Chapter 5

Spintronics and Synchrotron Radiation

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Abstract Having access to the electronic and magnetic properties of spintronic systems is of crucial importance in view of their future technological developments. Our purpose in this chapter is to elaborate how a variety of synchrotron radiation-based measurements provides powerful and often unique techniques to probe them. We first introduce general concepts in spintronics and present some of the important scientific advances achieved in the last 30 years. Then we will describe some of the key investigations using synchrotron radiation concerning voltage control of magnetism, spin-charge conversion and current-driven magnetization dynamics.

5.1 General Introduction to Spintronics: From Magnetoresistive Effects to the Physics of Spin-Transfer Phenomena

Whereas electronics relies on the charge of carriers (electrons and holes) to create and convey information, in spintronics the spin of these carriers is also used. The generation of spin currents, their detection as well as their manipulation are the basic principles of any spintronic devices. Since the discovery of giant magnetoresistance (GMR) effect in 1988, several breakthroughs have been achieved and strongly impacted several domains of applications such as, e.g. magnetic sensors, hard-disk drive read heads and more recently nonvolatile memories to cite the most emblematic ones. In this first part, we aim at introducing the basic principles of GMR and some of the pioneer experiments. Then we present some other important breakthroughs such as tunnelling magnetoresistive effects and the manipulation of magnetization without the application of a magnetic field but using spin-transfer effects.

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5.1.1 Giant Magnetoresistance: An Historical Point of View

5.1.1.1 Electric Conduction in Ferromagnets

In $3d$ ferromagnetic (FM) metals, electric conduction comes mainly from $4s$ band electrons, whereas magnetism originates from electrons in $3d$ bands. Resistivity in such materials arises from scattering events of carriers on impurities, phonons and local potentials. In particular, localized d orbitals act as diffusive centres for the s electrons which carry most of the charge current. Moreover, the non-zero net local magnetization comes up with a shift of the d -bands for the two electron spins (up and down), leading to different densities of states (DOS) at the Fermi level (Fig. 5.1 left). At low temperature, the electron spin is observed to be conserved during different sources of scattering (electron–phonon, electron–electron, electron–magnon, ...), allowing to describe the conduction properties in a ferromagnet using a "two-current" model [1]. Spin-up and spin-down electrons carry the current in two separated channels (Fig. 5.1 right). The difference in DOS of $3d$ electrons for spin-up and spin-down channels then leads to different number of scattering events for the two sub-bands, and thus different resistivities. Conduction in ferromagnet can hence be characterized by the spin-asymmetry coefficient α defined by the ratio between the resistivity of spin-down channel over the resistivity of spin-up channel. The spin-asymmetry coefficient is then strongly correlated to the asymmetry of the DOS at the Fermi level. For example, Fe has a spin-asymmetry coefficient lower than 1 whereas for Co and Ni this coefficient is larger than 1.

The experimental demonstration of the two-current model has been provided by a series of experiments in which Fert and Campbell showed that the conduction of a spin channel can be tuned by introducing different types of magnetic impurity [2–5]. Depending on the introduction of two different impurities with similar or opposite spin asymmetries, the impact on the two spin channels will be different (Fig. 5.2). Let us take a Ni matrix ($\alpha > 1$) doped with Co impurities ($\alpha > 1$) and Rh impurities ($\alpha < 1$). With the presence of only Co impurities, scattering in

Fig. 5.1 (left) Scheme of the density of states in a ferromagnetic metal which has d bands shifted. (right) Scheme illustrating the two-current model: spin-up and spin-down electrons carry the current in separated channels. As resistivity arises from s - d transition and is proportional to the $3d$ DOS, the resistivity of a ferromagnet can be schematized by two resistances ($\rho_{\uparrow}$ and $\rho_{\downarrow}$) in parallel

Fig. 5.2 Tailoring the channel conduction with impurities. **a** Resistivity of Ni as a function of Co and Rh doping. Co and Rh having opposite spin asymmetries, scattering is significant for both spin-up and spin-down channels **b** and **c** leading to an increase of resistivity. The resistivity dependence as a function of Co and Rh concentration is not linear **a**. **d** Resistivity of Ni as a function of Co and Au doping. Co and Au having similar spin asymmetries, electrons of only spin-down channel are strongly scattered **e**. The spin-up channel is providing a short-circuit for the electrons **f**. A small linear dependence of the resistivity with Co and Au impurity concentrations is then obtained **d**. Adapted from [5] with permission (Copyright 2008, American Physical Society)

the spin-down channel is enhanced, but the spin-up channel still experiences low resistivity: the resulting global resistivity is low (short-circuit effect). As Rh impurities are introduced, scattering becomes significant for both channels [Fig. 5.2(b,c)], progressively suppressing the short-circuit effect, and the global resistivity quickly increases. The resistivity dependence as a function of Co and Rh concentrations is not linear [Fig. 5.2a]. Instead, when Rh impurities are replaced by Au ones ($\alpha > 1$, i.e. more electrons of the down channel are scattered, like for Ni and Co), such an enhancement of resistivity is not observed. Only electrons of the spin-down channel are strongly scattered by Co and Au impurities and a small linear dependence of the resistivity with Co and Au impurity concentrations is observed [Fig. 5.2d]. The spin-up channel is providing a short-circuit for the electrons [Fig. 5.2e, f].

These experiments indeed brought experimental demonstration of the validity of the two-current model and the proof that the mean resistivity can be varied by tailoring of the channel conduction through impurities. These pioneer experiments can be indeed considered as the pre-concept of the GMR effect in which different resistance states will be achievable by applying a magnetic field in a single heterostructure.

Thanks to the development of molecular beam epitaxy of metallic materials in the late 80's, it became hence possible to grow nanometre-thick magnetic metallic heterostructures (multilayers) involving several separated FM layers. Controlling the magnetic configurations of the magnetic films (parallel or antiparallel configurations of their magnetizations) then grants the possibility to tailor the spin-dependent transport of the system. This has led to the discovery of the GMR effect in 1988.

5.1.1.2 First Observation of GMR Effects

The first studies of GMR effects were done in Fe/Cr superlattices (Fe: FM, Cr: non-FM metal). A key point of these heterostructures was the control of the magnetic configuration. Exploiting the indirect exchange interaction between the magnetic layers transmitted by the conduction electron (called RKKY interaction) through the non-magnetic spacer layer, P. Grünberg and co-workers proved in 1986 that it is possible to align the magnetization of consecutive Fe layers in opposite direction at zero magnetic field by choosing adequate Cr interlayer thickness [6]. The parallel magnetic configuration can then be reached by applying a large magnetic field in order to overcome the RKKY coupling. First magnetoresistance curves have been obtained by the groups of A. Fert in Orsay, France (Fig. 5.3 left) and P. Grünberg in Jülich, Germany (Fig. 5.3 right). A higher resistance state is observed at zero magnetic field at which the antiparallel magnetic configuration is stabilized. By applying a magnetic field, the parallel magnetic configuration is progressively reached inducing a large decrease of the resistance. In the experiments done in Grünberg's group, a GMR effect of 1.5% at low temperature was obtained in the case of a Fe/Cr/Fe trilayer (Fig. 5.3 right). The striking point is that this effect is about one order of magnitude larger in amplitude than the anisotropic magnetoresistance of a single Fe thin film (also shown on Fig. 5.3 right). In Fert's group, measurements were performed in Fe/Cr superlattices instead of trilayers, which allowed to considerably increase the amplitude of the effect that reached 80% for sixty Fe/Cr bilayers (Fig. 5.3 left).

An important development towards the implementation of the GMR systems for applications in read heads for hard-disk drives or sensors was made in 1991 by B. Dieny et al. [7] at IBM who reported the first observation of GMR in simple

Fig. 5.3 Pioneer observation of GMR in Fe/Cr superlattices (left) and in Fe/Cr/Fe trilayer (right). At zero magnetic field, magnetization of consecutive layers is pointing in opposite direction leading to a high resistance state. By applying a magnetic field, a parallel magnetic configuration is reached and so a low resistance state is obtained. For a Fe/Cr/Fe trilayer, the GMR ratio is about 1.5% and one order of magnitude larger than the anisotropic magnetoresistance of a single Fe layer (right). By using Fe/Cr superlattices (60 Fe/Cr repetitions) the GMR ratio reaches 80% at low temperature. Reproduced from [9] (left) and [10] (right) with permission (Copyright 1988 and 1989, American Physical Society)

trilayered structures with two distinct coercive fields, later called spin valves. In these structures, one of the layers switches using low fields while the second is stable up to large fields. It is to be emphasized that the time between the first discovery in a lab and the use of the GMR effect in cutting-edge technology devices has been extremely fast as less than 10 years after the discovery of GMR, and IBM introduced in 1997 spin valves in read heads of hard-disk drives. GMR-based magnetic sensors is now used in a multitude of applications such as monitoring wheel speed, detecting charge current, and fluid flow.

The research on magnetic multilayers and GMR became rapidly a very hot topic. It is not our purpose here to make here a review of all experimental and theoretical results that followed up the pioneer results. A complete review can be consulted in [8].

5.1.1.3 A Simple Model to Describe the GMR

The *two-current model*, described previously, was developed to explain the spin-dependent resistivity in magnetic materials. It can, in a rather simple way, be adapted to describe the giant magnetoresistive effect in these magnetic multilayers. This model is based on two assumptions: 1) $\alpha > 1$: the minority-spin electrons (opposed to local magnetization) are more scattered than those of majority spin (aligned with local magnetization); 2) the spin is conserved when the electrons are scattered. These two conditions are fulfilled at low temperature.

