Correlation of magnetic moments and local structure of FePt nanoparticles

C. Antoniak, M. Spasova, A. Trunova, K. Fauth, M. Farle, and H. Wende

1 Fachbereich Physik and Center for Nanointegration Duisburg-Essen (CeNIDE), Universität Duisburg-Essen, Lotharstr. 1, D-47048 Duisburg (Germany)
2 Experimentelle Physik IV, Universität Würzburg, Am Hubland, D-97074 Würzburg, Germany

E-mail: carolin.antoniak@uni-due.de

Abstract. The influence of structural and compositional changes within FePt nanoparticles on their magnetic properties was studied by means of x-ray absorption spectroscopy in the near-edge regime and its associated magnetic circular dichroism as well as by analysis of the extended x-ray absorption fine structure. The magnetic moments at the Fe sites were found to be a sensitive monitor to changes of the local surrounding: While compositional inhomogeneities in the nanoparticles yield significantly reduced magnetic moments (by 20–30%) with respect to the corresponding bulk material, thermally induced changes in the crystal structure yields strongly enhanced orbital contributions (up to 9% of the spin magnetic moment). Also the break of crystal symmetry at the surface leads to an enhanced orbital magnetism which was confirmed by determination of the ratio of orbital-to-spin magnetic moment for FePt particles with different sizes between 3 and 6 nm in diameter.

1. Introduction

Driven by their potential use as new ultra-high density storage media, nanoparticles of ferromagnetic alloys have been the subject of intense research activities over the last decade (see e.g. [1–6]). Since FePt in the chemically ordered state is one of the materials with the highest magnetocrystalline anisotropy which is around 6 × 10^6 J/m^3 [7–10], it is a promising candidate to overcome the so-called superparamagnetic limit, i.e. data loss due to thermally activated fluctuations of magnetisation.

The high-temperature part of the phase diagram of the bulk Fe_xPt_{1-x} system is shown in Fig. 1. A detailed description can be found e.g. in [11, 12]. In thermodynamical equilibrium, there are three chemically ordered phases: For Fe contents around 25at% and 75at% it crystallises in the fcc structure with an L1_2 crystal symmetry that is schematically drawn in Fig. 1. Around the equiatomic composition the chemically ordered alloy consists of atomic layers of Fe and Pt atoms along the [100] direction. This structure leads to a tetragonal distortion along the stacking direction (L1_0 crystal symmetry). In the as-prepared state of nanoparticles and thin films of FePt the formation of the chemically ordered state is kinetically suppressed but can be obtained by enhancing the volume diffusion by annealing. Depending on the composition the magnetic properties of Fe_xPt_{1-x} alloys are very different: For example while alloys around the equiatomic composition exhibit a ferromagnetic ordering, FePt_3 is an antiferromagnet in the...
chemically ordered state. In this work, we focus on the structural and magnetic properties of spherical Fe$_x$Pt$_{1-x}$ nanoparticles with Fe contents of about 50at% and diameters between 3.4 nm and 6.3 nm. Various aspects of the x-ray absorption spectroscopy are used to answer the following questions:

- How do changes in the crystal symmetry due to thermally induced chemical ordering influence the element-specific magnetic moments of FePt nanoparticles?
- What happens to the magnetic moments at the surface of nanoparticles?
- Is there a measurable correlation between the local composition and the magnetic properties of FePt nanoparticles?

We present the results of a detailed structural characterisation by analysis of the extended x-ray absorption fine structure (EXAFS) of pure metallic FePt nanoparticles supported by investigations employing high-resolution transmission electron microscopy (HR-TEM), electron diffraction (ED) and x-ray diffraction (XRD) before turning to the discussion of the magnetic properties focussing on the element-specific magnetic moments at both the Fe and Pt sites determined by means of the x-ray magnetic circular dichroism (XMCD).

