacks: (i) it requires a long time to manufacture individual samples, which significantly reduces the output and (ii) Ga\(^+\) ions are easily incorporated into the objects being fabricated, thus modifying their chemical composition and structure and, in turn, their physical properties. \(^6\)

To overcome all these issues, new cost-effective methods able to miniaturize high aspect ratio nc-metallic films at relatively fast rates are being explored. Electrodeposition is a non-vacuum, simple and well-known method for depositing micrometer-thick films of metals and alloys onto conductive substrates. The fabrication of large aspect ratio metallic nanowires in the cylindrical pores of membranes such as anodized alumina, polycarbonate and polymethyl methacrylate (PMMA) is well documented in the literature. \(^5\) The diameter of the nanowires obtained ranges from hundreds down to tens of nanometres depending on the pore size of the template. The application of these 1D nanostructures in sensors, field-emission and magnetic devices often requires the nanostructures to be supported by a substrate (Si, glass, etc.) in order to ensure the robustness and mechanical stability of the device. For this reason, the nanowires are kept embedded in the template to

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serve as high-performance platforms. However, in certain applications, the template needs to be removed because it can either interfere with or hinder the nanowires’ response. In such cases, when the template is etched away, the nanowires often bend and partially (or completely) collapse. Thus, for many applications where the active material needs to be exposed, nanopillars (with intermediate length, filling the gap between dots and nanowires) are more adequate than nanowires (with high aspect-ratio) due to their superior mechanical stability. Self-organized metallic ferromagnetic nanopillars have been proposed not only as MEMS/NEMS counterparts but also as building blocks in spintronic devices. Interestingly, there is also an increasing interest in magnetic oxide nanowires, both ferrimagnetic and diluted magnetic semiconductors. Although disordered or pseudo-ordered nanowire arrays of many different elements and alloys have been grown by electrodeposition in anodic aluminium oxide (AAO) templates, research on the growth of nanopillars is far scarcer. Up to now, a few pure metals (e.g. Co, In, Sn, Au and Cu) have been prepared using EBL combined with electrodeposition. Examples of alloy nanopillars grown using this approach are even more limited (NiFe is almost the only example available), mostly due to the difficulty in achieving the co-deposition of metals in nano-sized cavities. Compared to AAO membranes, e-beam lithographed substrates offer the possibility to prepare fully ordered arrays of diverse nanostructure geometries on pre-defined substrate areas, enabling easier integration in NEMS and spintronics devices, as well as their implementation in high-density magnetic storage systems.

Regarding magnetic recording media, interest has been directed in recent years toward the development of the so-called “graded anisotropy materials” as possible future magnetic recording media. These materials exhibit a tunable and gradual variation of their magnetic anisotropy leading to a controllable decrease of coercivity without a significant loss of the thermal stability. Although the first graded anisotropy materials consisted of Co/Pd multilayers with variable Co thickness, recently, compositionally graded FePtCu films have also been demonstrated. Remarkably, in spite of the numerous works on continuous films with graded magnetic anisotropy, reports on patterned graded anisotropic structures remain rather scarce. Thus, development of novel types of lithographed pillars with graded magnetic anisotropy holds promise for the improvement of current state-of-the-art in patterned recording media.

Cu-Ni is a binary isomorphous alloy that can be electrodeposited from aqueous solutions. Alloying Cu with Ni at the nanoscale has been demonstrated to be a suitable way to synergistically combine the properties of the two pure metals, thereby expanding the range of technological applications. This has been proven not only for nanocrystalline thin films, where tuneable ferromagnetic and mechanical properties have been demonstrated as a function of the Cu/Ni ratio, but also for disordered/pseudo-ordered nanostructures electrodeposited in AAO (e.g. nanoparticles and nanowires). In all these cases, the composition is kept constant along the thickness/length of the objects. Interestingly, the possibility of producing fully ordered arrays of graded-anisotropy ferromagnetic Cu-Ni nanopillars on pre-defined substrate areas that can be potentially useful for a variety of applications (magnetic MEMS/NEMS, spintronics or recording media) has not been attempted.