Two geometries can be considered to evaluate the resistance of such a structure: either the current flows in the direction of the layer planes (known as "current-in-plane GMR", CIP-GMR), or the current flows in a direction perpendicular to the layer plane (known as "current-perpendicular-to-the-plane GMR", CPP-GMR). A similar description can be used to account for the magnetoresistive properties for both geometries, without entering into the fine discussion about the actual physical spin-transport mechanisms, as long as the layer thickness remains small compared to a characteristic length associated with each geometry: the mean free path for the CIP case and the spin-flip length for the CCP case [11].

Let us note r the resistance of a FM layer for the majority-spin channel and R the resistance for the minority-spin channel, with $r < R$. Here the resistance of the NM separating layer is neglected. As illustrated in Fig. 5.4, an electron will have to pass through adjacent FM layers. Depending on whether these layers have parallel (P) or anti-parallel (AP) magnetizations, the resulting resistance shall differ.

In the (P) configuration, the electrons with spin ($\uparrow$) and ($\downarrow$) are either the electron spin majority or minority in all magnetic layers. Spin ($\uparrow$) electrons then experience identical resistances $r_{\uparrow} = r$ when crossing each magnetic layer, while spin ($\downarrow$) electrons experience identical resistances $r_{\downarrow} = R$. The mean resistance then writes $r_P = \frac{Rr}{r+R}$. In case of materials with large spin asymmetry ($\alpha \gg 1$ and $r \ll R$), the multilayer is short-biased by the spin ($\uparrow$) channel and the total resistance is $r_P \approx r$.

For the (AP) configuration, the electrons of the two ($\uparrow$) and ($\downarrow$) channels correspond alternately to electron spin majority and minority in consecutive magnetic

Fig. 5.4 Illustration of the two-current model: conduction of an electron in a ferromagnetic metal/normal metal/ferromagnetic metal (FM/NM/FM) multilayer. The electrons with a spin aligned parallel (resp. anti-parallel) to the local magnetization see a resistance r (resp. R) through this magnetic layer. The equivalent resistance circuit is presented for two configurations: consecutive ferromagnetic layers with parallel magnetization or anti-parallel magnetization

layers, and thus the effect of short-circuit by one of the channels is suppressed. Consequently the two channels have the same resistance $\frac{R+r}{2}$ and the total resistance is $\frac{R+r}{4}$, which is in general much larger than $r_P = r$.

This brings the possibility of switching between high and low resistance states by simply changing the relative orientation of the magnetization of consecutive layers. If we can engineer these structures to stabilize the two (P) and (AP) magnetic states in absence of any external field, then such system defines a magnetic bit to store the information, using the powerful stability of the magnets, already well known through the development of hard-disk drives. The state can then be read through a very simple process, i.e. by measuring the resistance of the stack.

Following this model, the amplitude of the GMR ratio can be simply deduced

$$GMR = \frac{R_{AP} - R_P}{R_P} = \frac{(R - r)^2}{4Rr} \quad (5.1)$$

In short, the understanding and expertise acquired during the 60s and 70s for tailoring the spin-transport properties combined with the development of metallic magnetic multilayers in the 80s has led to the discovery of the GMR effect in 1988. Less than 10 years after the discovery of the GMR effect considered as the birth of spintronics, this effect has largely participated to the explosion of the amount of data storage in the mid-90s through the commercialization of the first generation of spintronic devices such as magnetic sensors used as read heads in hard-disk drives.

5.1.2 Tunnelling Magnetoresistance

The pioneer observation of tunnelling magnetoresistance (TMR) in magnetic tunnel junctions (MTJs) at room temperature in 1995 is considered to be the second breakthrough in spintronics, leading to a second generation of spintronic applications such as magnetic random access memory (MRAM). In this section, we describe the first TMR measurements, then the two standard models describing TMR effects.

5.1.2.1 First Experimental and Theoretical Studies

A magnetic tunnel junction is a device in which two FM electrodes are separated by an ultra-thin insulating barrier. The first MTJ exhibiting TMR has indeed been reported in 1975 by M. Jullière [12] who measured a TMR ratio of 14% at 4.2K in a Fe/Ge/Co junction. Twenty years later TMR effects have been observed at room temperature by J. S. Moodera [13] and T. Miyazaki [14] using amorphous alumina as tunnel barrier. In Fig. 5.5, the TMR curve obtained for a CoFe/Al₂O₃/Co MTJ [13] is displayed. The TMR ratio is around 12%, at room temperature. This increase in the resistance difference between the two states allows an easier detection of them and let appear the potential of such spintronic devices for memory applications.

Such TMR effect has been first explained by M. Jullière in 1975. Keeping the free-electron approximation, he proposed an additional assumption: the *electron spin conservation* during the tunnelling process, meaning that electrons can tunnel from one FM electrode toward the second FM electrode only into empty states having identical spins. This simple model easily allows to understand that the tunnelling current in the (P) and (AP) alignments of magnetizations will differ.

As illustrated in Fig. 5.5, for the parallel (P) configuration of the magnetizations: when a small voltage is applied, majority-spin (denoted as $\uparrow$) electrons from the injector will tunnel toward majority-spin ($\uparrow$) empty states of the collector, allowing "high" current. At the same time, minority-spin electrons from the injector will tunnel toward minority-spin ($\downarrow$) empty states of the collector, allowing "low" current.¹ The conductance then writes

$$G_P \propto D_1^\uparrow(E_F)D_2^\uparrow(E_F) + D_1^\downarrow(E_F)D_2^\downarrow(E_F) . \quad (5.2)$$

In the case of antiparallel (AP) configuration of the magnetizations, majority-spin ($\uparrow$) electrons from the injector will tunnel toward minority-spin ($\downarrow$) empty states of the collector. Minority-spin ($\downarrow$) electrons from the injector will tunnel toward majority-spin ($\uparrow$) empty states of the collector. Both currents are then low. The conductance then writes

¹Note here that the arrow does not define any direction of the spin, but rather the fact that it is aligned with the local magnetization. In the (AP) case, a spin with one direction will correspond to ($\uparrow$) spin in one electrode and ($\downarrow$) spin in the other.

Fig. 5.5 TMR curve of a CoFe/Al₂O₃/Co MTJ recorded at room temperature [Adapted from [13] with permission (Copyright 1995, American Physical Society)]. Schematic of the spin-dependent tunnelling process through an insulating barrier when their magnetizations are aligned parallel (left) and antiparallel (right) to one another. The process is assumed to be purely elastic, so that no mixing of spin states occurs during the tunnelling process

$$G_{AP} \propto D_1^\uparrow(E_F)D_2^\downarrow(E_F) + D_1^\downarrow(E_F)D_2^\uparrow(E_F). \quad (5.3)$$

Using these results for the conductance, the TMR ratio, characterizing the resistance difference in (P) and (AP) configurations, is expressed as

$$TMR = \frac{G_P - G_{AP}}{G_{AP}} = \frac{R_{AP} - R_P}{R_P} = \frac{2P_1P_2}{1 - P_1P_2}, \quad (5.4)$$

where P_i defines the spin polarization for each electrode as

$$P_i = \frac{D_i^\uparrow(E_F) - D_i^\downarrow(E_F)}{D_i^\uparrow(E_F) + D_i^\downarrow(E_F)}. \quad (5.5)$$

A key issue to get an accurate prediction of TMR is to properly estimate the actual amplitude and sign of the spin polarization for a given FM material. Obviously, the largest the spin polarization is, the highest will be the TMR amplitude. It thus explains the strong research activity in the last decade on material science to investigate novel families of materials (magnetic oxides, Heusler alloys, etc.) for which some of their

compounds are predicted to be half metallic, i.e. to have a 100% spin polarization at the Fermi level.

Jullière's model has enjoyed much success in correctly predicting TMR amplitudes in most MTJs with amorphous barriers, until the emergence of MTJs based on crystalline insulating barriers, notably magnesium oxide (MgO) barriers in the beginning of the 2000s [15, 16]. The model described previously is actually oversimplified in that it considers an identical tunnelling process of every electron, independently on their band belonging, hence forgetting about the interplay of the band structures throughout the MTJ heterostructure.

5.1.2.2 Coherent Tunnelling in Epitaxial Magnetic Tunnel Junctions

Another major breakthrough in the field of spintronics corresponds to the introduction of crystalline MgO tunnel barriers, that is today's standard in MTJ implemented in MRAMs, read heads and field sensors working with TMR effect. These experimental developments have been stimulated by some theoretical calculations made by W. Butler [17] predicting that huge TMR ratios as large as 1600% are anticipated for *epitaxially grown* Co or Fe electrodes on crystalline, instead of amorphous MgO. These calculations, developed with *ab initio* methods, derive the tunnelling probability of each kind of electrons, depending on their orbital symmetry.

Contrary to amorphous alumina [Fig. 5.6a left], in crystalline MgO tunnel junctions, the electron wave functions in the FM material are coupled with evanescent wave functions having the same symmetry in the barrier [Fig. 5.6a right]. Through *ab initio* calculations, it was then predicted that the tunnelling probability of an electron strongly depends on the orbital symmetry of the electron (of the band it belongs to). Beyond the two-current (up-spin and down-spin) model, the system behaves as if it exists an independent current channel for each band and each spin, leading to a possible effective *symmetry filtering of the tunnelling current*. The tunnel barrier can, therefore, filter the wave functions and thus select the spins in the electronic transport.

