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Enhancement of damping constant in Nd-doped L1₀-FePt at elevated temperatures for HAMR

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ABSTRACT

Hard disk drives (HDDs) are main storage devices in data centers handling petabyte-scale AI data, and achieving higher recording density with heat-assisted magnetic recording (HAMR) requires media material design that ensures a sufficiently large damping constant (α) at elevated temperatures to enable reliable magnetization reversal. We investigated Nd-doped L1₀-FePt continuous films to clarify the effect of rare-earth element doping on α at elevated temperatures where recording occurs. (FePt)_{100-x}(Nd)_x thin films with Nd content ranging from 0.0 to 4.1 at.% were prepared by sputtering and characterized for microstructure and magnetic properties. Nd was completely rejected from FePt, and Nd and Pt segregated into Fe-deficient void regions in the FePt layer. Nevertheless, enhanced spin-orbit coupling (SOC) locally at the FePt/Nd interface increased magnetic anisotropy with a small amount of Nd doping. The magnetization dynamics were measured by time-resolved magneto-optical Kerr effect (TRMOKE) from room temperature to near T_C . Effective α near T_C increased from 0.048 (undoped) to 0.066 (Nd 4.1 at.%). These findings demonstrate that rare-earth element doping can modulate α even in FePt, which already exhibits strong SOC, and highlight a media design strategy to optimize α for HAMR performance.

1. Introduction

The volume of data being generated and consumed is accelerating rapidly, driven by technologies such as Artificial Intelligence (AI), genomic sequencing, large-scale simulations, and high-resolution video streaming. These applications produce petabytes of information that must be stored cost-effectively and reliably for long periods [1,2]. Among large-capacity storage solutions, hard disk drives (HDDs) offer key advantages, delivering significantly lower cost per terabyte and proven long-term reliability for archival use compared to solid-state drives (SSDs). Although SSDs are utilized in data centers for high-performance critical workloads requiring fast access and low latency, HDDs remain the primary choice for large-scale capacity storage. Their cost efficiency and durability make them essential for meeting the rapidly growing demand for data storage [3].

Industry forecasts project that total HDD shipment capacity will

grow at a 20 % compound annual growth rate (CAGR) from 2024 to 2029, effectively doubling every 4 years [4]. At the same time, data centers face a major challenge: power consumption. They already require vast amounts of electricity, and simply increasing the number of drives to meet capacity needs would further raise energy usage and operational costs. As a result, increasing the capacity of each individual drive is critical to prevent proportional growth in energy consumption and carbon footprint. Achieving higher capacity without increasing energy usage depends on achieving areal density capability (ADC).

In HDDs, one bit consists of several media grains and the number of grains in one bit determines the signal-to-noise ratio. Reducing the grain size is necessary to increase ADC while maintaining the number of grains per bit. On the other hand, a small grain size results in a loss of thermal stability which is described by $K_u V / k_B T$ where K_u , V , k_B , T are magnetic anisotropy constant, grain volume, Boltzmann constant, and absolute temperature, respectively. Although high- K_u material maintains thermal

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stability with small grain size, limited magnetic field from the writing coil is insufficient to write on high- K_u media. This problem is known as magnetic recording trilemma. Heat-assisted magnetic recording (HAMR) enables writing on high- K_u media by locally heating the recording layer near its Curie temperature (T_C). $L1_0$ -FePt is employed in the recording layer of HAMR media because it exhibits high- K_u at room temperature and possesses a T_C that is sufficiently high for thermal stability yet low enough to be reached by laser heating during the write process. The Advanced Storage Research Consortium (ASRC) proposed an HDD technology roadmap in which recording technology shifts to HAMR to increase ADC to over 4 Tbit/in² [5].

During HAMR writing, magnetization reversal occurs via precessional dynamics described by the Landau-Lifshitz-Gilbert (LLG) equation:

$$\frac{d\mathbf{m}}{dt} = -\gamma\mu_0(\mathbf{m} \times \mathbf{H}_{\text{eff}}) + \alpha\left(\mathbf{m} \times \frac{d\mathbf{m}}{dt}\right) \quad (1)$$

where $\mathbf{m}$, γ , $\mathbf{H}_{\text{eff}}$, and α are the unit magnetization vector, the gyro-magnetic ratio, the effective magnetic field, and the Gilbert damping constant, respectively. α controls the relaxation rate of precessional motion during magnetization reversal. At elevated temperatures near T_C , high thermal fluctuations can cause magnetization back-switching after reversal, which is a key contributor to bit error rate [6]. HAMR mitigates this by applying a large magnetic field and heating the media above T_C [7]. On the other hand, these conditions broaden the write width which limits ADC, and introduce reliability issues such as head smear and near-field transducer (NFT) degradation [8,9]. Increasing α in HAMR media offers a pathway to suppress back-switching, enhance switching probability, and reduce jitter [6,10–14]. Moreover, higher α enables faster switching. This is critical under steep thermal gradients that shorten the available switching time, which is necessary for higher ADC [15].

Previous studies on NiFe films have shown that doping with rare-earth elements can enhance α . Bailey *et al.* and Reidy *et al.* demonstrated that rare-earth doping enhances α through increased spin-orbit coupling (SOC) induced by the localized rare-earth 4f shell [16,17]. Woltersdorf *et al.* also explained the enhancement of α by slow-relaxing impurity mechanism [18]. In NiFe, a 3d transition metal alloy, the SOC is relatively weak, whereas in FePt, an intermetallic compound consisting of 3d transition metal Fe and 5d transition metal Pt, the heavy element Pt contributes to an enhancement of the SOC through its 5d bands. While SOC enhancement by doping rare-earth elements has been reported in NiFe, rare-earth doped FePt has been theoretically predicted to exhibit enhanced magnetocrystalline anisotropy energy, which is intrinsically driven by strong SOC [19]. However, in FePt, the impact of rare-earth doping on α , potentially mediated through a SOC-related mechanism, has not yet been experimentally confirmed. Consequently, clear guidelines to enhance α in FePt thin films have not been established for HAMR-HDD material design [20].

Among rare-earth elements that have a large orbital moment, Fe₅₀Pt₄₄Nd₆ is reported to retain the $L1_0$ -ordering and a high- K_u , which are requirements for the application of HAMR media [21]. Therefore, Nd-doped HAMR media has the potential to increase α , thereby improving switching probability and recording performance.

The magnetization dynamics of $L1_0$ -FePt has been measured by ferromagnetic resonance (FMR) and time-resolved magneto-optical Kerr effect (TRMOKE) [22–26]. Since $L1_0$ -FePt has high- K_u , the measurable temperature range using conventional FMR with microwaves is restricted to the high temperature range, 62 K to 22 K below T_C . At low temperatures, FMR field shifts to negative value due to large K_u , while at high temperatures close to T_C , the FMR signal becomes weak because of the reduced magnetization moment [22,23]. TRMOKE with a femtosecond laser is capable of magnetization dynamics up to the terahertz (THz) frequency range. TRMOKE covers the precession frequency of $L1_0$ -FePt over a wide temperature range from room temperature to near

T_C .