In this work, a straightforward approach to obtain compositionally graded Cu$_{1-x}$Ni$_x$ nanopillars by electrodeposition onto e-beam lithographed silicon-based substrates is presented. Template-free, ordered and straight arrays of compositionally graded nanopillars have been obtained in all cases after removing the e-beam resist used for patterning. The structural characterization demonstrates the compositional gradient along the length of the nanopillars. The obtained pillars, with diameters of 100 nm (300 nm pitch size) and 200 nm (600 nm pitch size), and aspect ratios of 4.5 and 2.25, respectively, exhibit a tuneable magnetic response as a function of the average Cu/Ni ratio.

2 Results and discussion

Fig. 1 shows the layered structure of the substrates used for electrodeposition (left image). The role of the upper 15 nm thick Ti layer was to protect the Au seed-layer during the lithography process and subsequent wafer slicing into single chips (substrates). This layer was selectively etched by a 0.1% HF solution immediately before electrodeposition to ensure the adhesion of the nanopillars to the Au surface after resist removal; otherwise, the nanopillars become detached from the substrate due to the mismatch between the crystallographic structure of Ti (hexagonal closed packed, P6$_3$mmmc spacial group) and that of Cu$_{1-x}$Ni$_x$ structures (face-centred cubic, Fmm3 spacial group) (Fig. S1a†). Nanopillars grown onto the Au seed layer show good adherence to the substrate since their crystallographic structures are the same.

An electrolytic solution containing a [Ni(n)][Cu(n)] molar ratio of 29 and a number of additives to control metal co-deposition and grain size were initially used for electrodeposition. This bath was proven to be effective in achieving the deposition of nanocrystalline, smooth, continuous Ni-rich Cu-Ni films (up to 55 at% Cu). However, in contrast to the results in continuous films, the composition of the nanopillars could not be tuned as a function of the applied potential/current density using this electrolytic solution. Indeed, almost pure Cu nanopillars were obtained regardless of the applied potential/current density, likely because the potential inside the
cylindrical cavities is not low enough to achieve the necessary overpotential for the reduction of Ni. To overcome this issue, the bath was re-formulated by increasing the \([\text{Ni(II)}]/[\text{Cu(II)}]\) molar ratio in solution to 95.2, keeping the overall ion concentration constant. The next step was to choose the appropriate potential of deposition potentials so as to obtain deposits with tuneable Cu/Ni ratios, and also to set the deposition time to avoid overplating (Fig. S1b†); otherwise, mushroom-like nanostructures are obtained upon stripping of the resin, as observed in other work34 (inset of Fig. S1b†). Note that for recording or magnetic sensor applications, overplating must be avoided because it can induce undesirable magnetic dipolar and exchange interactions between nanodisks.\(^\circ\) The optimized parameters for the deposition of non-overplated structures (namely, the concentration of metal salts in solution, applied potential, deposition time, stirring rate and temperature) are shown in Table 1. The high \([\text{Ni(II)}]/[\text{Cu(II)}]\) ratio was also selected to induce significant composition gradients along the alloy nanopillar’s axis.

The potentiostatic curves for \(\text{Cu}_{1-x}\text{Ni}_x\) deposition onto the lithographed substrates from the optimized bath are shown in Fig. 2a. It can be observed that the absolute value of current density increases as the applied potential is made increasingly negative. The applied potentials (from \(E = -0.900\) V to \(E = -1.150\) V) allowed growing nanopillars with varying composition. The nickel percentage obtained from energy dispersive X-ray spectroscopy (EDXS) increased, on average, from \(34\) at% to \(70\) at% as \(E\) was made more negative (Fig. 2b), since greater overpotentials favour nickel discharge.