2. Nanoparticle synthesis and sample preparation

The Fe$_x$Pt$_{1-x}$ nanoparticles were synthesised using the wet-chemical approach by Sun et al. [13] based on the reduction of a Pt-containing salt and thermal decomposition of Fe(CO)$_5$ in an organic solvent. To stabilise the particles in dispersion, oleic acid and oleylamine are added that form a shell around the particles and prevent agglomeration. A submonolayer of particles is deposited onto a naturally oxidised Si wafer using the spin-coating technique. For all samples, the size distribution is log-normal with a standard deviation of about 10–15%. HR-TEM studies reveal the single-crystalline fcc structure of the nanoparticles [15]. In Fig. 4 (b) an example of FePt nanoparticles with a mean diameter of 6.3 nm deposited onto the Si substrate is shown. In the as-prepared state, the nanoparticles are not only surrounded by the organic ligands, but are also partially oxidised. Therefore, a soft in situ hydrogen plasma treatment [14] was used to remove all ligands and to reduce the Fe oxides. The efficiency of the plasma cleaning was checked.

![Figure 1. Phase diagram of bulk Fe$_x$Pt$_{1-x}$. The focus in this work is on alloys around the equiatomic composition (grey coloured area).](image-url)
Figure 2. XANES at the carbon K edge (left panel) and Fe L\textsubscript{3,2} absorption edges (right panel) of 6 nm FePt nanoparticles [15]. Dashed lines correspond to the particles in the as-prepared state, solid lines to hydrogen plasma cleaned particles.

by recording the x-ray absorption near-edge structure (XANES) as can be seen in Fig. 2. The as-prepared nanoparticles show a strong absorption signal at the carbon K edge typical for oleic acid and oleylamine. At the Fe L\textsubscript{3,2} absorption edges, a clear multiplet structure indicates the presence of Fe oxides. After the plasma treatment, no absorption signal at the carbon K edge is obtained, and the multiplet structure at the Fe L\textsubscript{3,2} edges vanished confirming the reduction of all oxides. Thus, the in situ hydrogen plasma treatment gives us the possibility to study the properties of pure metallic Fe\textsubscript{x}Pt\textsubscript{1−x} nanoparticles [15].

In order to obtain the chemically ordered L\textsubscript{10} phase, a post-deposition thermal treatment is necessary. A well-known problem is sintering of the nanoparticles due to the enhanced mobility on the substrate at elevated temperatures. Several approaches to prevent agglomeration during annealing have been suggested in the literature, e.g. a strong bonding to the substrate via functional amino-silane molecules [16–18] or embedding in a NaCl matrix [19, 20]. However, since this may influence the electronic structure and magnetic properties, we present results on the largest nanoparticles, i.e. the particles with the lowest mobility on the substrate without any surface modifications. For these particles with a mean diameter of 6.3 nm scanning electron microscopy (SEM) images show an agglomeration of less than 20% of all particles after annealing at 600°C for 30 min. This leads to almost no change of the mean diameter which is 6.4 nm after annealing, but the standard deviation increases from \( \sigma = 0.14 \) to 0.25.

3. Crystal structure and alloying

By XRD and HR-TEM and ED investigations of FePt nanoparticles with a diameter around 4.4 nm, a lattice expansion of 1–2% compared to the corresponding bulk material was obtained [15]. The lattice constant of larger particles (d = 6.3 nm) was found to be in agreement to the values of the bulk material. However, these results were obtained on nanoparticles in the as-prepared state and thus the lattice expansion may be related to the influence of Fe oxides at the surface or the organic ligands surrounding the particles. Therefore, the lattice constant was determined by the analysis of the EXAFS at the Pt L\textsubscript{3} and Fe K edges of bulk material, as-prepared FePt nanoparticles and plasma cleaned FePt nanoparticles. As reported elsewhere [21], the lattice expansion could be confirmed for the case of the plasma cleaned particles indicating that this is
an intrinsic property of the FePt nanoparticles with diameters below 6 nm and is not caused by surface modifications.