This mechanism of orbital selection for the tunnel conductance is presented in Fig. 5.6. In Fig. 5.6b, band dispersion of *bcc* Fe(001) for the minority and majority spins is shown. Δ_1 Bloch states are present at the Fermi level only for the majority spins. In Fe(100)/MgO/Fe(100) systems, band structure calculations have demonstrated that majority-spin electrons [see Fig. 5.6c] are mainly filling Δ_1 symmetry states (hybridized states with *spd* characters), whereas minority-spin electrons [see Fig. 5.6d] are filling Δ_2 symmetry states (*d* type states) and Δ_5 symmetry states (hybridized *pd* states). Moreover, this is crucial for getting a large TMR ratio, the tunnelling exponential decay is much stronger for Δ_2 and Δ_5 states compared to Δ_1 states. For $d_{\text{MgO}} = 8$ monolayers, which is a typical barrier thickness that can be achieved experimentally, the probability of transmission of Δ_1 electrons is larger than for Δ_5 electrons by 10 orders of magnitude. Ultimately, only Δ_1 electrons contribute significantly to the current. It is this filtering effect which can explain the large values of TMR expected on epitaxial or highly textured structures and that has made

Fig. 5.6 **a** Schemes illustrating electron tunnelling through an amorphous Al-O barrier (left) and through a crystalline MgO barrier (right). **b** Band dispersion of *bcc* Fe(001) for the minority and majority spins. Δ_1 Bloch states are present at the Fermi level only for the majority spins. Tunnelling DOS for Fe/MgO/Fe at $k_{\parallel} = 0$ for majority **c** and minority **d** spins. Decay of Δ_1 Bloch states in MgO is less attenuated than the Δ_1 Bloch states. **e** TMR of 600% obtained at room temperature for a CoFeB/MgO/CoFeB MTJ. Adapted from [18] (**a**, **b**) with permission (Copyright 2007, Institute of Physics), from [17] (**c**, **d**) with permission (Copyright 2001, American Physical Society) and from [19] (**e**) with permission (Copyright 2008, American Institute of Physics Publishing)

the MgO-based tunnel junctions at the core of the development of new spintronic devices like the MRAMs.

After these theoretical predictions, a strong research effort has been made to obtain epitaxial growth of structures for Fe/MgO/Fe or CoFeB/MgO/CoFeB [20, 21]. These efforts have resulted in a TMR of about 600% obtained in 2009 [19] in CoFeB/MgO/CoFeB MTJs at room temperature [see Fig. 5.6e]. An excellent review on the physics of tunnelling transport in MgO-based systems and the experimental state of the art for TMR has been written by S. Yuasa [18].

Since the measurements of TMR effects at room temperature in 1995, a lot of work has been done in order to both increase the TMR ratio and reduce the MTJ resistance. This has been achieved by developing high-quality crystalline MgO-based MTJs. This effort has led to the development of a new class of magnetic memories called MRAMs.

5.1.3 Magnetization Manipulation without Magnetic Fields

For practical applications, e.g. MRAMs, the possibility to manipulate the magnetization of FM electrodes in spin valves or MTJs without using magnetic field is required. This corresponds to the next breakthrough with the prediction and the observation of spin-transfer effects, first providing a new way to reverse the magnetization in a nanostructure but also to generate in some cases steady precession of the magnetization. In this section, we briefly present the physics of spin-transfer torque (STT) and also discuss an alternative approach consisting in the manipulation of the magnetization through the application of a voltage.

5.1.3.1 Spin-Transfer Torque

In the first generation of MRAM devices developed in 2004, the magnetization reversal between the two possible magnetic states was realized by using local magnetic fields generated by electrical currents flowing in lines close to each magnetic element. This writing process rapidly suffered from both the large energy consumption needed to generate large enough magnetic fields and from cross-talk problems due to the difficulty to write a single bit. A solution came out from a major breakthrough in spintronics in 1996 when it has been proposed that a spin-polarized current flowing through magnetic multilayers provides a new way to manipulate the magnetization of a ferromagnet. This new effect, which was called spin-transfer effect, has become rapidly a very hot topic. Moreover, it is at the basis of a new generation of magnetic memories, called spin-transfer torque-MRAMs (STT-MRAMs).

The concept of spin-transfer effect was proposed in 1996, concomitantly by J. Slonczewski [22] and L. Berger [23]. To describe this effect, one can take a standard spintronic structure composed of a fixed FM electrode F_1 and a free FM layer F_2 separated by a NM spacer. When the electrons flow from layer F_1 to F_2 , the current becomes spin-polarized after passing through F_1 . This non-zero spin polarization is aligned along the magnetization direction in F_1 and propagates into the NM metallic spacer or tunnels in the case of an insulating barrier, so that it arrives at the NM/ F_2 interface. If the magnetization $\vec{M}_1$ and $\vec{M}_2$ are non-collinear, it results that a component of the spin current transverse to F_2 exists at the NM/ F_2 interface [green arrows in Fig. 5.7a]. When the electrons penetrate into F_2 , the spin of the conduction electron becomes aligned over a very short distance (within a few atomic distances) toward the magnetization direction in F_2 because of a strong exchange interaction between conducting (s) electrons and localized (d) electrons, the latter being responsible of the magnetic moments.

During all this process, the total spin angular momentum is conserved. Thus the transverse component of the spin current $\Delta \vec{m}$ lost by the electrons when passing through F_2 is indeed absorbed and transferred to the local magnetization of F_2 . This transfer of spin angular momentum results in a torque exerted by the spin-polarized current on the local magnetization. For this current polarity, when the

Fig. 5.7 **a** Scheme of a magnetic trilayer structure for illustrating the concept of spin-transfer torque. **b** The transverse component of the angular momentum Δm of the spin-polarized current is transferred to the magnetization. It results in a torque exerted on the magnetization $\vec{M}_2$, of the thin magnetic layer, which aligns along magnetization $\vec{M}_1$ for positive current. **c** Schematic representation of the torques acting on the magnetization including spin-transfer effect

current increases enough, it tends to align the magnetization $\vec{M}_2$ along the direction of the spin polarization of the current, i.e. along the magnetization $\vec{M}_1$ [Fig. 5.7b]. As all the process occurs in the first atomic planes after the interface, the spin-transfer mechanism is an interfacial effect.

After having introduced the spin-dependent transport mechanisms at the origin of STT, our aim is now to address the influence of this torque on the magnetization dynamics of the layer F_2 . A classical approach to describe the dynamical motion of a magnetization is a differential equation, named the Landau–Lifschitz–Gilbert equation, to which we add Slonczewski's component of spin torque.² This equation has the

²For simplicity we have only introduced the main STT called "in-plane torque". A second spin torque, called "field-like torque" or "out-of-plane torque", is similar to a torque exerted by a field along the magnetization of F_2 and its action might, therefore, be included into $\vec{H}_{eff}$.

following form, taking magnetizations normalized to the saturation magnetization, $m_{i=1,2}$ for the two FM layers

$$\frac{d\vec{m}_2}{dt} = -\gamma_0(\vec{m}_2 \times \vec{H}_{eff}) + \frac{\alpha}{\gamma_0 M_{S2}}(\vec{m}_2 \times \frac{d\vec{m}_2}{dt}) - P_{spin} \frac{J}{e} \frac{h}{2} \frac{g\mu_B}{M_{S2}^2 t} [\vec{m}_2 \times (\vec{m}_2 \times \vec{m}_1)] \quad (5.6)$$

The first term corresponds to the tangential force describing the magnetization precession around effective field $\vec{H}_{eff}$ [green arrow in Fig. 5.7c], which takes into account the external applied magnetic field, magnetic anisotropy fields, coupling fields, etc. The second term gives a phenomenological description of magnetization dissipation of the system. The coefficient α , named Gilbert damping, (about 10^{-2} for standard FM materials), describes the damping rate of the motion of $\vec{m}_2$ towards the equilibrium position oriented along $\vec{H}_{eff}$ [blue arrow in Fig. 5.7c]. The magnetization relaxation is induced by a damping force tangential to the magnetization trajectory. The third term is the so-called Slonczewski torque (or in-plane torque) where μ_B is the Bohr magneton, t the layer thickness, J the injected current density, and P_{spin} the amplitude of spin polarization at the interface NM/F₂ and M_{S2} the magnetization of the ferromagnet F₂. This simplified description allows making clear the nature of the main contribution of the spin-transfer force that can be described as a non-conservative force acting in the same direction than the natural damping, i.e. perpendicularly to the magnetization trajectory. Depending on the sign of the injected current, the spin-transfer torque decreases or increases the effective damping [purple and red arrows in Fig. 5.7c]; it favours the stability of the parallel or the antiparallel magnetic configuration.

For large current density ($\sim 10^7$ A cm⁻² for a typical material system), the STT fully compensates the damping torque, steady magnetization precession occurring at the ferromagnetic frequency (typical in the GHz range for the ferromagnetic materials and in the THz range for antiferromagnetic ones) can be established. The spin transfer induced magnetization dynamics can convert into oscillations of resistance through the magnetoresistive effect described previously, and in turn into a radiofrequency voltage signal. Spin-torque effect thus makes it possible to convert a *dc* current into a *rf* voltage and so to build microwave oscillators at the nanoscale.

For larger current density, the torque can become sufficient to commute the magnetization between the two stable configurations. A negative current will, for instance, destabilize the parallel magnetization configuration, while stabilizing the anti-parallel configuration, allowing to commute from P to AP configuration. Identically, a positive current allows to commute from AP to P configuration.