Pure Cu and Ni metals were also electrodeposited onto the lithographed substrates for comparison. Importantly, pure Cu nanopillars could be obtained from the same bath formulation provided that the applied potential was made more positive. In particular, for deposition potentials around \(-0.8\) V, only Cu is electrochemically reduced (Fig. 2b). To deposit pure Ni, though, the bath composition had to be modified since Ni deposition is always accompanied by some Cu deposition even at high overpotentials. Therefore, the Cu sulphate in solution was replaced by Ni sulphate while keeping the overall ion concentration constant (Table 1).

<table>
<thead>
<tr>
<th>Electrodeposition conditions</th>
<th>NiSO(_4) (g L(^{-1}))</th>
<th>CuSO(_4) (g L(^{-1}))</th>
<th>(E) (V)</th>
<th>(t) (s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Pure Ni</td>
<td>630</td>
<td>0</td>
<td>(-0.900)</td>
<td>30</td>
</tr>
<tr>
<td>Pure Cu</td>
<td>624</td>
<td>6.24</td>
<td>(-0.800)</td>
<td>(T = 30) °C</td>
</tr>
<tr>
<td>Cu–Ni alloy 34 at% Ni</td>
<td>624</td>
<td>6.24</td>
<td>(-0.900)</td>
<td>30</td>
</tr>
<tr>
<td>41 at% Ni</td>
<td>-1.000</td>
<td>30</td>
<td></td>
<td></td>
</tr>
<tr>
<td>45 at% Ni</td>
<td>-1.025</td>
<td>15</td>
<td></td>
<td></td>
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<tr>
<td>70 at% Ni</td>
<td>-1.150</td>
<td>5</td>
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</table>

\(70\) at% \(\text{Ni}\) alloy nanopillars is shown in Fig. 3a. 70° tilted (side-view) typical FESEM images of 200 nm and 100 nm-diameter nanopillars are displayed in Fig. 3b and c, respectively. Both images show well defined, periodically ordered nanopillars with a constant height of 450 nm. The diameter at the top is slightly smaller than at the bottom (near the substrate), so the structures are in fact hill-shaped. Interestingly, the Cu–Ni nanopillars are compositionally graded along their long axis (Fig. 4). The EDX mappings show, within the lateral resolution (roughly around 50 nm), that all the different alloyed nanopillars are Ni-rich at the bottom with the Cu/Ni ratio progressively increasing towards the top, \(i.e.,\) a compositional gradient (see Fig. 4). The degree of gradient depends on the average \(\text{Cu}_{1-x}\text{Ni}_x\) composition. For example, for \(\text{Cu}_{0.55}\text{Ni}_{0.45}\), the gradient is moderate and the composition ranges from \(\text{Cu}_{0.46}\text{Ni}_{0.54}\) at the bottom to \(\text{Cu}_{0.68}\text{Ni}_{0.32}\) at the top, \(i.e.,\) the Ni content increases by 19% between the top and the bottom (see Fig. 4b). On the other hand, for pillars with higher Ni contents, \(e.g.,\) \(\text{Cu}_{0.37}\text{Ni}_{0.63}\), the gradient becomes considerably higher and can be as much as 30% between the top and the bottom (Fig. 4c). Considering that the molar concentration of \(\text{Ni}(u)\) is almost 100 times as that of \(\text{Cu}(u)\), the observed gradation indicates that the codeposition is strongly dependent on

![Fig. 2](image-url)
the mass transport conditions and the current distribution inside the cylindrical holes. Two factors may be the reason for favoured deposition of nickel at the bottom of the pillar: (a) copper deposition is controlled by diffusion; (b) the confinement of the deposition event at the cavities of the template can lead to local increases of current density over the deposition area. Once the growing alloy approaches the bulk solution, the deposition of copper then becomes more favoured.

Magneto-optical Kerr effect (MOKE) measurements indicate that the Cu$_{1-x}$Ni$_x$ nanopillars (diameter of 100 nm) are ferromagnetic at room temperature in the whole compositional range.

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**Fig. 3** FESEM images of Cu$_{0.66}$Ni$_{0.34}$ nanopillars: (a) planar view of 200 nm diameter nanopillars, and (b and c) 70° tilted images for the 200 and 100 nm diameter nanopillars, respectively.