In addition, a clear deviation of the composition around the probe atoms was found compared to the averaged value determined by energy-dispersive x-ray spectroscopy (EDS) in the TEM. This analysis of the local structure was possible since the backscattering amplitudes for Fe and Pt differ significantly as shown in the left panel of Fig. 3. Whereas in the case of Fe backscatterer, the amplitude shows a maximum at a wavenumber \( k \approx 60/\text{nm} \), Pt has its maximum backscattering amplitude at much higher wavenumbers \( k \approx 150/\text{nm} \). At the \( k \) position of maximum backscattering amplitude of Fe, Pt exhibits a local minimum of the amplitude. The strong reduction in the Pt amplitude over a small range at this point is connected to a more rapidly changing phase. This effect is known in literature as generalised Ramsauer-Townsend effect \([22, 23]\). In a simple picture, the wavelength of the outgoing photoelectron (about 0.1 nm for \( k \approx 60/\text{nm} \)) is well-matched to the size of the scatterer. In this case, the photoelectron may tunnel through the scattering potential and the scattering cross-section vanishes leading to a dip in the backscattering amplitude at a fairly distinct wavenumber. In the case of a Pt-rich environment, this yields a reduced amplitude of EXAFS oscillations in this range of wavenumbers. Thus, by standard Fourier based analysis employing the FEFF code \([25-30]\), the wavenumber dependent amplitude of simulated EXAFS oscillations can be fitted to experimental data only by changing the local composition around the absorber atom. For the FePt nanoparticles \((40 \pm 8)\text{at}\% \text{Fe}\) has to be assumed for a proper simulation of the experimental data measured at the Pt L\(_3\) edge which does not match the value found by EDS \((56 \pm 5)\text{at}\% \text{Fe}\). This difference is not a conflict between EXAFS and EDS results, but simply reflects that an averaging technique like EDS does not allow for the detection of an inhomogeneous composition of the investigated sample whereas the EXAFS technique does. The reason is that EXAFS stems from scattering of the photoelectron at the local surrounding of the probe atoms. Thus, it can be concluded that the Pt absorbing atoms are in a Pt-rich environment and the Fe atoms are in an Fe-rich environment. From the Fe content around the Pt atoms and the averaged value, the Fe content in the near environment of the Fe atoms

![Figure 3.](image)

**Figure 3.** Backscattering amplitude for Fe and Pt backscatterers (left panel) and 3D surface plot of the difference between wavelet transformed EXAFS of Pt absorber atoms in nanoparticles and bulk material (right panel) \([24]\). The difference reveals the change in the surrounding of Pt sites. The positions of the Fe and Pt nearest neighbour (nn.) atom backscattering maxima are marked.
Figure 4. (a) Exemplary XANES and XMCD spectra of plasma cleaned FePt nanoparticles with a mean diameter around 6.3nm measured at the Fe and Pt L_{3,2} absorption edges at low temperatures in maximum external field applied (see text for detailed parameters). (b) Scanning electron microscopy image of the examined sample showing a submonolayer of nanoparticles on a naturally oxidised Si wafer.

is expected to be around 72\text{at}\%. The analysis of the experimental data EXAFS measured at the Fe K edge support this conclusion, i.e. the Fe probe atoms are in an Fe-rich environment containing (70 \pm 12)\text{at}\% Fe. From this analysis, we concluded that Fe has to be in an Fe-rich environment and Pt is in a Pt-rich environment [24].

A method to visualise this result is the use of wavelet transformation instead of Fourier transformation. The main idea behind the wavelet transformation is to replace the infinitely expanded periodic oscillations in a Fourier Transformation by located wavelets as kernel for the integral transformation yielding transformed data not only as a function of radial distance but also as a function of wavenumber in the case of EXAFS data. The \( k \) position of maxima in the wavelet transformed signal is connected to the different elements via the individual \( k \) position of their maximum backscattering amplitude. Therefore it can be distinguished between the contribution of the Fe and Pt backscatters in Fe\textsubscript{2}Pt\textsubscript{1-x} alloys: A higher Fe content leads to a higher amplitude of the transformed data at \( k \approx 60/\text{nm} \), whereas a higher Pt content is connected to a higher amplitude at \( k \approx 150/\text{nm} \). Since the difference in local and averaged composition is rather small, in Fig. 3 the difference between the wavelet transformed experimental data measured on FePt nanoparticles at the Pt L\textsubscript{3} absorption edge and measured on bulk material with the same averaged composition is shown. A clear minimum in this difference at \( k \approx 60/\text{nm} \) indicates less Fe nearest neighbours around the Pt probe atom in the nanoparticles with respect to the bulk material. This is connected to an increasing amplitude at high wavenumbers indicating more Pt nearest neighbours around the Pt probe atom in the nanoparticles. Measurements at the Fe K edge [24] confirm this result.