First experimental results allowing to confirm the theoretical predictions of the existence of STTs have been obtained by M. Tsoi et al., using point contact geometry for injection of a large current into a magnetic layer [24]. Then after, it has been demonstrated that magnetization commutation can be achieved back and forth by the STT effect in Co/Cu/Co spin valves [25]. Figure 5.8a represents magnetization rever-

Fig. 5.8 **a** Current-induced magnetization switching in a Co/Cu/Co nanopillar spin valve. Jumps in the curves correspond to the magnetization reversal of the ferromagnetic free layer. **b** Current-induced magnetization switching in a CoFeB_{2.5}/AlOx/CoFeB_{2.5} MTJ. **c** Domain wall motion induced by a current in Co/Cu/Py spin valves. State 1 and 2 correspond to different positions of the domain walls in the strip. By applying a large enough current it is possible to move the domain walls and thus reach the parallel (P) or antiparallel (AP) magnetic configuration. Reproduced from [25] **(a)** with permission (Copyright 2000, American Physical Society), from [26] **(b)** and [27] **(c)** with permission (Copyright 2005 and 2002, American Institute of Physics Publishing)

sal induced by a current for two different applied magnetic fields. Current density in the 10^7 A cm^{-2} range corresponding to few mA for their nanopillars is needed to switch the Co free electrode magnetization. Few years later similar results have been obtained in MTJs using AlOx and MgO tunnel barriers [26]. Figure 5.8b illustrates the magnetization reversal by the current for a CoFeB/AlOx/CoFeB magnetic tunnel junction. A smaller current (less than 1 mA) is needed to commute the magnetization and thus to switch between the parallel and antiparallel magnetic configurations. This result shows that STT effects can be used not only in metallic spin valves but also in MTJs and has opened the door for the development of new STT-MRAM technologies. Finally, STT has been also used to induce magnetic domain wall motion. Figure 5.8c represents the first observation of domain wall motion by spin transfer in a Co/Cu/Py trilayer, where Py stands for permalloy (Ni₈₀Fe₂₀) [27]. Starting from configuration 1 or 2 (the domain wall is located at the two-third of the strip) it is possible to move the domain wall by a current to reach the parallel (P) or antiparallel (AP) magnetic configuration. The current density needed to move these domain walls is about 10^7 A cm^{-2} . Note that for all these pioneer experiments, a magnetic field is applied. Since then, a lot of work has been done and now magnetization manipulation without applying magnetic field is feasible. Later, STT has been also used to manipulate other magnetic textures such as magnetic vortices [28] or skyrmions, or generate dynamics [29, 30]. To conclude, STT is now used to write the information of a single bit in magnetic memories.³

³For more information on the latest MRAM development, see <https://www.mram-info.com/>

Fig. 5.9 (left panel) Voltage can be used to modify magnetic anisotropy, coercive fields, magnetization values, exchange bias, Curie temperature or magnetoresistance. This can be achieved by different mechanisms such as charge, strain exchange coupling, orbital ordering or electromigration (right panel). Reproduced from [31] with permission (Copyright 2017, Elsevier)

5.1.3.2 Voltage Control of Magnetism

Although STT is very efficient to manipulate magnetization, alternative or complementary approaches have been explored such as the voltage control of magnetism. Magnetic anisotropy, coercive fields, magnetization magnitude, exchange bias or Curie temperature can be tuned by a voltage [Fig. 5.9 (left panel)]. Different mechanisms such as charge accumulation/depletion, electromigration, strain or orbital ordering are involved [Fig. 5.9 (right panel)].

The samples to study the effect of an electric field on the magnetization are generally composed of an ultra-thin FM film in contact with either dielectric, ferroelectric or piezoelectric materials. A wide variety of FM materials (metals, oxides or diluted magnetic semiconductors) have been also studied [31]. Depending on the mechanism used to control magnetism through an applied voltage, the ferromagnet thickness has to be adjusted. As charge, orbital ordering or electrochemistry are mainly interfacial effects, the thickness of the ferromagnet should be close to the screening length which is in the angstrom range for metal and in nanometre range for dilute magnetic semiconductors. Thicker ferromagnets, up to the micrometre range, can be used if strain is involved. In the following, several examples illustrating the variety of devices studied are displayed. Readers are invited to refer to the article of C. Song et al. [31] for a complete review.

In Fig. 5.10, we show different examples where the Curie temperature, the coercive field and the magnetic anisotropy are modified by an electric field. This magnetization control involves different mechanisms such as charge or strain effects. One of the first demonstration of magnetization manipulation by an electric field has been done using a dilute magnetic semiconductor [32]. In Fig. 5.10a, b, Hall effect curves recorded at different gate voltages for a (In,Mn)As-based field-effect transistor are presented. Magnetic properties and notably the Curie temperature of the diluted magnetic semiconductor (In,Mn)As being dependent on the hole density, it is thus possible to tune the Curie temperature by an electric field modulating the carrier density [33]. Setting the sample temperature close to its Curie temperature, the switching

Fig. 5.10 Examples of voltage control of the magnetization. **a, b** Hall effect curves recorded for different gate voltages in a dilute magnetic semiconductor (In, Mn) As field-effect transistor. The gate voltage induces a modulation of the hole concentration in (In, Mn)As and thus a modulation of the Curie temperature. **c, d** Variation of the coercive field of a FePt thin film as a function of bias voltage. The bias voltage modifies the number of 3d electrons and thus changes the magnetocrystalline anisotropy. **e, f** Modification of the magnetic anisotropy induced by strain in a FeGaB/PZN-PT multi-ferroic heterostructure. The voltage leads to lattice modulation of the PZN-PT ferroelectric inducing a strain modulation of the magnetostrictive FeGaB layer. Under an electric field of 8 kV cm^{-1} a magnetic field of 70 mT is needed to align the magnetization along the [100] direction. Adapted from [32] (**a, b**) with permission (Copyright 2000, Nature Publishing Group), from [34] (**c, d**) with permission (Copyright 2007, American Association for the Advancement of Science), from [31] (**e**) with permission (Copyright 2017, Elsevier) and from [35] (**f**) with permission (Copyright 2009, John Wiley and Sons)

between ferromagnetic and paramagnetic states can be realized by applying, respectively, a negative and a positive gate voltage (125 V). A similar approach has been investigated to tune the coercive field of the FePt FM intermetallic compound [34]. Using an electrolyte to modify electron density at the FePt interface, a modification of 4.5% of the coercive field is observed [Fig. 5.10c, d]. The variation in the number of 3d electrons directly affects the magneto-crystalline anisotropy and thus the coercive field. Interestingly, this device geometry allows tuning the magnetic properties of a ferromagnet by applying a low bias voltage (below 1 V). Although the modification of the coercive field is quite small, this first demonstration of voltage control of a FM metal was quite encouraging as it opens the door to room temperature modulation of magnetism by applying small voltage. Another approach to modify the magnetic anisotropy is to apply a strain. By combining ferroelectric and magnetostrictive materials, it is possible to modify the magnetic properties by a voltage. Under voltage, the lattice of the ferroelectric layer is modulated through the inverse piezoelectric effect and will thus induce a strain modulation in the ferromagnet. In Fig. 5.10(e,f), we show an example where the in-plane magnetic anisotropy is modified by an electric field. Whereas at zero electric field, the remanent magnetization is along the [100] direction, an electric field of 8 kV cm^{-1} allows rotating the magnetization and so a magnetic field of 70 mT is now needed to saturate it along the [100] direction.

The voltage control of magnetization is obviously less mature than STT or spin-orbit torque (SOT) technologies. However, the large variety of ferromagnets (metals, semiconductors, oxides) and gating materials (dielectric, ferroelectric, electrolyte) that can be used, makes this field fascinating and promising to reduce power consumption in memory technologies, which remains still a crucial issue.

5.1.4 Summary

We have introduced some breakthroughs in the field of spintronics achieved during the last 30 years. Giant magnetoresistance has emerged rapidly as a promising effect to build efficient magnetic sensors working at room temperature. The ability to manipulate magnetization by STT in MTJs has allowed developing new magnetic memories such as MRAMs that are now commercially available. This is of course not exhaustive and there are a lot of other exciting and promising research fields such as spin-orbitronics, magnonics, molecular spintronics or antiferromagnetic spintronics, to cite only few of them. Spin-orbitronics is certainly the most active field nowadays. Emblematic topics are conversion between charge and spin currents, spin-polarized surface and interface states or novel chiral magnetic textures (skyrmions and domain walls). See for example A. Soumyanarayanan et al. for a review [36].

In all the spintronic effects that we have just introduced, like in many other fields, interfaces play a key role, and therefore a deep knowledge of the electronic and magnetic properties of these interfaces is desired. Synchrotron radiation-based measurements such as absorption and photoelectron spectroscopies and microscopies are powerful techniques to probe these interfaces.

5.2 Examples of Synchrotron Radiation Contribution to Spintronics

In the following, several examples for which synchrotron radiation-based measurements have allowed a better understanding of spintronic devices are presented. We have selected three spintronics topics: (i) voltage control of magnetism; (ii) spintronics with pure spin currents; (iii) current-driven magnetization dynamics.