This study aims to clarify how Nd doping modulates α in $L1_0$ -FePt continuous thin films and its temperature dependence from room temperature to near T_C , which is critical for HAMR performance. We employ continuous thin films rather than granular thin films. Granular thin films, which consist of FePt grains and non-magnetic segregant materials, are more directly relevant to HAMR media applications. However, they introduce additional complexity in isolating the effect of Nd doping on FePt itself. In contrast, continuous thin films allow a more direct evaluation of the Nd doping effect on α . By sweeping temperature up to near T_C , we investigate α under conditions relevant to HAMR writing, and also take advantage of the high-temperature regime where extrinsic effects such as anisotropy dispersion are minimized, enabling a more intrinsic assessment of α . A previous study reported that α near T_C is comparable between granular and continuous FePt thin films, suggesting that the intrinsic α behavior observed in continuous films is largely preserved in granular structures [25]. Therefore, demonstrating the Nd doping effect on α in continuous films provides a meaningful feasibility study for its application to granular HAMR media structures.

In addition to magnetization dynamics measurements using TRMOKE, we investigated the microstructure and static magnetic properties to provide fundamental data on anisotropy and T_C . This comprehensive approach offers insights into how rare-earth element doping can optimize switching reliability and recording performance in HAMR.

2. Experimental method

Film samples of (FePt)_{100-x}(Nd)_x ($x = 0.0 - 4.1$ at.%) with thickness from 30 to 50 nm were deposited on an MgO (001) single-crystal substrate using the combinatorial high-throughput sputtering system [20]. Sputtering of (FePt)_{100-x}(Nd)_x was conducted by co-sputtering Fe, Pt, and Nd targets at a substrate temperature of 500°C and an Ar gas pressure of 10 mTorr. Nd concentration was tuned via the position-dependent compositional gradient and the Nd-target power. To reduce the Nd deposition rate, a mesh was placed over the Nd target. A carbon cap layer of 5 nm was deposited on the (FePt)_{100-x}(Nd)_x layer at room temperature.

Composition of Nd was analyzed by x-ray fluorescence (XRF) and the $L1_0$ -structure was evaluated by the superlattice peaks of x-ray diffraction (XRD) patterns. The thickness of the (FePt)_{100-x}(Nd)_x layer was evaluated by X-ray reflectivity (XRR) measurements. Microstructure and elemental analysis were conducted by scanning transmission electron microscopy (STEM) using a Spectra Ultra (Thermo Fisher Scientific). Three-dimensional atom probe tomography (APT) was carried out to evaluate atomic structure using LEAP5000XS (CAMECA) at the sample temperature of 30 K. A laser energy was set to 30 pJ. STEM and needle-shaped APT samples were fabricated by focused ion beam (FIB) with a scanning electron microscopy (SEM) dual-beam system Helios5UX (Thermo Fisher Scientific). A low accelerating voltage of 2 kV was selected in the final cleaning of the tip to reduce the FIB-damage in the sample [27]. Static magnetic properties, including T_C , were measured by using MPMS3, SQUID-VSM and PPMS Dynacool (Quantum Design). The temperature dependence of saturation magnetization ($\mu_0 M_s$) was performed to evaluate T_C with an analysis based on the Kuz'min formula [28].

Magnetization precessional motion was investigated by all-optical TRMOKE with a pump-probe method using a femtosecond laser system with a wavelength of 1028 nm, pulse width 290 fs, and repetition frequency 10 kHz. A probe beam with wavelength of 514 nm was frequency-doubled using a BBO crystal. An external magnetic field ($\mu_0 H_{\text{ext}}$) up to 7 T was applied using a superconducting magnet (Quantum Design) at a fixed angle of 80° with respect to the film normal. The sample was heated up to 450°C during TRMOKE measurements in a vacuum chamber with a base pressure of $\sim 10^{-4}$ Pa. Fig. 1 shows a schematic of sample heating and temperature measurement system for

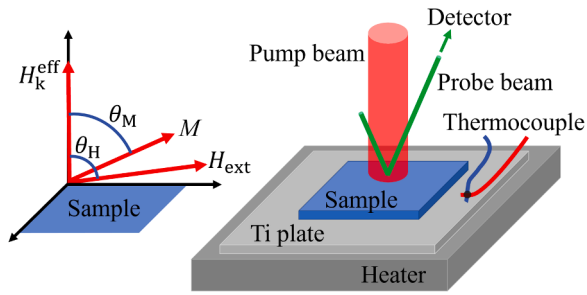


Fig. 1. Schematic of the sample heating and temperature measurement system in the TRMOKE experiments. A thermocouple was spot-welded onto the Ti plate, and $\mu_0 H_{\text{ext}}$ was applied at an angle of $\theta_H = 80^\circ$ during the measurements.

TRMOKE measurements. The heater was mounted onto the sample holder. A titanium (Ti) plate was affixed in direct contact with the heater, and the sample was placed on top of this Ti plate to ensure efficient thermal conduction. The temperature measured by a thermocouple spot-welded onto the Ti plate was used as the representative sample temperature throughout the experiments.

3. Results and discussion

3.1. Microstructure

Fig. 2(a) shows XRD patterns of $(\text{FePt})_{100-x}(\text{Nd})_x$ thin films with $x =$

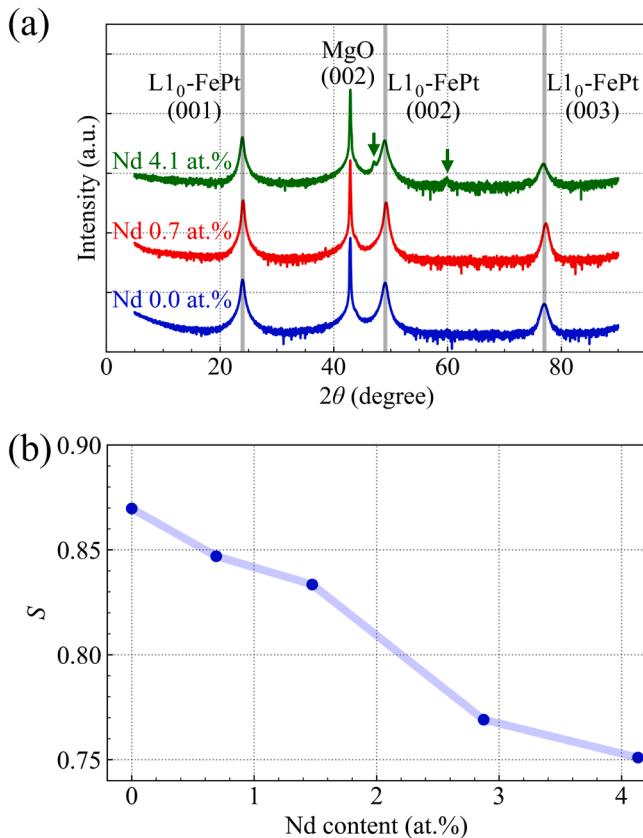


Fig. 2. (a) XRD patterns of $(\text{FePt})_{100-x}(\text{Nd})_x$ with $x = 0.0, 0.7,$ and 4.1 at.%. The presence of the superlattice (001) and (003) peaks and the fundamental (002) peak of L1_0 -FePt suggests chemical ordering and a (001) texture. The Nd 4.1 at. % sample exhibits additional peaks at 2θ values of 47.1° and 59.9° , indicated by arrows. The peak at a 2θ value of 47.1° corresponds to the L1_0 -FePt (110) reflection. (b) Nd content dependence of the L1_0 -ordering parameter (S). S decreases monotonically with increasing Nd content.