4. Magnetic properties determined by x-ray absorption

Element-specific magnetic moments were determined by means of XMCD at the Fe and Pt sites. Measurements in the soft x-ray regime at the Fe L_{3,2} absorption edges were performed at the bending magnet beamline PM3 at BESSY II, Berlin (Germany) in magnetic fields of up to \( \mu_0 H_{\text{ext}} = 3 \text{T} \). After each scan, the magnetic field was reversed while the helicity of
the incident photons was kept constant. For verification, some experiments were repeated with reversed helicity as well. All spectra were recorded in the total electron yield (TEY) mode by measurement of the sample drain current. Spectra at the Pt L\textsubscript{3,2} absorption edges in the hard x-ray regime were taken at the undulator beamline ID-12 at the ESRF, Grenoble (France) in maximum magnetic fields of \(\mu_0 H_{ext} = 0.6\) T in fluorescence yield (FY) mode. After each scan, either the magnetic field or the helicity was reversed.

In Fig. 4, an example of the experimental XANES data and its associated XMCD is shown for both Fe and Pt L\textsubscript{3,2} absorption edges. To separate the transitions into the 3d states of the Fe atoms and into the 5d final states of the Pt atoms from transitions into higher levels (or into the continuum), a two-step like function was subtracted in the case of the XANES at the Fe L\textsubscript{3,2} edges. Since the absorption at the Pt L\textsubscript{3,2} edges is not well-pronounced, this procedure would lead to a large error. Therefore, reference spectra of Au are shifted in energy and subtracted instead [31] after stretching the energy scale to account for the different lattice constants of Au and FePt. From the XANES and XMCD spectra, the effective spin magnetic moment \(\mu_S^{\text{eff}}\) and the orbital magnetic moment \(\mu_l\) can be determined according to the sum rules [32–34]. For the numbers of unoccupied final states 3.41 at the Fe sites and 1.74 at the Pt sites obtained from band structure calculations [24, 35] were used. Note, that the effective spin magnetic moment \(\mu_S^{\text{eff}} = \mu_S + 7\mu_t\) consists of the spin magnetic moment and a magnetic dipole moment \(\mu_t\) accounting for a possible asphericity of the spin density distribution.

Absorption spectra recorded in the TEY mode are influenced by saturation effects [36] that lead to an underestimation of the magnetic moments also in the case of nanoparticles [37]. The values discussed in this work are all corrected for these effects. The most significant corrections were made in the case of the largest particles: For an ensemble of spherical FePt nanoparticles with a diameter of 6 nm, the effective spin magnetic moments were corrected by 2% and the orbital magnetic moments by about 20%.

4.1. Influence of chemical order

One hint for a successful transformation into a (partial) L\textsubscript{10} phase is an enhanced coercive field since the magnetocrystalline anisotropy is expected to increase by about one order of magnitude.

![Figure 5](image-url)

Figure 5. Field-dependent magnetisation measured by XMCD at the Fe L\textsubscript{3} edge at 15 K for FePt nanoparticles in the chemically disordered state (left) and after annealing in the chemically ordered state (right) [38]. Note the different scaling of the abscissa.
Figure 6. Field-dependent magnetisation measured by XMCD at the Fe L\textsubscript{3} edge at different temperatures for FePt nanoparticles in the chemically ordered state (left). Temperature dependence of the coercive field (right) as obtained from experiment (symbols) and by simulation (solid line) [15]. The dashed line is a guide-to-the-eye.

compared to the chemically disordered state. Indeed, we found an increase from (36 ± 5) mT for 6.3 nm FePt particles at \( T = 15 \) K and the external magnetic field applied perpendicular to the sample plane to (292 ± 8) mT after annealing. The field-dependent magnetisation shown in Fig. 5 was measured at the L\textsubscript{3} absorption edge of Fe for two different angles \( \theta \) between external magnetic field and sample normal.