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**Fig. 4** (a) FESEM image of two Cu$_{0.55}$Ni$_{0.45}$ 200 nm-diameter nanopillars, (b) the corresponding EDX mapping analysis along their length, where green color denotes Cu and red color Ni. (c) FESEM image of a Cu$_{0.37}$Ni$_{0.63}$ 200 nm-diameter nanopillar, for which the Ni percentage at different locations along its length is indicated.

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**Fig. 5** Room-temperature MOKE hysteresis loops for: (a) Cu–Ni nanopillar arrays with a diameter of 100 nm, measured in longitudinal configuration, (b) Cu–Ni nanopillar arrays with a diameter of 100 nm, measured in polar configuration and (c) continuous (unpatterned) films, measured in longitudinal configuration. The averaged Cu–Ni compositions are indicated for the nanopillars. The behaviour of pure Ni is also shown for the aim of comparison.
range. The hysteresis loops measured in longitudinal configuration (with the magnetic field applied along the in-plane direction, i.e., perpendicular to the nanopillar axis) are rather square (except for the average Cu0.37Ni0.63 composition) and exhibit coercivity values, $H_c$, around 100–150 Oe (Fig. 5a). Similar results were recorded for the nanopillars with a diameter of 200 nm. The occurrence of ferromagnetic behaviour in all the investigated Cu$_{1-x}$Ni$_x$ nanopillars, particularly for the Cu-rich ones, (e.g., Cu$_{0.55}$Ni$_{0.45}$ which is paramagnetic in bulk and thin film form)27,28 is consistent with the presence of a compositional gradient.

Namely, even if the average Ni content is below the range for room temperature magnetism in these alloys, the bottom of the nanopillars has sufficiently high Ni contents to make the pillars ferromagnetic. However, two ferromagnetic contributions seem to be present in the hysteresis loops corresponding to the Cu$_{0.30}$Ni$_{0.70}$ and Cu$_{0.37}$Ni$_{0.63}$ nanopillars. In principle, if the compositional gradient was continuous, exchange coupling should render smooth hysteresis loops.27 Thus, the two-component loops may indicate in some alloys the segregation of various non-exchange coupled Ni-rich regions with different Ni contents (i.e., separated by non-magnetic regions). To elucidate the origin of such an unexpected ferromagnetic response, some of the electroplated nanopillars were prepared for atom probe tomography (APT) analysis in order to locally probe the distribution of Cu and Ni elements. APT is a very powerful technique although the spatial resolution of atom probe tomography is slightly anisotropic, meaning its resolution along the analysis axis (typically the Z-axis) can be sub-nanometer and laterally (X and Y directions) ca. 1 nm. As shown in Fig. 6a, the distribution of Cu and Ni elements at the top of Cu$_{0.53}$Ni$_{0.45}$ nanopillars is not homogeneous, indeed, showing the presence of Ni-rich regions embedded in a Cu-rich matrix. Moreover, the concentration profile across the cylinder depicted in Fig. 6b shows that the Ni content at the centre of it is around 57 at% (Fig. 6c), close to the threshold for ferromagnetism. Since the Ni content progressively increases toward the bottom of the pillar, one can straightforwardly assume that the probability of finding clusters with higher Ni contents increases toward the bottom of the pillar, thus causing the ferromagnetic response. Note that the occurrence of Ni-rich clusters could be understood in mind that the enthalpy of mixing between Cu and Ni is slightly positive.29

Remarkably, square hysteresis loops are also obtained using a polar MOKE (i.e., applying the magnetic field along the nanopillars’ axis) for pure Ni nanopillars with a diameter of 100 nm (Fig. 5b). In fact, the coercivity in this case ($H_c \sim 500$ Oe) is much larger than for the longitudinal hysteresis loops. Such an increase of $H_c$ is probably due to the role of shape anisotropy, which is particularly pronounced for the nanopillars with a smaller diameter (i.e., with a larger aspect ratio). Again, two ferromagnetic contributions are observed using a polar MOKE in the Cu$_{0.37}$Ni$_{0.63}$ and Cu$_{0.30}$Ni$_{0.70}$ nanopillars (Fig. 5b). The coercivity is larger for pure Ni than for Cu–Ni nanopillars because the addition of Cu tends to progressively reduce the magnetocrystalline anisotropy.48