$0.0, 0.7,$ and 4.1 at.%. Both the superlattice (001) and (003) peaks and the fundamental (002) peak of L1_0 -FePt were observed in all samples, indicating chemical ordering and preferred (001) orientation. Fig. 2(b) shows Nd content dependence of L1_0 -ordering (S), calculated from the integrated intensity ratio of the (001) and (002) peaks [29–31]. S decreased with increasing Nd content. Moreover, additional peaks at 2θ values of 47.1° and 59.9° appeared in the Nd 4.1 at. % sample. The peak at 47.1° corresponds to the L1_0 -FePt (110) reflection.

Fig. 3(a)–(h) show high-angle annular dark field scanning TEM (HAADF-STEM) images and energy dispersive X-ray spectroscopy (EDS) elemental maps of $(\text{FePt})_{100-x}(\text{Nd})_x$ films with $x = 0.7$ and 4.1 at.%. For Nd 0.7 at.%, the sample exhibited Fe-deficient void regions within the otherwise continuous FePt layer, and Nd and Pt segregated into these voids. For Nd 4.1 at.%, pronounced Fe-deficient void regions induced striated segregation of Nd and Pt, which disrupted the continuous FePt layer and generated columnar, granular-like FePt structures.

To investigate whether Nd forms a solid-solution state within FePt grains, elemental analysis was performed using atom probe tomography (APT), which provides higher sensitivity and a lower detection limit of ~ 0.001 at.% compared to EDS, which has a detection limit of ~ 1 at.% [32]. Fig. 4 shows APT analysis of $(\text{FePt})_{99.3}(\text{Nd})_{0.7}$ thin film. Fig. 4(b) shows APT elemental map of Nd element. Nd was concentrated at the MgO/FePt interface and Nd segregation regions, as indicated by the arrows. Fig. 4(c) shows the mass spectrum of the signal from the whole sample and from the FePt region excluding the Nd segregation regions. While Nd peaks were detected in the spectrum from the whole sample, Nd peaks were absent in the spectrum from FePt region. This indicates that Nd was completely rejected from FePt.

The Nd 4.1 at. % sample exhibited more pronounced Fe-deficient void regions than the Nd 0.7 at. % sample, resulting in a granular-like microstructure of FePt due to the breakup of the continuous FePt layer. The appearance of the L1_0 -FePt (110) reflection peak in the Nd 4.1 at. % sample in XRD indicates the formation of poorly (001)-oriented grains. Such extensive segregation at high Nd content is considered to contribute to the decrease in S . The broad peak at a 2θ value of 59.9° may originate from a secondary phase related to an Nd-Pt compound, as these regions correspond to Fe-deficient void regions. The granular-like FePt structure induced by Nd and Pt segregation, combined with the reduction in L1_0 -ordering, is expected to influence magnetic properties such as anisotropy and T_C . These effects will be discussed in detail in the following section.

3.2. Magnetic properties

Figs. 5 and 6 show magnetization curves and temperature dependence of saturation magnetization ($\mu_0 M_s$) of $(\text{FePt})_{100-x}(\text{Nd})_x$ thin films with $x = 0.0, 0.7,$ and 4.1 at.%. The Nd 4.1 at. % sample exhibited high coercivity of 1.3 T, as a result of the microstructural change from a continuous FePt layer to a granular-like FePt structure.

Magnetization curves and temperature dependence of $\mu_0 M_s$ were utilized to calculate anisotropy field ($\mu_0 H_k$) and T_C , shown in Fig. 7(a) and (c), respectively. $\mu_0 H_k$ was determined from the in-plane field required to saturate magnetization in Fig. 5, while T_C was obtained by fitting the temperature dependence of $\mu_0 M_s$ in Fig. 6 using the Kuz'min formula [28].

Fig. 7 shows Nd content dependence of (a) $\mu_0 H_k$, (b) $\mu_0 M_s$, and (c) T_C . $\mu_0 H_k$ increased up to 0.7 at. % Nd, then decreased at higher Nd contents. Similarly, T_C exhibited a maximum at 1.5 at. % Nd. In the case of FePt doped with another rare-earth element, Er, ab initio calculations have predicted that the anisotropy energy is enhanced, which has been attributed to contributions from the large orbital moment and strong SOC associated with Er [19]. Experimental studies also reported an increase in anisotropy field with a small amount of Er doping [20]. These results agree with the increase of $\mu_0 H_k$ by Nd content until 0.7 at. % and suggest that the large orbital moment of Nd contributed to the anisotropy enhancement. Although Nd was rejected from FePt and segregated

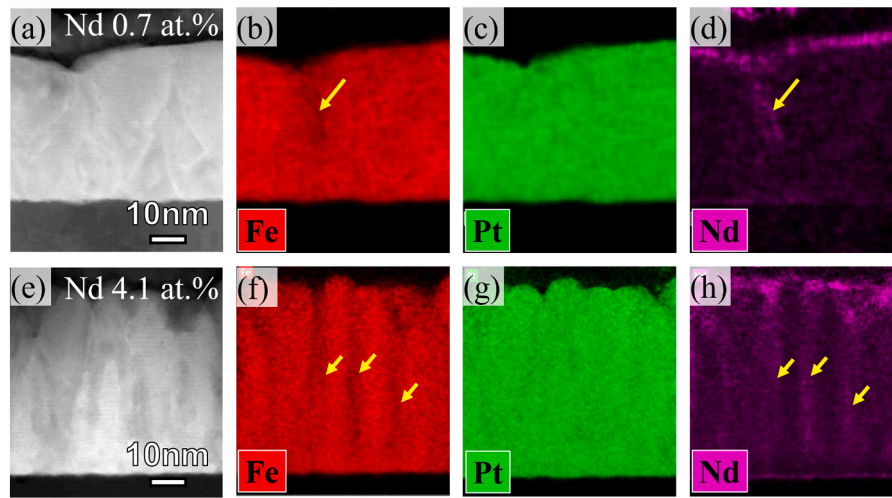


Fig. 3. (a) HAADF-STEM image of the $(\text{FePt})_{99.3}(\text{Nd})_{0.7}$ thin film. (b)-(d) Corresponding EDS elemental maps of Fe, Pt, and Nd for (a). (e) HAADF-STEM image of the $(\text{FePt})_{95.9}(\text{Nd})_{4.1}$ thin film. (f)-(h) Corresponding EDS elemental maps of Fe, Pt, and Nd for (e). Regions with weak Fe intensity (indicated by arrows), hereafter referred to as Fe-deficient voids, are observed in both the Nd 0.7 and 4.1 at.% samples. Nd and Pt segregate into these voids. In the Nd 4.1 at.% sample, the Fe-deficient voids become more pronounced, breaking up the continuous FePt layer and producing columnar, granular-like FePt structures. Nd also segregates along the MgO/FePt interface and on the surface.