5.2.1 Voltage Control of Magnetism

5.2.1.1 Effect of Charge Accumulation/Depletion and Electromigration

Perpendicular magnetic anisotropy (PMA) has been widely investigated in NM/FM/oxide heterostructures. For example, by varying the oxidation time of top Al layer in Pt/Co/AlOx devices, it has been demonstrated that the PMA is induced by Co–O bonds [37]. Figure 5.11(a,b) represent extraordinary Hall effect curves and X-ray absorption spectroscopy (XAS) measurements at the Co $L_{2,3}$ edges for Pt/Co/AlOx samples with different oxidation times. The extraordinary Hall effect allows detecting the magnetic easy axis and XAS measurements probe the electronic properties of the cobalt layer. A metallic Co/Al interface resulting from Al under-oxidation, or the formation of a CoO layer due to over-oxidation, both produce magnetization aligned within the plane of the film. Hence optimized oxidation, leading to Co–O–Al bonds, induces a perpendicular magnetic anisotropy. Instead of playing with oxidation time, D. Chiba et al. [38] have shown that it was possible to switch the easy magnetic axis from in-plane to out-of-plane by applying a bias voltage. In Fig. 5.11c, we show that a bias voltage of 10 V allows to modify the magnetic

Fig. 5.11 **a** Extraordinary Hall effect as a function of oxidation time measured in a Pt/Co/AlOx trilayer. **b** XAS spectra recorded at the Co $L_{2,3}$ edges for different oxidation times. These measurements show that optimized Al oxidation induces perpendicular magnetic anisotropy due to Co–O bonds. **c** Switching between in-plane magnetization to out-of-plane magnetization by applying a bias voltage for a Pt/Co/MgO sample. Reproduced from [37] (**a**, **b**) with permission (Copyright 2008, American Institute of Physics Publishing), and from [38] (**c**) with permission (Copyright 2011, Nature Publishing Group)

anisotropy in Pt/Co/MgO [38]. This magnetic anisotropy modification with voltage can originate from charge accumulation or oxygen electromigration.

C. Bi et al. have performed Hall resistivity and XAS measurements at the Co L_3 edge on a Pt/Co/GdOx sample [39]. A clear correlation between the evolution of the magnetic anisotropy and the Co oxidation was demonstrated [Fig. 5.12(a,b)]. In Fig. 5.12b, one can see that the shape of the Co XAS spectra are modified with voltage. Whereas a negative voltage induces some fine structures in the XAS spectra indicating a Co oxidation, a positive voltage reduces the Co layer, and therefore Co spectra become similar to that of metallic Co. Thus the Co magnetic anisotropy in this Pt/Co/GdOx structures can be reversibly controlled by voltage via Co oxidation and reduction. This result confirms the earlier experiments performed by F. Bonell et al. [40] on Au/CoFe/MgO samples. Hence, in these examples, the modification of magnetic anisotropy is rather due to oxygen electromigration than charge accumulation.

In V/Fe/MgO devices, S. Miwa et al. [41] have indeed observed a different behaviour. A slight change of coercive field with bias voltage has been measured [see Fig. 5.12c] but without any modification of the Fe $L_{2,3}$ XAS and X-ray magnetic circular dichroism (XMCD) spectra with the applied voltage, suggesting that Fe is not oxidized [Fig. 5.12d]. This is quite surprising since the electric field applied is similar in Pt/Co/GdOx and V/Fe/MgO experiments. This result shows that modifications of the coercive field is not induced by electromigration but is rather due to charge accumulation.

To conclude, this careful investigation of the electronic properties of the FM/oxide interface has allowed to unveil the origin of voltage control of magnetic anisotropy in each NM/FM/oxide structures and thus to discriminate between electromigration and charge effects.

5.2.1.2 Effect of Strain

Strain has been also used to tune magnetic anisotropy and can be very efficient if magnetostrictive materials are used. By combining piezoelectric and magnetostrictive materials, it is possible to control the magnetic anisotropy by voltage. Here we show an example where photoelectron emission microscopy (PEEM) measurements have been performed to highlight the correlation between strain and magnetism [42]. A magnetostrictive CoFe_2O_4 layer has been deposited on a piezoelectric BaTiO_3 substrate. BaTiO_3 can have domains with different unit cell parameters, and therefore can induce different domains with different strains in the CoFe_2O_4 layer. X-ray linear dichroism (XLD) being sensitive to local coordination, different XLD spectra will be obtained. In Fig. 5.13d, XLD-PEEM⁴ image recorded at the Fe L_3 edge are

⁴XAS spectra can be also recorded on the different domains allowing to measure the local electronic properties.

Fig. 5.12 **a** Hall resistance for a Pt/Co/GdOx sample as deposited (red) and after an applied bias voltage leading to an electric field of -625 kV cm^{-1} (blue) and $+625 \text{ kV cm}^{-1}$ (purple) showing an in-plane to out-of-plane magnetization transition. **b** XAS and XMCD spectra recorded at the Co L_3 edge. A negative voltage induces an oxidation of the Co layer, whereas a positive voltage reduces the Co layer. **c** Magnetic hysteresis loop of a V/Fe/MgO sample obtained for two bias voltages showing a modification of the coercive field. **d** Fe $L_{2,3}$ XAS and XMCD spectra recorded at $+4 \text{ V}$ and -4 V . Fe oxidation/reduction is not observed demonstrating that the coercive field modification is rather attributed to a charge effect. **a, b** Adapted from [39] with permission (Copyright 2014, American Physical Society) and **c, d** from [41] with permission (Copyright 2015, American Institute of Physics Publishing)

presented showing different strain domains for the CoFe_2O_4 layer induced by the BaTiO_3 substrate. By recording XMCD-PEEM images at the Fe L_3 edge, domains with different magnetic signals are observed. In Fig. 5.13, it appears clearly that a correlation exists between the strain state and the magnetic signal. Black domains in the XLD-PEEM [see Fig. 5.13d] correspond to a magnetization along the $[100]$ direction [see Fig. 5.13a], whereas for white domains, the magnetization is along the $[010]$ direction [see Fig. 5.13c]. Hence, by performing XLD-PEEM and XMCD-

Fig. 5.13 XMCD-PEEM (a–c) and XLD-PEEM (d) images recorded at Fe L_3 edge for a BaTiO₃/CoFe₂O₄ sample. XMCD-PEEM images show magnetic stripes along the [100] and [010] directions. These magnetic stripes are correlated to domains observed in the XLD-PEEM image **d** which indicates different strain states of CoFe₂O₄ induced by the BaTiO₃ substrate. Reproduced from [42] with permission (Copyright 2012, American Physical Society)

PEEM measurements, it is possible to locally observe the influence of strain on magnetism.

We have shown that XAS has allowed determining the origin of the voltage control of magnetism in NM/FM/oxide structures. In particular, it is possible to disentangle charge and electromigration effects. By performing PEEM measurements with both circular and linear polarized X-rays, the correlation between strain and magnetism can be directly probed. Hence, synchrotron radiation-based measurements are useful to better understand the mechanisms of the voltage control of magnetism.

5.2.2 Spintronics with Pure Spin Current

The generation of pure spin current from heat, charge current, light or vibration is an active and promising research field [43]. A wide variety of materials (ferromagnets,

heavy metals, semiconductors, insulators, topological insulators, etc.) and devices are currently investigated. In this paragraph, we aim at discussing the spin-charge conversion using topological insulators as well as the heat-spin current conversion in ferromagnetic insulator/paramagnetic metal devices.

5.2.2.1 Spin-Charge Conversion

Conversion between charge and spin currents has been a very active branch of spintronics in the last couple of years. Such a conversion can be achieved in bulk materials (the so-called spin Hall effect) or at interfaces (Edelstein–Rashba effect) relying on the spin–orbit interaction and/or extrinsic effects. The studied materials are usually NM metals with a large spin–orbit coupling. A second approach is to rely on spin–orbit properties at the interfaces either at Rashba interfaces or through the surface states of topological insulators (TIs). A complete review of this approach can be found in J. Sinova et al. [44] or A. Soumyanarayanan et al. [36].

In the following, we provide an example where angle-resolved photoemission spectroscopy (ARPES) measurements have allowed to understand the spin-charge conversion from the α -Sn TI [see Fig. 5.14a]. In this spin-charge conversion study, the spin current is generated through the magnetization dynamics induced at the magnetization resonance of a Fe layer by an external rf field. The generated spin current then diffuses to the α -Sn top surface. Injection of a spin current into a TI induces a spin accumulation on one side of the Fermi contour of the Dirac cone as well as a spin depletion on the other side [see Fig. 5.14b]. As a consequence, this spin injection results in a charge current. Importantly, note that the spin-charge conversion is not observed when Fe is deposited directly on α -Sn, but is observed when a thin Ag layer is inserted at the interface [see Fig. 5.14c]. By performing ARPES measurement to probe the DOS, it has been shown that the deposition of a sub-monolayer of Fe on α -Sn indeed suppresses the Dirac cone, that is a signature of the TI, whereas it is still observable after deposition of a Ag layer [see Fig. 5.14d]. Hence the absence of spin-charge conversion at Fe/ α -Sn interfaces is clearly ascribed to the loss of TI surface states after the Fe deposition. This study shows that characterization of the DOS by ARPES measurements is a very useful and unique technique to understand spin-charge conversion at such spinorbitronic interfaces.