The coercivity values obtained from the nanopillars are larger than those corresponding to Cu–Ni thin films with the same Cu/Ni ratio (Fig. 5c). The increase in $H_c$ can be ascribed to size effects, i.e., either to the formation of single domain states or the hindrances imposed by the limited lateral size of the nanopillars on the propagation of domain walls. Note that, as expected, continuous films with the Cu$_{0.55}$Ni$_{0.45}$ composition are not ferromagnetic. While for continuous Cu–Ni films the coercivity tends to progressively decrease as the Cu content increases (basically due to the decrease of the magnetocrystalline anisotropy), the evolution of coercivity with the Cu percentage in the Cu–Ni pillars is more complex. This can be ascribed to the interplay between different effects. First, the formation of Ni-rich and Cu-rich clusters can give rise to different magnetic contributions to the hysteresis loops, often exchange coupled to each other. The size of these clusters (and the concomitant surface anisotropy) and the exchange, dipolar and Ruderman–Kittel–Kasuya–Yosida (RKKY) interactions between them (similar to Cu/Ni multilayer nanowires41) can influence the magnetic state of the cluster and their magnetization reversal, thus strongly affecting $H_c$. For example, the large coercivity observed for Cu$_{0.53}$Ni$_{0.45}$ in Fig. 5a could be due to the isolated ferromagnetic regions surrounded by a non-magnetic Cu-rich matrix, resulting in even larger coercivity than for pure Ni pillars. In addition, in the nanopillars, the magnetocrystalline anisotropy competes with the shape anisotropy. Shape anisotropy would promote larger coercivity along the pillar axis (i.e., polar configuration). However, the formation of non-magnetic inclusions can have a negative effect on the shape anisotropy, eventually resulting in small changes in coercivity when comparing polar with longitudinal hysteresis loops (e.g., Cu$_{0.37}$Ni$_{0.63}$).

Fig. 6 (a) Atom map (Ni and Cu atoms are green and orange dots, respectively) of a local-electrode atom-probe (LEAP) tomography reconstruction of a Cu$_{0.55}$Ni$_{0.45}$ nanopillar with volume dimensions in nm (actual dimensions are: 38 nm × 38 nm × 26 nm). The isocentre surfaces, displaying 49 at% Cu (orange) and 50 at% Ni (green), reveal where local concentration gradients exist. The clustered regions are higher (>50 at%) in Ni. Concentrations within one such Ni rich region is shown in (c) along the 8 nm diameter cylinder depicted in (b), using 0.5 nm bin sizes along the length of the cylinder.
3 Conclusions

Ordered arrays of Cu$_{x}$–Ni$_{y}$ (0.34 ≤ x ≤ 0.70) nanopillars, with diameters of 100 nm and 200 nm and length of 450 nm, have been obtained by e-beam lithography combined with electrodeposition onto Au/Ti/Si(111) substrates. The structures remain adhered to the substrate and do not collapse after resin removal. The nanopillars are compositionally graded along their height, as demonstrated by EDX analyses, and exhibit tuneable ferromagnetic behaviour depending on the average Cu/Ni ratio. Interestingly, their coercivity is larger compared to continuous films of the same composition. Atom probe tomography of Cu-rich nanopillars reveals the presence of Ni-rich clusters that account for the observed ferromagnetism in spite of the large Cu content. Owing to their graded-anisotropy ferromagnetic properties, the synthesized Cu–Ni nanopillar arrays could find applications in a number of fields where the active material needs to be fully exposed (template-free) and mechanical stability is a prerequisite. Among them, magnetic NEMS actuation and recording media are envisioned.