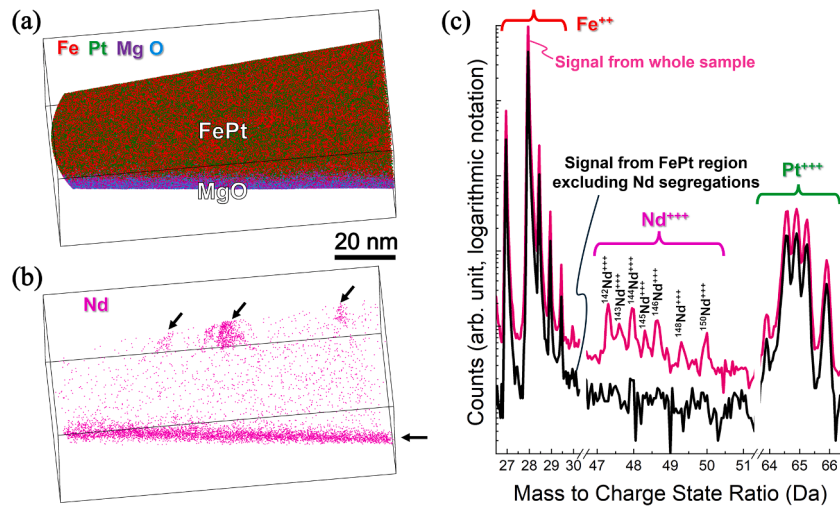


Fig. 4. APT map and mass spectra of the $(\text{FePt})_{99.3}(\text{Nd})_{0.7}$ thin film. (a) APT maps of Fe, Pt, Mg and O. (b) APT map of Nd. Nd is concentrated in regions corresponding to the MgO/FePt interface and Nd segregation regions, as indicated by the arrows. (c) Mass spectrum obtained from the whole sample, and the FePt region excluding the Nd-segregated areas. Nd peaks are detected in the spectrum from the whole sample (pink), whereas no Nd peaks are observed in the spectrum from the FePt region excluding the Nd-segregated areas (black), indicating that Nd is fully rejected from FePt.

into Fe-deficient void regions, as shown in Figs. 3 and 4, its large orbital moment could contribute to local anisotropy through SOC effects.

At Nd contents higher than 0.7 and 1.5 at.%, $\mu_0 H_k$ and T_C decreased, respectively. In contrast to $\mu_0 H_k$ and T_C , which exhibited a maximum at low Nd content, $\mu_0 M_s$ monotonically decreased with Nd content. Reduction of S is known to decrease $\mu_0 H_k$ and T_C [33]. Therefore, while the initial increase in $\mu_0 H_k$ and T_C at low Nd content is likely due to large orbital moment of Nd, their subsequent decrease at higher Nd content is primarily attributed to the decrease in S , rather than a direct Nd effect.

3.3. Magnetization dynamics

In HAMR, suppressing thermally driven back-switching is essential for improving recording reliability [6]. Increasing α is expected to reduce back-switching and enable faster magnetization reversal, which is required for achieving higher ADC [10–13,34]. Therefore, evaluating α is crucial for media design, especially in engineering approaches that

aim for higher α .

Magnetization dynamics was measured by TRMOKE in the temperature range from room temperature to near T_C . For each temperature, magnetization dynamics was measured under an external field ($\mu_0 H_{\text{ext}}$) from 5 to 7 T at a fixed angle of 80° from film normal. Fig. 8(a) shows TRMOKE signals of $(\text{FePt})_{99.3}(\text{Nd})_{0.7}$ thin film at 633 K under multiple $\mu_0 H_{\text{ext}}$. Incident of pump pulse to the film surface at delay time (t) of 0 ps induces ultrafast demagnetization and triggers magnetization precession. Precession motion is detected by Kerr rotation angle and appears as a sinusoidal oscillation during the magnetization recovery [35]. The signals were fitted using Eq. (2).

$$S(t) = A_0 + A_1 \exp\left(-\frac{t}{\tau_r}\right) + A_2 \exp\left(-\frac{t}{\tau_f}\right) \sin(2\pi f t + \varphi) \quad (2)$$

where, A_0 , A_1 and A_2 denote the amplitude coefficients of offset, magnetization recovery, and magnetization precession, respectively. τ_r , τ_f , f and φ denote the time constant of magnetization recovery, lifetime of

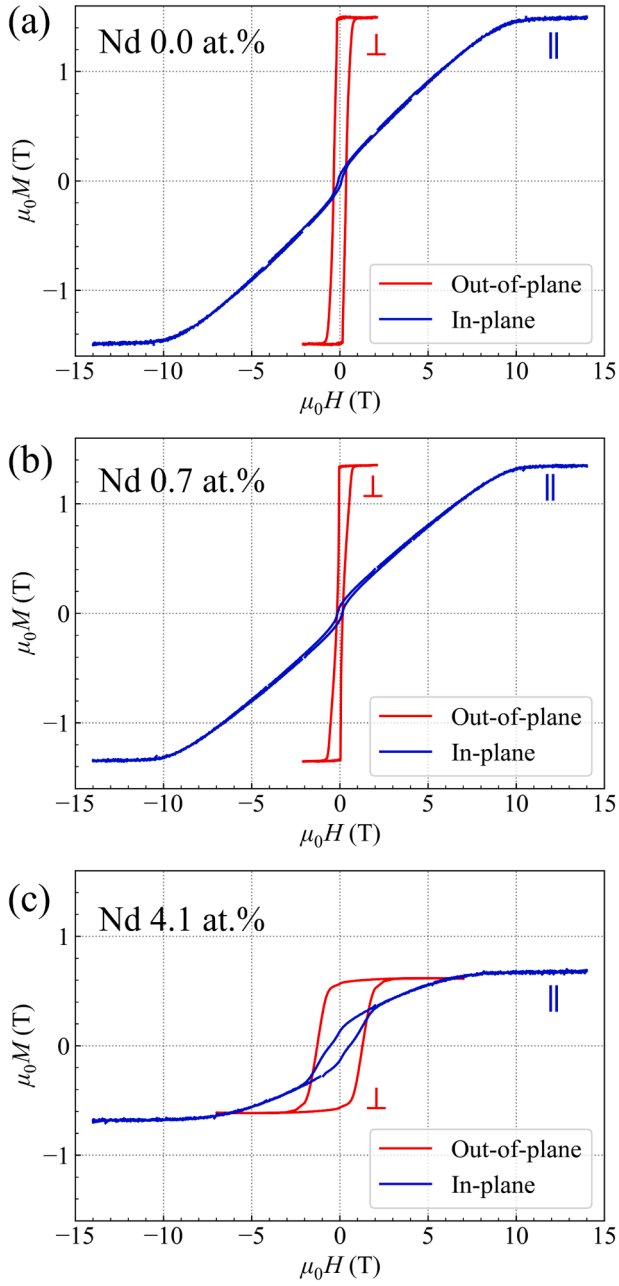


Fig. 5. Out-of-plane ($\perp$, shown in red) and in-plane ($\parallel$, shown in blue) magnetization curves of (FePt)_{100-x}(Nd)_x thin films with Nd contents of (a) 0.0 at.%, (b) 0.7 at.%, and (c) 4.1 at.%. The Nd 4.1 at.% sample exhibits a coercivity of 1.3 T, attributed to a change in microstructure from a continuous layer to a granular-like FePt structure.