5.2.2.2 Heat-Spin Conversion

Similarly to the Seebeck effect, the spin Seebeck effect describes the generation of a spin voltage from a temperature gradient in a FM conductor or insulator. Longitudinal spin Seebeck effect (LSSE) refers to experiments where the spin current generated is parallel to the temperature gradient. Materials used are FM materials (conducting

Fig. 5.14 **a** Scheme of the device studied for the spin to charge conversion by spin pumping into the topological insulator α -Sn. **b** Illustration of the inverse Edelstein effect. A spin current injected in the topological insulator induces an accumulation of charge for one spin direction inducing a shift of the Fermi contour creating a charge current. **c** Ferromagnetic resonance and dc charge current signals for α -Sn/Fe/Au and α -Sn/Ag/Fe/Au samples. Only the second sample shows a dc current signal. **d** ARPES measurement along the [100] or [110] directions on the free surface of α -Sn (top) and after deposition of Fe (left) or Ag (right). The Dirac cone subsists after deposition of Ag but not for Fe. Adapted from [45] with permission (Copyright 2016, American Physical Society)

or insulating) in contact with heavy metals. In order to extract LSSE efficiency, it is necessary to discard the possible artefacts such as an anomalous Nernst effect that arises from the potential induced magnetization in the NM layer [46].

Prototype material systems contain a thin film of yttrium iron garnet $\text{Y}_3\text{Fe}_5\text{O}_{12}$ (YIG) as a FM insulator and a thin layer of Pt as high spin-orbit non-FM material. In order to detect and quantify the magnetic proximity effect (MPE) at the YIG/Pt interface, XMCD measurements appear to be the most accurate method. In Fig. 5.15, XMCD measurements at the Pt $L_{2,3}$ edges performed on YIG/Pt samples are shown. Whereas no MPE have been detected for one study [see Fig. 5.15a] [47, 48], a clear XMCD signal has been observed for another experiment [Fig. 5.15b] [49]. This discrepancy can be attributed to different interface qualities. Actually, x-ray absorption near-edge spectra (XANES) have clearly different shapes. The white line intensity (XANES step height) depends on the number of holes in the 5d band and is, therefore, sensitive to the oxidation state of Pt. In the case where no MPE has been detected, the white line intensity is about 1.3 which is similar to the Pt metal value. On the contrary, when MPE is observed, the white line intensity is about 1.45 which is close to what is observed for $\text{PtO}_{1.36}$ samples [50]. These observations mean that the observed MPE is probably due to intermixing at the YIG/Pt interface.

Fig. 5.15 XANES and XMCD recorded at the Pt $L_{2,3}$ edges for YIG/Pt samples. For apparently similar samples a clear XMCD signal is observed for one sample (right), whereas no induced magnetic moment is measured for the other (left). The higher intensity of the step edge observed for the sample exhibiting an XMCD signal (right) meaning an increase of the number of holes in the $5d$ band indicates a possible Pt oxidation. Hence different interface qualities induce different XMCD signals. Reproduced from [48] (left) and from [49] (right) with permission (Copyright 2013, American Physical Society)

This behaviour has been clearly evidenced by further experiments on insulating ferrite/Pt samples. In Fig. 5.16 (left panel), XMCD measurements performed at the Pt M_3 edge on CoFe₂O₄/Pt samples are presented [51]. A clear XMCD is observed for samples where Pt has been deposited at high temperature (HT in Fig. 5.16), whereas no XMCD signal is detected when Pt is grown at room temperature (RT in Fig. 5.16). By performing XAS and XMCD measurements at Co and Fe $L_{2,3}$ edges [see Fig. 5.16 (right panel)] it appears clearly that the deposition of Pt at high temperature induces some intermixing at the CoFe₂O₄/Pt interface. Indeed, Co XAS spectra are close to those of metallic Co films for the sample with Pt grown at high temperature, whereas for the sample with Pt deposited at room temperature, spectra shapes are similar to those of CoFe₂O₄ thin films. Hence the presence of MPE in ferrite/Pt interface comes from interface alloying, whereas for clean interfaces, the induced magnetization by proximity effect is absent. In fact, XAS is not the only measurement to detect MPE; X-ray resonant magnetic reflectivity is also a powerful technique. Thanks to the latter, T. Kuschel et al. have confirmed that no MPE occurs at the clean NiFe₂O₄/Pt interface [52]. Finally, it has been shown that no MPE occurs for another ferrite (MnFe₂O₄/Pt) and for magnetite, both in the conducting and insulating states [53]. This suggests that the absence of MPE at the magnetic oxide/Pt clean interface is a general rule.

Fig. 5.16 XAS and XMCD spectra recorded at the Pt M_3 edge for $\text{CoFe}_2\text{O}_4/\text{Pt}$ where Pt is grown at room temperature (RT) and high temperature (HT). A clear XMCD signal is observed at the Pt M_3 edge when Pt is grown at high temperature. A clear difference can be also seen in the Fe and Co $L_{2,3}$ edges. For Pt grown at HT Co $L_{2,3}$ spectra look like Co metal signifying an intermixing at the $\text{CoFe}_2\text{O}_4/\text{Pt}$ interface. When Pt is grown at RT, Co and Fe $L_{2,3}$ XAS spectra are similar to those of CoFe_2O_4 thin films. Adapted from [51] with permission (Copyright 2018, American Chemical Society)

These experiments hence highlight that synchrotron radiation-based spectroscopies are powerful to probe the electronic and magnetic properties of interfaces and can be very useful to discard artefacts in LSSE experiments.

5.2.3 Current-Induced Magnetization Dynamics

Besides STT-MRAMs, other conceptual memory devices have been proposed and largely studied in the decade. A flagship example is the domain racetrack memory, proposed by S. S. P. Parkin [54], which is a nanoscale shift register memory in which bits are defined by magnetic domains separated by domain walls. Another version of such devices has been recently introduced where the bits are stored thanks to the presence of magnetic skyrmions [55]. In this section, we show that magnetization dynamics imaging using synchrotron radiation-based measurements that can be useful to better understand these STT-MRAM and racetrack memory devices.

5.2.3.1 Spin–Orbit Torque Driven Magnetization Reversal

SOT is an efficient tool to manipulate magnetization in spintronic devices integrating only one FM electrode (see, for example, R. Ramaswamy et al. for a recent review [56]). M. Baumgartner et al. [57] have studied Co nanodot magnetization reversal by SOT. In Fig. 5.17a, the magnetization reversal [measured by XMCD at the Co L_3 edge (black)] induced by a current pulse (red) is shown. Then, by performing time-resolved scanning transmission X-ray microscopy (STXM) measurements, it has

Fig. 5.17 **a** Magnetization reversal probed by XMCD measurements at Co L_3 edge (black) induced by a current pulse (red). **b** STXM images recorded at intervals of 100 ps during a 2 ns injected current pulse. The four rows correspond to different applied current (yellow arrows) and magnetic field (blue arrows) conditions. Red dots indicate domain wall nucleation and green arrows the direction of the domain wall propagation. Adapted from [57] with permission (Copyright 2017, Nature Publishing Group)

been possible to perform a direct observation of the actual path leading to the magnetization reversal during the current pulse injection. Figure 5.17b represents STXM images recorded at the Co L_3 edge with 25 nm spatial resolution at intervals of 100 ps during a 2 ns current pulse. From these images, nucleation (red dot) and propagation (green arrow) of magnetic domain walls can be clearly observed. Depending on the applied magnetic field direction and the current polarity, the main characteristics of these nucleation and propagation might change. These measurements have allowed demonstrating that this diagonal motion of the domain wall originates from a combination of the damping-like and field-like SOTs and the Dzyaloshinskii–Moriya interaction.

5.2.3.2 Current-Induced Domain Walls and Skyrmions Motion

Racetrack memory-based on domain wall or skyrmion motion is an attractive field for future magnetic memories. Synchrotron radiation-based techniques such as PEEM and STXM or even X-ray magnetic resonant scattering (XMRS) have allowed to study the fine structures of magnetic textures such as chiral domain walls [58], vortex [59] and skyrmions [60, 61], see, for example, X. Cheng et al. for a review [62]. A key point is also to optimize the motion velocity and determine the factors that could

Fig. 5.18 (left panel) XMCD images of Co and NiFe electrodes of a NiFe/Cu/Co spin valves recorded at the Co L_3 edge **a** and Fe L_3 edge **b**, **c**. Different magnetic domains are observed depending on the parallel **b** or antiparallel **c** magnetic configuration (right panel). **d** Sketch of the Pt/Co/Ta stripe deposited on a Si_3N_4 membrane. **e** STXM images recorded after current pulses showing that three of the four skyrmions move (orange, yellow and red circles) after the current pulse. The fourth skyrmion (white circle) is pinned. **f** Skyrmion velocity as a function current density injected for Pt/Co/Ta and Pt/CoFeB/MgO devices. from [63] (left panel) with permission (Copyright 2010, American Physical Society) and from [29] (right panel) with permission (Copyright 2016, Nature Publishing Group)

limit this velocity. X-ray imaging is a powerful tool to investigate magnetic object motion induced by a current. In the following, we show two examples of domain wall and skyrmion motions probed by XMCD-PEEM and STXM.

Large domain wall velocities ($\sim 600 \text{ m s}^{-1}$) have been measured in NiFe/Cu/Co spin valves [63]. Unexpectedly, the domain wall motion is altered when longer electrical pulses or higher current densities are applied [63]. By performing XMCD-PEEM measurements, it has been shown that the dipolar interaction between the NiFe and Co electrodes was a source of domain wall pinning. Thanks to the chemical selectivity of X-ray photoemission, magnetic configuration of the NiFe and Co electrodes can be probed by recording images at the Fe and Co L_3 edges [see Fig. 5.18a, c]. In the parallel magnetic configuration, the stray field of the Co domain wall locally reverses the magnetization in the NiFe layer, leading to the three domain walls [white circle in Fig. 5.18b]. On the other hand, in the antiparallel magnetic configuration, the magnetic flux closes naturally and a single domain wall is formed [Fig. 5.18c]. This magnetic imaging of the Co and NiFe electrodes demonstrates that the stray field prevents domain wall motion across the corners in the NiFe layer.