4 Experimental details

4.1 Substrates preparation

First, Ti (100 nm)/Au (400 nm)/Ti (15 nm) were evaporated onto Si(111) substrates. An electron sensitive resist (ZEP-520A) of about 500 nm in thickness was then spin-coated onto the substrate and patterning was carried out by electron-beam lithography (EBL) with the CRESTEC CABL-9500C Nanolithography equipment. 60 × 60 µm$^2$ arrays of cylindrical holes of 100 nm and 200 nm in diameter and pitch sizes of 300 nm and 600 nm, respectively, were patterned. Following the lithographic procedure, the resist was developed in anisol.

Before electrodeposition, all samples were submerged in HF solution for 1 minute to remove the 15 nm-thick Ti layer to allow the deposition of the metals directly onto the Au layer. Finally, the substrates were dipped in diluted sulphuric acid to remove oxides and organic residues. Before deposition, the substrates were immersed in the electrolyte for 10 min, which had been previously deaerated with argon gas.

4.2 Electrodeposition

Cu, Ni and the Cu–Ni alloy were deposited by direct current electrodeposition in a one-compartment thermostatted three-electrode cell using a PGSTAT302N Autolab potentiostat/galvanostat (Ecochemie). The analytical solutions were prepared from analytical grade reagents and Millipore Milli-Q water (18 MΩ cm).

Cu and the CuNi alloy were deposited from a bath containing 6.24 g L$^{-1}$ CuSO$_4$·5H$_2$O, 624 g L$^{-1}$ NiSO$_4$·6H$_2$O and 70 g L$^{-1}$ Na$_2$C$_6$H$_5$O$_7$·2H$_2$O (sodium citrate), 0.2 g L$^{-1}$ NaC$_6$H$_5$SO$_4$ (sodium dodecylsulfate) and 1.4 g L$^{-1}$ C$_2$H$_5$NO$_3$S (saccharine). The electrolyte volume was 50 mL. The pH was fixed at 4.5 and the temperature at 30 °C in all cases. Pure Ni was deposited from an electrolyte containing 630 g L$^{-1}$ NiSO$_4$·6H$_2$O (nickel sulphate) and the same amount of the other chemicals (sodium citrate, sodium dodecylsulfate and saccharine).

The Si(111)/Ti (100 nm)/Au (400 nm)/ZEP-520A layers were used as working electrodes, which were positioned vertically into the electrode. A double junction Ag|AgCl (E = +0.210 V per SHE) reference electrode was used with a 3 M KCl inner solution and a 1 M Na$_2$SO$_4$, outer solution. All the potentials are referred to this electrode. A Pt sheet was used as the counter electrode.

Deposition was conducted potentiostatically, by applying a constant potential in the range of −0.800 to −1.150 V, under mild stirring (200 rpm) using a magnetic stirrer bar. After the deposition, the samples were thoroughly rinsed in water and the ZEP-520A resin was removed with dimethylacetamide (DMAc) at 35 °C for 1 min to get template-free nanopillars.

4.3 Characterization

FESEM images of the nanopillars were obtained on a MERLIN FESEM microscope operated at 5 kV. The chemical composition was determined by energy dispersive X-ray (EDX) spectroscopy at 7 kV. Metal proportions are expressed in atomic percentage (at%). Mapping analysis was carried out with the aim to clarify element distribution through the nanopillar length. Aspect-ratio values and nanopillar element distribution images were taken at a 70° tilt angle.

The magnetic properties of nanopillars were investigated by acquiring hysteresis loops using a magneto-optical Kerr effect magnetometer (MOKE) in longitudinal and polar modes (perpendicular and parallel to the long axis of the structure, respectively) with a maximum applied field of 825 Oe at room temperature.

Atom probe tomography of a single pillar was achieved using a Cameca LEAP 4000-X HR. Controlled field evaporation from the specimen cooled to 33 K was instigated by applying 150 pJ laser energy pulses (of 355 nm wavelength) at a pulse rate of 160 kHz, resulting a Ni$^{++}$/Ni$^+$ charge-state-ratio of 0.2. The resulting reconstruction and analysis were completed with IVAS 3.6.4.

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Notes and references