magnetization precession, frequency of magnetization precession, and phase of magnetization precession, respectively. Fig. 8(b) shows the $\mu_0 H_{\text{ext}}$ dependence of f , which is expressed by Kittel's formula (3):

$$f = \frac{\mu_0 \gamma}{2\pi} \sqrt{H_1 H_2} \quad (3)$$

where, μ_0 , γ denote the permeability of vacuum and the gyromagnetic ratio, respectively. Terms of effective magnetic field, $\mu_0 H_1$ and $\mu_0 H_2$ are expressed by Eqs. (4) and (5),

$$\mu_0 H_1 = \mu_0 H_{\text{ext}} \cos(\theta_H - \theta_M) + \mu_0 H_k^{\text{eff}} \cos^2 \theta_M \quad (4)$$

$$\mu_0 H_2 = \mu_0 H_{\text{ext}} \cos(\theta_H - \theta_M) + \mu_0 H_k^{\text{eff}} \cos(2\theta_M) \quad (5)$$

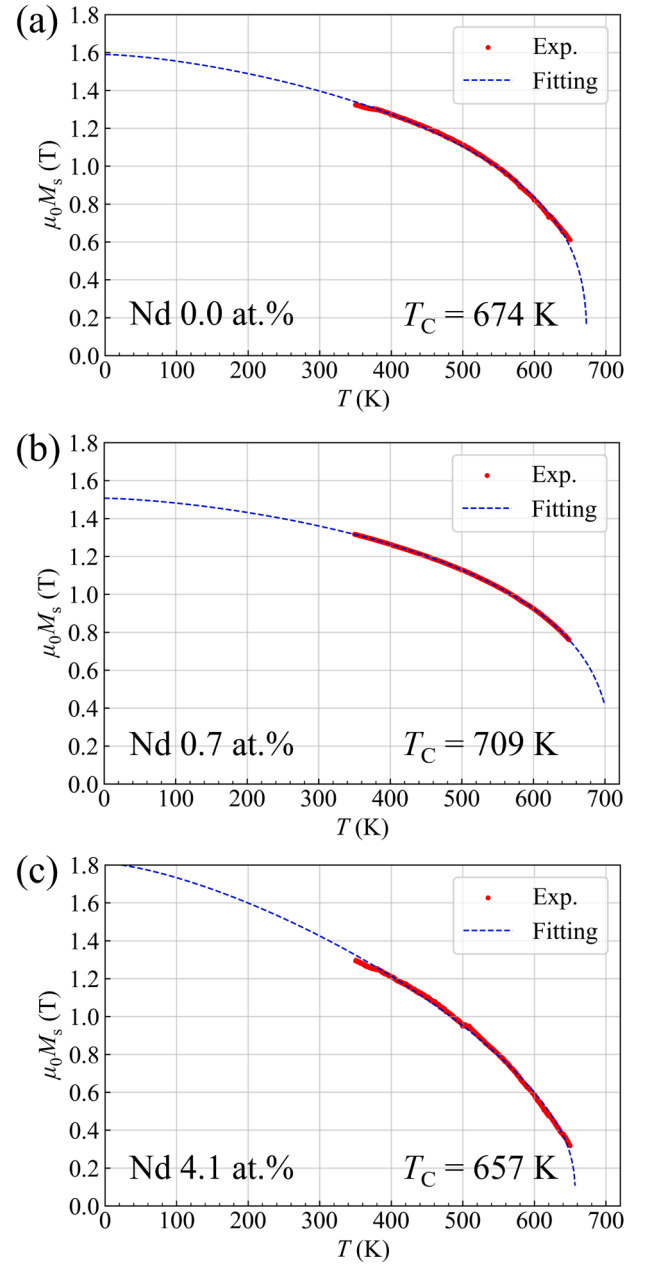


Fig. 6. Temperature dependence of $\mu_0 M_s$ of (FePt)_{100-x}(Nd)_x thin films with Nd contents of (a) 0.0 at.%, (b) 0.7 at.%, and (c) 4.1 at.%. T_C was obtained by fitting the experimental data (red) with the Kuz'min formula (blue dashed line) in the temperature range of 350–650 K [28].

where, θ_H , θ_M and $\mu_0 H_k^{\text{eff}}$ denote the angle of applied field from film normal, the angle of magnetization from film normal, and the effective anisotropy field. θ_H of 80° was applied for all measurements. Total energy (E) consists of Zeeman energy and uniaxial anisotropy energy as described in Eq. (6).

$$E = -\mu_0 M H_{\text{ext}} \cos(\theta_H - \theta_M) + \frac{\mu_0 M H_k^{\text{eff}}}{2} \sin^2 \theta_M \quad (6)$$

θ_M satisfies the energy equilibrium condition in Eq. (7):

$$\frac{dE}{d\theta_M} = \mu_0 M H_{\text{ext}} \sin(\theta_H - \theta_M) + \frac{\mu_0 M H_k^{\text{eff}}}{2} \sin(2\theta_M) = 0 \quad (7)$$

Fig. 8(b) shows the $\mu_0 H_{\text{ext}}$ dependence of f . $\mu_0 H_k^{\text{eff}}$ and γ were calculated by fitting that satisfies Eq. (3)–(7). $\mu_0 H_k^{\text{eff}}$ and g factor