Before annealing, a clear angular dependence of the ratio of remanence-to-saturation magnetisation \( M_r/M_S \) is visible indicating magnetic dipole-dipole interactions between the particles. This leads to a hard direction of magnetisation parallel to the sample normal (\( \theta = 0 \)). After annealing, \( M_r/M_S \) is independent of the angle of the external magnetic field and equals 0.5 as expected within the Stoner-Wohlfarth model for non-interacting particles with uniaxial anisotropy and randomly distributed easy axes of magnetisation. This shows that the interparticle interactions become negligible with respect to the intra-atomic anisotropy of the annealed particles. Note, that a small angular dependence of the coercive field suggest a slight preferential order of easy axes that may be substrate induced.

After annealing, a clear hysteresis of the field-dependent magnetisation is visible even at 300 K as can be seen in Fig. 6. However, the wasp-waist shape of the hysteresis loop indicates a mixture of ferromagnetically blocked and superparamagnetic particles which means that the mean blocking temperature in this experiment, i.e the blocking temperature of particles with the mean diameter, is below 300 K.

The mean blocking temperature and the anisotropy can be quantified by analysis of the temperature-dependent coercive field according to [39]

\[
H_c(T) \approx \frac{K_{\text{eff}}}{M} \left[ 1 - \left( \frac{25k_B T}{K_{\text{eff}} V} \right)^{2/3} \right]
\]  

Using this equation, the effective anisotropy \( K_{\text{eff}} \) of the annealed nanoparticles can be estimated to be around \((4.5 \pm 1) \times 10^5\) J/m\(^3\) which is in agreement to the value determined by simulations of the field-dependent magnetisation by describing the magnetisation on the basis of the Landau-Lifshitz-Gilbert equation [15].
Table 1. Element-specific magnetic moments of FePt bulk material and 6 nm nanoparticles (NPs) in the A1 state, wet-chemically synthesised nanoparticles in the (partial) L1_0 state and nanoparticles in the L1_0 state from the gas phase (last row). The values taken from [38] were recalculated using the number of unoccupied final states mentioned in the text and a saturation correction of 2% for the spin and 20% for the orbital magnetic moment respectively.

| system    | Fe atoms in FePt | Pt atoms in FePt |
|-----------|------------------|------------------|
|           | $\mu_s^{eff} [\mu_B]$ | $\mu_l [\mu_B]$ | $\mu_s^{eff} [\mu_B]$ | $\mu_l [\mu_B]$ | $\mu_l / \mu_s^{eff} [%]$ |
| bulk (A1) | 2.92 ± 0.29      | 0.083 ± 0.012    | 0.47 ± 0.02       | 0.045 ± 0.006   | 3.6 ± 1          |
| NPs (A1)  | 2.28 ± 0.25      | 0.048 ± 0.010    | 0.41 ± 0.02       | 0.054 ± 0.006   | 3.8 ± 1          |
| NPs (L1_0) [38] | 2.38 ± 0.26      | 0.204 ± 0.020    | 0.41 ± 0.04       | 0.042 ± 0.008   | 8.8 ± 1          |
| NPs (L1_0) [40]: | 2.21 ± 0.20      | 0.194 ± 0.020    | -                 | -               |                 |

The mean blocking temperature is the temperature at which the simulation gives a vanishing coercivity. Here it is around 180 K (Fig. 6). The non-vanishing coercivity measured above the mean blocking temperature is caused by particles with larger diameters than the mean one since no volume distribution but only the mean volume of the nanoparticles enters equation (1).

Since the magnetic anisotropy is connected to the anisotropy of the orbital magnetic moment, we turn to the discussion of the influence of chemical order on the element-specific spin and orbital magnetic moments which can be found in Table 1. The magnetic moments at the Pt sites remain largely unaffected by thermally induced structural changes. At the Fe sites the effective spin magnetic moment is slightly smaller in the chemically disordered state which may be related to the inhomogeneous composition in that phase (cf. section 4.3). The most significant change occurs for the orbital magnetic moment at the Fe sites: In the chemically ordered state, a fourfold enhanced orbital magnetic moment is found compared to the value in the chemically disordered state. This is related to the reduction of crystal symmetry: In the chemically disordered state, the cubic crystal structure yields a quenching of the orbital magnetic moment in a non-relativistic approximation. A small orbital moment is recovered by the relativistic spin-orbit coupling. After annealing, the crystal structure has changed to a tetragonal distorted one. In this case, even in the non-relativistic approximation the orbital moment does not vanish yielding an enhanced orbital magnetism as presented here. Approximately the same values for magnetic moments at the Fe sites and coercivities were found in FePt nanoparticles prepared by gas-phase condensation [40]. In this method, the particles are in-flight annealed before landed onto a substrate. Note that also a slight change in the spectral shape and intensity of the XANES occurs that may be related to changes in the electronic structure due to chemical ordering [40].