An important property of the racetrack memories is that the magnetic objects (domain walls or skyrmions) move all together. In Fig. 5.18(d-f), we show that STXM measurements can be used to probe current-driven skyrmion motion in a Pt/Co/Ta stripe as shown by K. Woo *et al.* [29]. Each image is recorded after injection of current pulses. Three of the four skyrmions (red, yellow and orange circles) move forward and backward depending on the current polarity [see Fig. 5.18e]. Note that the fourth skyrmion (white circle) is not showing any motion under current injection

(for both polarities) and is probably pinned by a material defect. Skyrmion velocities can be extracted [see Fig. 5.18f] and it is shown that they increase from $\sim 50 \text{ m s}^{-1}$ to $\sim 120 \text{ m s}^{-1}$ by replacing Co by CoFeB. These measurements allow to conclude that lower pinning materials such as amorphous CoFeB layer, i.e. without grain boundaries, are probably more appropriate to get an efficient skyrmion motion.

X-ray microscopy is thus a powerful tool to study in real time the current-driven magnetization dynamics in various magnetic systems. We have shown three examples where magnetic reversal, domain walls and skyrmion motion are investigated by STXM or XMCD-PEEM. This is not limited to FM systems as antiferromagnetic domains can be also probed by performing linear dichroism. Recently, M. J. Grzybowski et al. have measured current-induced antiferromagnetic domain switching in CuMnAs by performing XMLD-PEEM measurements [64].

5.3 Conclusion

We have shown through few examples that synchrotron radiation-based spectroscopies are powerful techniques to perform advanced characterization of various spintronic systems. For example, XAS (including circular and linear dichroism) allows probing the electronic and magnetic properties of surfaces, interfaces and bulk materials. Spin- and angle-resolved photoemission is an ideal tool to measure the actual spin polarization and reveal the density of states of surfaces and interfaces. It can be applied to a wide variety of materials used in spintronic devices such as ferromagnets, antiferromagnets, topological insulators, Rashba interfaces, hybrid ferromagnet/molecules interfaces, 2D materials and multiferroics. XPEEM and STXM are also perfect techniques to study the dynamics of magnetic structures (ferromagnetic, ferrimagnetic and antiferromagnetic) but also to measure locally the electronic structure of surfaces. This is not exhaustive; other techniques such as X-ray magnetic scattering, resonant inelastic X-ray scattering, hard X-ray photoelectron spectroscopy, nano-ARPES can be also used to probe spintronic device properties. Finally, the development of new techniques such as magnetic X-ray nanotomography [65] and novel X-ray sources (new generation of synchrotron and X-ray free-electron lasers) improving coherence, spatial and temporal resolution, will allow obtaining deeper characterization and understanding of magnetic textures and spintronic devices.

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References

1. N.F. Mott, The electrical conductivity of transition metals. Proc. R. Soc. Lond., Ser. A **153**, 699 (1936). <https://doi.org/10.1098/rspa.1936.0031>
2. A. Fert, I. Campbell, Two-current conduction in nickel. Phys. Rev. Lett. **21**, 1190 (1968). <https://doi.org/10.1103/PhysRevLett.21.1190>
3. A. Fert, I. Campbell, Transport properties of ferromagnetic transition metals. J. Phys. **32** C1, 46 (1971). <https://doi.org/10.1051/jphyscol:1971109>
4. A. Fert, I. Campbell, Electrical resistivity of ferromagnetic nickel and iron based alloys. J. Phys. F: Metal Phys. **6**, 849 (1976). <https://doi.org/10.1088/0305-4608/6/5/025>
5. A. Fert, Nobel lecture: origin, development, and future of spintronics. Rev. Mod. Phys. **80**, 1517 (2008). <https://doi.org/10.1103/RevModPhys.80.1517>
6. P. Grünberg, R. Schreiber, Y. Pang, M. Brodsky, H. Sowers, Layered magnetic structures: evidence for antiferromagnetic coupling of Fe layers across Cr interlayers. Phys. Rev. Lett. **57**, 2442 (1986). <https://doi.org/10.1103/PhysRevLett.57.2442>
7. B. Dieny, V.S. Speriosu, S.S.P. Parkin, B.A. Gurney, D.R. Wilhoit, D. Mauri, Giant magnetoresistive in soft ferromagnetic multilayers. Phys. Rev. B **43**, 1297 (1991). <https://doi.org/10.1103/PhysRevB.43.1297>
8. A. Barthélémy, A. Fert, F. Petroff, Giant magnetoresistance in magnetic multilayers, in *Handbook of Magnetic Materials*, ed. by K.H.J. Buschow, vol. 12 (North-Holland, Amsterdam, 1999), p. 1. [https://doi.org/10.1016/S1567-2719\(99\)12005-5](https://doi.org/10.1016/S1567-2719(99)12005-5)
9. M.N. Baibich, J.M. Broto, A. Fert, F.N. Van Dau, F. Petroff, P. Etienne, G. Creuzet, A. Friederich, J. Chazelas, Giant magnetoresistance of (001)Fe/(001)Cr magnetic superlattices. Phys. Rev. Lett. **61**, 2472 (1988). <https://doi.org/10.1103/PhysRevLett.61.2472>
10. G. Binasch, P. Grünberg, F. Saurenbach, W. Zinn, Enhanced magnetoresistance in layered magnetic structures with antiferromagnetic interlayer exchange. Phys. Rev. B **39**, 4828 (1989). <https://doi.org/10.1103/PhysRevB.39.4828>
11. T. Valet, A. Fert, Theory of the perpendicular magnetoresistance in magnetic multilayers. Phys. Rev. B **48**, 7099 (1993). <https://doi.org/10.1103/PhysRevB.48.7099>
12. M. Jullière, Tunneling between ferromagnetic films. Phys. Lett. A **54**, 225 (1975). [https://doi.org/10.1016/0375-9601\(75\)90174-7](https://doi.org/10.1016/0375-9601(75)90174-7)
13. J.S. Moodera, L.R. Kinder, T.M. Wong, R. Meservey, Large magnetoresistance at room temperature in ferromagnetic thin film tunnel junctions. Phys. Rev. Lett. **74**, 3273 (1995). <https://doi.org/10.1103/PhysRevLett.74.3273>
14. T. Miyazaki, N. Tezuka, Giant magnetic tunneling effect in Fe/Al₂O₃/Fe junction. J. Magn. Magn. Mater. **139**, L231 (1995). [https://doi.org/10.1016/0304-8853\(95\)90001-2](https://doi.org/10.1016/0304-8853(95)90001-2)
15. M. Bowen, V. Cros, F. Petroff, A. Fert, C. Martínez Boubeta, J.L. Costa-Krämer, J.V. Anguita, A. Cebollada, F. Briones, J.M. de Teresa, L. Morellón, M.R. Ibarra, F. Güell, F. Peiró, A. Cornet, Large magnetoresistance in Fe/MgO/FeCo(001) epitaxial tunnel junctions on GaAs(001). Appl. Phys. Lett. **79**, 1655 (2001). <https://doi.org/10.1063/1.1404125>
16. J. Faure-Vincent, C. Tiusan, E. Jouguelet, F. Canet, M. Sajieddine, C. Bellouard, E. Popova, M. Hehn, F. Montaigne, A. Schuhl, High tunnel magnetoresistance in epitaxial Fe/MgO/Fe tunnel junctions. Appl. Phys. Lett. **82**, 4507 (2003). <https://doi.org/10.1063/1.1586785>
17. W.H. Butler, X.-G. Zhang, T.C. Schulthess, J.M. MacLaren, Spin-dependent tunneling conductance of Fe|MgO|Fe sandwiches. Phys. Rev. B **63**, 054416 (2001). <https://doi.org/10.1103/PhysRevB.63.054416>
18. S. Yuasa, D. Djayaprawira, Giant tunnel magnetoresistance in magnetic tunnel junctions with a crystalline MgO(001) Barrier. J. Phys. D: Appl. Phys. **40**, R337 (2007). <https://doi.org/10.1088/0022-3727/40/21/R01>
19. S. Ikeda, J. Hayakawa, Y. Ashizawa, Y.M. Lee, K. Miura, H. Hasegawa, M. Tsunoda, F. Matsukura, H. Ohno, Tunnel magnetoresistance of 604% at 300 K by suppression of Ta diffusion in CoFeB/MgO/CoFeB pseudo-spin-valves annealed at high temperature. Appl. Phys. Lett. **93**, 082508 (2008). <https://doi.org/10.1063/1.2976435>