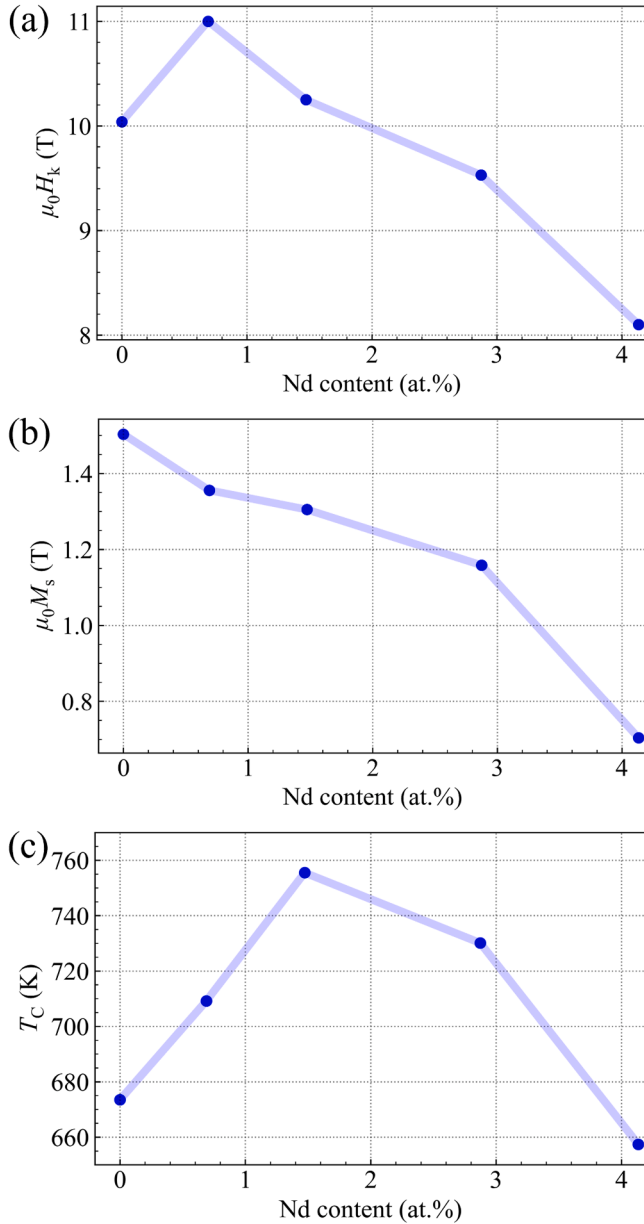


Fig. 7. Nd content dependence of (a) $\mu_0 H_k$, (b) $\mu_0 M_s$, and (c) T_c for $(\text{FePt})_{100-x}(\text{Nd})_x$ thin films. $\mu_0 H_k$ is determined from the in-plane field required to saturate magnetization in Fig. 5. $\mu_0 M_s$ is obtained from both out-of-plane and in-plane magnetization curves in Fig. 5. $\mu_0 H_k$ and T_c increase with Nd content up to 0.7 and 1.5 at.%, respectively, and then decrease at higher Nd contents. In contrast, $\mu_0 M_s$ decreases monotonically.

calculated from Fig. 8(b) were 3.94 T and 2.05, respectively. Fig. 8(c) shows the $\mu_0 H_{\text{ext}}$ dependence of effective damping constant (α_{eff}). α_{eff} is calculated using f and τ by Eq. (8),

$$\frac{1}{\tau} = 2\pi\alpha_{\text{eff}}f \quad (8)$$

α_{eff} decreased with increasing $\mu_0 H_{\text{ext}}$ (Fig. 8(c)). The variation of α_{eff} originates from the combined effect of intrinsic Gilbert damping and extrinsic contributions such as anisotropy distribution [22,23,25]. To minimize extrinsic effects, the value at $\mu_0 H_{\text{ext}}$ of 7 T was taken as the representative α_{eff} at each temperature hereafter.

Fig. 9(a) shows magnetization dynamics of $(\text{FePt})_{99.3}(\text{Nd})_{0.7}$ film measured at various temperatures at $\mu_0 H_{\text{ext}}$ of 7 T. Using signals measured in the range of $\mu_0 H_{\text{ext}}$ from 5 to 7 T for each temperature,

$\mu_0 H_k^{\text{eff}}$ was determined by fitting. Fig. 9(b) shows the temperature dependence of $\mu_0 H_k^{\text{eff}}$ for $(\text{FePt})_{100-x}(\text{Nd})_x$ thin films with $x = 0.0, 0.7$, and 4.1 at.%. $\mu_0 H_k^{\text{eff}}$ decreased with increasing temperature. Fig. 9(c) shows temperature dependence of α_{eff} . α_{eff} decreased with increasing temperature at first, then increased near T_c , where T_c in Fig. 9(c) was measured by temperature dependence of $\mu_0 M_s$ in Fig. 6. Although magnetization dynamics above T_c is generally considered unobservable due to disappearance of spontaneous magnetization, a TRMOKE signal was detected slightly above T_c . This discrepancy may be attributed to the setup difference.

As reported in previous studies on the temperature dependence of the magnetic damping constant for L1₀-FePt nano-granular thin films, α_{eff} exhibited a local minimum in the temperature range from 570 to 660 K, as shown in Fig. 9(c) [25]. α_{eff} included both intrinsic Gilbert damping and extrinsic contributions, which strongly depended on the magnetic field and temperature as reported in previous studies [22,23,25,36,37]. At high temperatures approaching T_c , intrinsic damping increases due to the reduction of magnetization and the associated enhancement of thermally driven spin fluctuations, which leads to stronger longitudinal relaxation in the Landau–Lifshitz–Bloch (LLB) model [37]. In addition to this LLB-type mechanism, previous studies have also reported that the increased magnon population at elevated temperatures enhances magnon-related scattering processes, further increasing the intrinsic damping component [23,38].

In contrast, extrinsic damping arising from magnetic inhomogeneity such as anisotropy distributions and voids in continuous FePt layer, which were observed in Nd-doped FePt samples, is suppressed at elevated temperatures. This is because the absolute magnitude of anisotropy distribution decreases as anisotropy itself diminishes, reducing the scattering strength associated with magnetic inhomogeneity [22,23]. As a result, extrinsic damping dominates near room temperature, whereas intrinsic damping becomes dominant at elevated temperatures, and the observed temperature dependence of α_{eff} can be interpreted as the combined effect of these opposing trends.

Fig. 10 shows Nd content dependence of α_{eff} near T_c . T_c of each sample was obtained from temperature dependence of $\mu_0 M_s$ in Fig. 6. Since T_c and measurement temperature were not exactly matched, α_{eff} near T_c was obtained by linear interpolation between two nearest measurement points. α_{eff} increased from 0.048 without Nd doping to 0.066 with Nd 4.1 at.% doping, providing the first experimental demonstration of α_{eff} enhancement through rare-earth element doping in FePt. Since Nd was rejected from FePt, this enhancement is attributed to localized effects at FePt/Nd interfaces.

The interface is an important factor for engineering α [26,39–44]. For example, it has been reported that lattice mismatch and subsequent misfit dislocation at the interface increase α due to electron scattering [26]. Ferromagnet/rare-earth interface effects on α have also been reported in multilayer systems with NiFe/rare-earth interfaces [39–43] and CoFe/rare-earth interfaces [44]. In those systems, the enhancement of α saturates with increasing rare-earth layer thickness, confirming its interface origin [39,40]. Several mechanisms have been discussed, including spin pumping at the ferromagnet/rare-earth interface [41], antiferromagnetic coupling between ferromagnet and rare-earth layers [42–44], and enhanced orbital moments of Co and Ni through antiferromagnetic coupling with the rare-earth layer [44].