### 4.2. Surface effects on the orbital magnetism

In order to study surface effects on the magnetic properties of FePt nanoparticles, the size-dependence of the magnetic moments was analysed for a composition around 50at% Fe content. The values are listed in Table 2. Compared to the corresponding bulk material, the effective spin magnetic moment is reduced by about 20% at the Fe sites and by 13% at the Pt sites of FePt nanoparticles with a mean diameter of 6.3 nm. This reduction can be explained by the inhomogeneous alloying within the nanoparticles as indicated by the analysis of EXAFS oscillations (section 4.3) and will be discussed in the next section.

In the literature, reductions of the spin magnetic moment are also discussed in terms of spin canting effects [14] due to a large surface anisotropy with respect to the exchange coupling [42,43] or negative contributions of $\mu_l$ that increases at the surface as in the case of Fe clusters [44].
Table 2. Effective spin magnetic moment and orbital magnetic moment at the Fe sites determined from the analysis of the XMCD for FePt nanoparticles of different sizes.

| diameter [nm] | $\mu_{\text{eff}}^S$ [$\mu_B$] | $\mu_l$ [$\mu_B$] | $\mu_{\text{eff}}^S / \mu_l(\text{Fe})$ |
|---------------|-------------------------------|-------------------|-------------------------------------|
| 6.3           | 2.28 ± 0.25                   | 0.048 ± 0.010     | 2.1%                                |
| 4.4           | 2.13 ± 0.21                   | 0.062 ± 0.014     | 2.9%                                |
| 3.4           | 2.01 ± 0.16                   | 0.068 ± 0.015     | 3.4%                                |

These effects cannot be excluded but are expected to be too small to explain the strong reduction of $\mu_s$ with respect to the bulk material. However, they might be the reason for the further decrease of $\mu_{\text{eff}}^S$ at the Fe sites with decreasing particle size to $(2.01 ± 0.16)\mu_B$ for particles with a mean diameter of 3.4 nm. In addition, a small change in the number of unoccupied final states may be present, that was assumed to be constant. The ratio of orbital-to-effective-spin magnetic moment increases with deceasing particle size as qualitatively expected due to the break of symmetry at the surface. In the case of the Fe atoms, it increases from 2.1% for 6.3 nm particles to 3.4% for 3.4 nm particles. These ratios are independent of the number of unoccupied final states and therefore, this change is significant.

4.3. Effect of local deviations from averaged composition

From the bulk Fe$_x$Pt$_{1-x}$ system it is known that the magnetic properties strongly depend on the composition. By analysing the XMCD of bulk-like epitaxial Fe$_x$Pt$_{1-x}$ films as well as from recent bandstructure calculations using the Munich spin-polarised relativistic Korringa-Kohn-Rostoker (SPR-KKR) package [35], it is known that the spin magnetic moments at the Fe sites are a sensitive monitor to compositional changes: The higher the Fe content, the smaller the magnetic moment at the fcc Fe sites [24]. Fe atoms in FePt nanoparticles are in an environment with a local composition containing more Fe than expected from the averaged value. Therefore, assigning the magnetic moment to the averaged composition leads to reduced moments compared to the bulk material (cf. table 1). Since the local composition around the Fe atoms is known from EXAFS analysis and was found to be significantly higher than the averaged value, the magnetic moment at the Fe sites in FePt nanoparticles should be compared to the ones in Fe-rich Fe$_x$Pt$_{1-x}$ alloys. And in fact, for bulk-like alloys with Fe contants between $(58 ± 3)$at% and $(67 ± 3)$at%, the effective spin magnetic moment at the Fe sites ranges between $2.5 \mu_B$ and $2.2 \mu_B$ [24, 41] with an error of 10%. The value of $(2.38 ± 0.25)\mu_B$ for the nanoparticles with a local Fe content of about $(63 ± 10)$ at% perfectly matches these values within experimental errors.