Notably, a larger enhancement of α is observed in NiFe/Nd systems than in NiFe/Gd due to the larger orbital moment of Nd, consistent with trends in bulk NiFe–Nd and NiFe–Gd alloys [39]. Similarly, the enhancement of α in NiFe adjacent to Tb was more pronounced than that in CoFe adjacent to Gd, reflecting the larger orbital moment of Tb relative to Gd [44]. This suggests that in the present Nd-doped FePt films, where Nd was rejected from FePt, the large orbital moment and strong SOC of Nd remain effective in enhancing α even when the Nd effect is confined to the interface.

This interpretation is further consistent with the observed increase in $\mu_0 H_k$ at low Nd concentrations (up to 0.7 at.%), as enhanced SOC at the

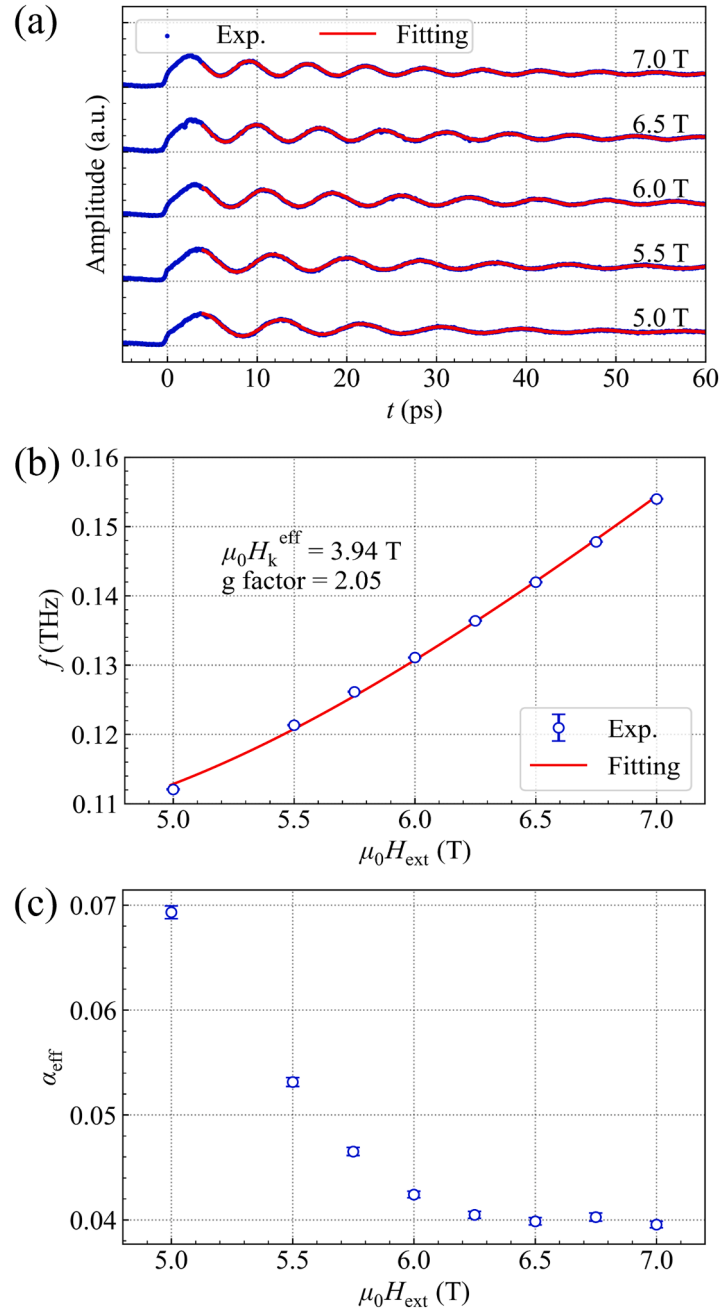


Fig. 8. Magnetization dynamics of the (FePt)_{99.3}(Nd)_{0.7} thin film at 633 K measured by TRMOKE. (a) TRMOKE signals under $\mu_0 H_{\text{ext}}$ from 5.0 to 7.0 T. Blue symbols represent experimental data points, and red lines represent fitting results. (b) $\mu_0 H_{\text{ext}}$ dependence of f with fitting. Blue symbols represent experimental data, and the red line represents the fitting curve. $\mu_0 H_k^{\text{eff}}$ and g factor were obtained by fitting. (c) $\mu_0 H_{\text{ext}}$ dependence of α_{eff} (blue symbols: experimental data). α_{eff} decreases and eventually saturates at high $\mu_0 H_{\text{ext}}$.

interface is expected to locally modify the magnetic anisotropy. Taken together, the increases in both α_{eff} and $\mu_0 H_k$ present a consistent picture of SOC enhancement at FePt/Nd interfaces.

In terms of two-magnon scattering, non-uniform microstructures are known to enhance α_{eff} [45]. Such damping enhancement, which originates from the extrinsic contributions, can be suppressed near T_C [22]. Therefore, the impact of microstructure on α_{eff} is likely negligible in Fig. 10.

From an application perspective, α_{eff} reaching 0.066 at 4.1 at.% Nd is notable. HAMR modeling predicts that $\alpha > 0.1$ significantly improves switching probability and the signal-to-noise ratio (SNR) under conditions of pulsed laser irradiation [12,14]. Conventionally HAMR drives utilize continuous-wave (CW) laser irradiation, where the available

switching time is longer than in pulsed laser irradiation [46]. Therefore, the required α for CW may be lower than 0.1. However, achieving high α remains critical because increasing thermal gradients for higher ADC reduces the available recording time.

Increasing α in granular FePt films, which are utilized in actual HAMR media, is critical for improving HAMR performance. The observed enhancement of α in continuous films originates from Nd segregation formed within FePt. In granular HAMR media, FePt grains form interfaces with surrounding non-magnetic material such as BN and C at grain boundaries. Through these interfaces, Nd is expected to interact with FePt, giving rise to similar interface effects on α as those observed in the continuous films. However, the spatial distribution and local concentration of Nd at FePt grain boundaries in granular films may

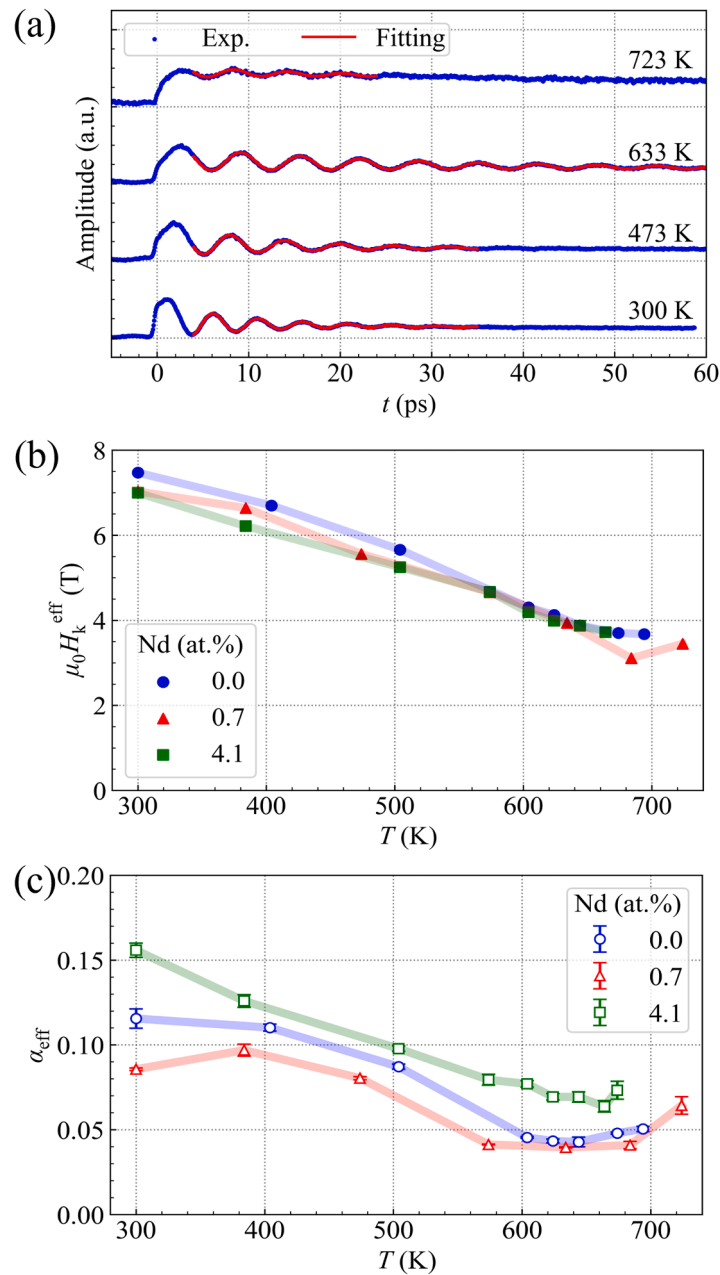


Fig. 9. Temperature dependence of magnetization dynamics. (a) TRMOKE signals of the $(\text{FePt})_{99.3}(\text{Nd})_{0.7}$ thin film at multiple temperatures under $\mu_0 H_{\text{ext}}$ of 7 T. (b) Temperature dependence of $\mu_0 H_k^{\text{eff}}$ of $(\text{FePt})_{100-x}(\text{Nd})_x$ thin films with $x = 0.0, 0.7$, and 4.1 at.%. $\mu_0 H_k^{\text{eff}}$ decreases with increasing temperature. (c) Temperature dependence of α_{eff} . T_C values in the labels were measured in Fig. 6.

influence the degree of α enhancement. Further work should therefore focus on quantifying this effect in granular films.

This study demonstrated that rare-earth element doping is effective in enhancing α near T_C even in FePt, which already exhibits strong SOC due to the heavy metal element Pt. These findings highlight the feasibility of controlling α through rare-earth element selection in FePt-based HAMR media. Since Nd did not dissolve into FePt, exploring other rare-earth elements or heavy elements that can form a solid solution with FePt presents a promising direction for designing media optimized for reliable recording.

4. Conclusion

Nd-doped L1₀-FePt continuous thin films were investigated in terms of microstructure, magnetic properties and magnetization dynamics.

L1₀-chemical ordering and (001)-orientation were maintained in the range of Nd content from 0.0 to 4.1 at.% Nd. Nd was completely rejected from FePt, and Nd and Pt segregated into Fe-deficient void regions. Although the Nd effect is limited to the interface between FePt and Nd, $\mu_0 H_k$ was enhanced by doping a small amount of Nd through local SOC enhancement.

Similarly, the increase in α by Nd doping was successfully confirmed near T_C . α_{eff} increased from 0.048 without Nd doping to 0.066 with Nd 4.1 at.% doping, an enhancement attributed to localized effects at the interface between FePt and Nd. This study shows the effectiveness of rare-earth element doping in L1₀-FePt, which already has strong SOC, and the possibility of designing α in HAMR media by doping rare-earth elements. Although Nd did not dissolve into FePt grains, another possible approach is to explore rare-earth elements that can form a solid solution with FePt.

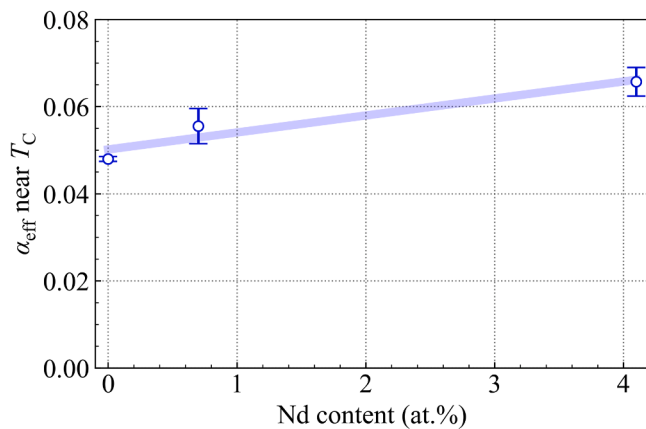


Fig. 10. Nd content dependence of α_{eff} near T_C . T_C was obtained from Fig. 6. α_{eff} near T_C was obtained by linear interpolation between the two nearest measurement points. Error bars were estimated by linearly interpolating the reported standard errors at the two measurement points used for the interpolation of α_{eff} at T_C . T_C was obtained from temperature-dependent $\mu_0 M_s$ shown in Fig. 9. α_{eff} at T_C increases with Nd content.

CRedit authorship contribution statement

Aiko Sakoguchi: Writing – original draft, Investigation, Formal analysis, Conceptualization. **Yuta Sasaki:** Writing – review & editing, Supervision, Resources, Methodology, Funding acquisition, Conceptualization. **Daisuke Ogawa:** Investigation, Formal analysis. **Jun Uzuhashi:** Writing – review & editing, Investigation, Formal analysis. **Hirofumi Suto:** Resources, Methodology, Funding acquisition. **Yukiko K. Takahashi:** Writing – review & editing, Supervision, Investigation, Funding acquisition, Formal analysis, Conceptualization.

Declaration of competing interest

The authors declare the following financial interests/personal relationships which may be considered as potential competing interests:

Yukiko K. Takahashi reports financial support was provided by JST CREST. Yukiko K. Takahashi reports financial support was provided by JSPS KAKENHI. Hirofumi Suto, Yukiko K. Takahashi reports financial support was provided by Advanced Storage Research Consortium (ASRC). Yuta Sasaki reports financial support was provided by MEXT-LEADER. Aiko Sakoguchi, Yuta Sasaki, Daisuke Ogawa, Jun Uzuhashi, Yukiko K. Takahashi has patent MAGNETIC RECORDING MEDIA WITH RARE-EARTH—CONTAINING SEGREGANT AND METHODS OF FABRICATING SAME pending to Western Digital Technologies, Inc., National Institute for Materials Science. If there are other authors, they declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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