5. Conclusion

Structural and magnetic properties of FePt nanoparticles with different sizes and crystal structures are studied by means of x-ray absorption spectroscopy. To answer the questions raised in the introduction, one can conclude:

- Due to the tetragonal distortion of the lattice in the chemically ordered (L1$_0$) state, the orbital magnetic moment at the Fe sites is fourfold enhanced in FePt nanoparticles [38, 40] while the spin magnetic moment and the magnetic moments at the Pt sites remain largely unchanged.
- As qualitatively expected, the orbital contribution to the magnetic moment increase with decreasing particle size due to the break of symmetry at the surface.
• By EXAFS analyses we found an inhomogeneous composition within the chemically disordered nanoparticles: Fe is an Fe-rich environment whereas Pt is in a Pt-rich environment [21,24]. This leads to smaller magnetic moments of the nanoparticles compared to bulk material of the same averaged composition.

Acknowledgements

We would like to thank S. Sun (Brown U.) for providing nanoparticles, T. Krenke and M. Acet (U. Duisburg-Essen) for preparing the bulk reference sample and J.-U. Thiele (Seagate) for growing Fe$_x$Pt$_{1-x}$ films. For help with the SPR-KKR package M. Košuth, J. Minár, H. Ebert (LMU München), H. Herper and R. Meyer (U. Duisburg-Essen) are gratefully acknowledged. For technical assistance and support during beamtimes we thank F. Wilhelm, A. Rogalev, P. Voisin, S. Feite (ESRF) as well as the BESSY II staff, especially T. Kachel and H. Pfau.

For help in the measurements, U. Wiedwald (U. Ulm), H.-G. Boyen (U. Hasselt), A. Schlachter, N. Friedenberger, and S. Stienen (U. Duisburg-Essen) are acknowledged. This work was financially supported by the DFG (SFB445), the BMBF (05 ES3XBA/5), the ESRF, and the EU (MRTN-CT-2004-0055667, "SystOrbMag").

References

[1] Chantrell R W, Weller D, Klemmer T J, Sun S, and Fullerton E E 2002, J. Appl. Phys. 91, 6866
[2] Antoniak C, Lindner J, and Farle M 2005, Europhys. Lett. 70, 250
[3] Bian B, Laughlin D E, Sato K, Hirotsu Y 2000, J. Appl. Phys. 87, 6962
[4] Kim C K, Kan D, Veres T, Normadin F, Liao J K, Kim H H, Lee S-H, Zahn M, and Muhammed M 2005, J. Appl. Phys. 97, 10G918
[5] Ulmeau M, Antoniak C, Wiedwald U, Farle M, Frait Z, and Sun S 2004, Phys. Rev. B 69, 054417
[6] Dorman J L, Fiorani D, and Tronc E 1997, Adv. Chem. Phys. 98, 283
[7] Ivanov O A, Solina L V, Demshina V A, and Magat L M 1973, Phys. Met. Metall. 35, 81
[8] Visokay M R and Sinclair R 1995, Appl. Phys. Lett. 69, 1692
[9] Thiele J-U, Folks L, Toney M F, and Weller D 1998, J. Appl. Phys. 84, 5686
[10] Shima T, Takamashi K, Takahashi Y K, and Hono K 2004, Appl. Phys. Lett. 85, 2571
[11] Rellinghaus B, Kästner J, Schneider T, Wassermann E F, and Mohn P 1995, Phys. Rev. B 51, 2983
[12] Shima T, Takamashi K, Takahashi Y K, and Hono K 2004, Appl. Phys. Lett. 85, 2571
[13] Boyen H-G, Fauth K, Stahl B, Ziemann P, Küstle G, Weigl F, Banhart F, Heßler M, Schütz G, Gajbhije N S, Ellrich J, Hahn H, Büttner M, Garnier M G, and Oelhafen P 2005, Acta mater. 53, 1648
[14] Sun S, Murray C B, Weller D, Folks L, Moser A 2000, Science 287, 1989
[15] Boyen H-G, Fauth K, Stahl B, Ziemann P, Küstle G, Weigl F, Banhart F, Heßler M, Schütz G, Gajbhije N S, Ellrich J, Hahn H, Büttner M, Garnier M G, and Oelhafen P 2005, Acta mater. 53, 1648
[16] Sun S, Murray C B, Weller D, Folks L, Moser A 2000, Science 287, 1989
[17] Antoniak C, Farle M 2007, Mod. Phys. Lett. B 21, 1111
[18] Muto M, Sasaki Y, Yu A C C, and Inoue M 2004, Adv. Mat. 17, 574
[19] Muto M, Sasaki Y, Yu A C C, and Inoue M 2004, Adv. Mat. 17, 574
[20] Antoniak C, Trunova A, Spasova M, Farle M, Wende H, Wilhelm F, and Rogalev A 2008, Phys. Rev. B 78, 041406(R)
[21] McCale A G, Veal B W, Paulikas A P, Chan S-K, and Knapp G S 1998, Phys. Rev. B 58, 10919
[22] Faxén H and Holtsmark J 1927, Z. Phys. 45, 307
[23] Antoniak C, Spasova M, Trunova A, Fauth K, Wilhelm F, Rogalev A, Minár J, Ebert H, Farle M, Wende H 2009 J. Phys.: Cond. Mat. 21, 336002
[24] Andikunin A L, Ravel B, Rehr J J, and Conradson S D 1998, Phys. Rev. B 58, 7565
[25] Zabinsky S I, Rehr J J, Andikunin A, Albers R C, and Eller M J 1995, Phys. Rev. B 52, 2995
[26] Newville M, Liviu P, Yacoby Y, Rehr J J, and Stern E A 1993, Phys. Rev. B 47, 14126
[27] Ravel B and Newville M 2005, J. Synchr. Rad. 12, 537 (2005)
[28] Newville M 2001, J. Synchr. Rad. 8, 96
[29] The FEFF project homepage, http://leonardo.phys.washington.edu/feff/
[30] Grange W, Maret M, Kappler J-P, Vogel J, Fontaine A, Petroff F, Krill G, Rogalev A, Goulon J, Finazzi M, and Brooks N B 1998, Phys. Rev. B 58, 6298
[32] Thole B T, Carra P, Sette F, and van der Laan G 1992, Phys. Rev. Lett. 68, 1943
[33] Carra P, Thole B T, Altarelli M, and Wang X 1993, Phys. Rev. Lett. 70, 694
[34] Chen C T, Idzerda Y U, Lin H-J, Smith N V, Meigs G, Chaban E, Ho G H, Pellegrin E, and Sette F 1995, Phys. Rev. Lett. 75, 152
[35] Ebert H et al., The Munich SPR-KKR package, version 3.6,http://olymp.cup.uni-muenchen.de/ak/ebert/SPRKKR;
Ebert H 2000, Fully relativistic band structure calculations for magnetic solids – Formalism and Application in Electronic Structure and Physical Properties of Solids, editor: H. Dreyssé, Lecture Notes in Physics 535 (Springer Berlin) p 191
[36] Nakajima R, Stöhr J, and Idzerda Y U 1999, Phys. Rev. B 59, 6421
[37] Fauth K 2004, Appl. Phys. Lett. 85, 3271
[38] Antoniak C, Lindner J, Spasova M, Sudfeld D, Acet M, Farle M, Fauth K, Wiedwald U, Boyen H-G, Ziemann P, Wilhelm F, Rogalev A, and Sun S 2006, Phys. Rev. Lett. 97, 117201 (2006)
[39] Sharrock M P 1994, J. Appl. Phys. 76, 6413
[40] Dmitrieva O, Spasova M, Antoniak C, Acet M, Dumpich G, Kästner J, Farle M, Fauth K, Wiedwald U, Boyen H-G, and Ziemann P 2007, Phys. Rev. B 76, 064414
[41] Fauth K et al., to be published
[42] Labaye Y, Crisan O, Berger L, Greniche J M, and Coey J M D 2002, J. Appl. Phys. 91, 8715
[43] Garanin D. A. and Kachkachi H 2003, Phys. Rev. Lett. 90, 065504
[44] Edmonds K W, Binns C, Baker S H, Thornton S C, Norris C, Goedkoop J B, Finazzi M, and Brookes N B 1999, Phys. Rev. B 60, 472