20. S. Yuasa, T. Nagahama, A. Fukushima, Y. Suzuki, K. Ando, Giant room-temperature magnetoresistance in single-crystal Fe/MgO/Fe magnetic tunnel junctions. *Nat. Mater.* **3**, 868 (2004). <https://doi.org/10.1038/nmat1257>
21. S.S.P. Parkin, C. Kaiser, A. Panchula, P.M. Rice, B. Hughes, M. Samant, S.-H. Yang, Giant tunnelling magnetoresistance at room temperature with MgO(100) tunnel barriers. *Nat. Mater.* **3**, 862 (2004). <https://doi.org/10.1038/nmat1256>
22. J.C. Slonczewski, Current-driven excitation of magnetic multilayers. *J. Magn. Magn. Mater.* **159**, L1 (1996). [https://doi.org/10.1016/0304-8853\(96\)00062-5](https://doi.org/10.1016/0304-8853(96)00062-5)
23. L. Berger, Emission of spin waves by a magnetic multilayer traversed by a current. *Phys. Rev. B* **54**, 9353 (1996). <https://doi.org/10.1103/PhysRevB.54.9353>
24. M. Tsoi, A.G.M. Jansen, J. Bass, W.-C. Chiang, M. Seck, V. Tsoi, P. Wyder, Excitation of a magnetic multilayer by an electric current. *Phys. Rev. Lett.* **80**, 4281 (1998). [Erratum: *Phys. Rev. Lett.* **81**, 493 (1998).] <https://doi.org/10.1103/PhysRevLett.80.4281>, <https://doi.org/10.1103/PhysRevLett.81.493>
25. J.A. Katine, F.J. Albert, R.A. Buhrman, E.B. Myers, D.C. Ralph, Current-driven magnetization reversal and spin-wave excitations in Co/Cu/Co pillars. *Phys. Rev. Lett.* **84**, 3149 (2000). <https://doi.org/10.1103/PhysRevLett.84.3149>
26. Z. Diao, D. Apalkov, M. Pakala, Y. Ding, A. Panchula, Y. Huai, Spin transfer switching and spin polarization in magnetic tunnel junctions with MgO and AlO_x barriers. *Appl. Phys. Lett.* **87**, 232502 (2005). <https://doi.org/10.1063/1.2139849>
27. J. Grollier, D. Lacour, V. Cros, A. Hamzic, A. Vaures, A. Fert, D. Adam, G. Faini, Switching the magnetic configuration of a spin valve by current-induced domain wall motion. *J. Appl. Phys.* **92**, 4825 (2002). <https://doi.org/10.1063/1.1507820>
28. A. Dussaux, B. Georges, J. Grollier, V. Cros, A. Khvalkovskiy, A. Fukushima, M. Konoto, H. Kubota, K. Yakushiji, S. Yuasa, K.A. Zvezdin, K. Ando, A. Fert, Large microwave generation from current-driven magnetic vortex oscillators in magnetic tunnel junctions. *Nat. Commun.* **1**, 8 (2010). <https://doi.org/10.1038/ncomms1006>
29. S. Woo, K. Litzius, B. Krüger, M.-Y. Im, L. Caretta, K. Richter, M. Mann, A. Krone, R.M. Reeve, M. Weigand, P. Agrawal, I. Lemesh, M.-A. Mawass, P. Fischer, M. Kläui, G.S.D. Beach, Observation of room-temperature magnetic skyrmions and their current-driven dynamics in ultrathin metallic ferromagnets. *Nat. Mater.* **15**, 501 (2016). <https://doi.org/10.1038/nmat4593>
30. W. Legrand, D. Maccariello, N. Reyren, K. Garcia, C. Moutafis, C. Moreau-Luchaire, S. Collin, K. Bouzehouane, V. Cros, A. Fert, Room-temperature current-induced generation and motion of sub-100 nm skyrmions. *Nano Lett.* **17**, 2703 (2017). <https://doi.org/10.1021/acs.nanolett.7b00649>
31. C. Song, B. Cui, F. Li, X. Zhou, F. Pan, Recent progress in voltage control of magnetism: materials, mechanisms, and performance. *Prog. Mater. Sci.* **87**, 33 (2017). <https://doi.org/10.1016/j.pmatsci.2017.02.002>
32. H. Ohno, D. Chiba, F. Matsukura, T. Omiya, E. Abe, T. Dietl, Y. Ohno, K. Ohtani, Electric-field control of ferromagnetism. *Nature* **408**, 944 (2000). <https://doi.org/10.1038/35050040>
33. T. Dietl, H. Ohno, Dilute ferromagnetic semiconductors: physics and spintronic structures. *Rev. Mod. Phys.* **86**, 187 (2014). <https://doi.org/10.1103/RevModPhys.86.187>
34. M. Weisheit, S. Fähler, A. Marty, Y. Souche, C. Poinsignon, D. Givord, Electric field-induced modification of magnetism in thin-film ferromagnets. *Science* **315**, 349 (2007). <https://doi.org/10.1126/science.1136629>
35. J. Lou, M. Liu, D. Reed, Y. Ren, N.X. Sun, Giant electric field tuning of magnetism in novel multiferroic FeGaB/lead zinc niobate-lead titanate (PZN-PT) heterostructures. *Adv. Mater.* **21**, 4711 (2009). <https://doi.org/10.1002/adma.200901131>
36. A. Soumyanarayanan, N. Reyren, A. Fert, C. Panagopoulos, Emergent phenomena induced by spin-orbit coupling at surfaces and interfaces. *Nature* **539**, 509 (2016). <https://doi.org/10.1038/nature19820>
37. A. Manchon, C. Ducruet, L. Lombard, S. Auffret, B. Rodmacq, B. Dieny, S. Pizzini, J. Vogel, V. Uhlir, M. Hochstrasser, G. Panaccione, Analysis of oxygen induced anisotropy crossover in Pt/Co/MOx trilayers. *J. Appl. Phys.* **104**, 043914 (2008). <https://doi.org/10.1063/1.2969711>

38. D. Chiba, S. Fukami, K. Shimamura, N. Ishiwata, K. Kobayashi, T. Ono, Electrical control of the ferromagnetic phase transition in cobalt at room temperature. *Nat. Mater.* **10**, 853 (2011). <https://doi.org/10.1038/nmat3130>
39. C. Bi, Y. Liu, T. Newhouse-Illige, M. Xu, M. Rosales, J.W. Freeland, O. Mryasov, S. Zhang, G.E. te Velthuis, W.G. Wang, Reversible control of Co magnetism by voltage-induced oxidation. *Phys. Rev. Lett.* **113**, 267202 (2014). <https://doi.org/10.1103/PhysRevLett.113.267202>
40. F. Bonell, Y.T. Takahashi, D.D. Lam, S. Yoshida, Y. Shiota, S. Miwa, T. Nakamura, Y. Suzuki, Reversible change in the oxidation state and magnetic circular dichroism of Fe driven by an electric field at the FeCo/MgO interface. *Appl. Phys. Lett.* **102**, 152401 (2013). <https://doi.org/10.1063/1.4802030>
41. S. Miwa, K. Matsuda, K. Tanaka, Y. Kotani, M. Goto, T. Nakamura, Y. Suzuki, Voltage-controlled magnetic anisotropy in Fe|MgO tunnel junctions studied by X-ray absorption spectroscopy. *Appl. Phys. Lett.* **107**, 162402 (2015). <https://doi.org/10.1063/1.4934568>
42. R.V. Chopdekar, V.K. Malik, A. Fraile Rodríguez, L. Le Guyader, Y. Takamura, A. Scholl, D. Stender, C.W. Schneider, C. Bernhard, F. Nolting, L.J. Heyderman, Spatially resolved strain-imprinted magnetic states in an artificial multiferroic. *Phys. Rev. B* **86**, 014408 (2012). <https://doi.org/10.1103/PhysRevB.86.014408>
43. Y. Otani, M. Shiraishi, A. Oiwa, E. Saitoh, S. Murakami, Spin conversion on the nanoscale. *Nat. Phys.* **13**, 829 (2017). <https://doi.org/10.1038/nphys4192>
44. J. Sinova, S.O. Valenzuela, J. Wunderlich, C.H. Back, T. Jungwirth, Spin Hall effects. *Rev. Mod. Phys.* **87**, 1213 (2015). <https://doi.org/10.1103/RevModPhys.87.1213>
45. J.C. Rojas-Sánchez, S. Oyarzún, Y. Fu, A. Marty, C. Vergnaud, S. Gambarelli, L. Vila, M. Jamet, Y. Ohtsubo, A. Taleb-Ibrahimi, P. Le Fèvre, F. Bertran, N. Reyren, J.M. George, A. Fert, Spin to charge conversion at room temperature by spin pumping into a new type of topological insulator: α -Sn films. *Phys. Rev. Lett.* **116**, 096602 (2016). <https://doi.org/10.1103/PhysRevLett.116.096602>
46. S.Y. Huang, X. Fan, D. Qu, Y.P. Chen, W.G. Wang, J. Wu, T.Y. Chen, J.Q. Xiao, C.L. Chien, Transport magnetic proximity effects in platinum. *Phys. Rev. Lett.* **109**, 107204 (2012). <https://doi.org/10.1103/PhysRevLett.109.107204>
47. S. Geprägs, S. Meyer, S. Altmannshofer, M. Opel, F. Wilhelm, A. Rogalev, R. Gross, S.T. Goennenwein, Investigation of induced Pt magnetic polarization in Pt/Y₃Fe₅O₁₂ bilayers. *Appl. Phys. Lett.* **101**, 262407 (2012). <https://doi.org/10.1063/1.4773509>
48. S. Geprägs, S.T.B. Goennenwein, M. Schneider, F. Wilhelm, K. Ollefs, A. Rogalev, M. Opel, R. Gross, Comment on “Pt magnetic polarization on Y₃Fe₅O₁₂ and magnetotransport characteristics” (2013), [arXiv:1307.4869v1](https://arxiv.org/abs/1307.4869v1) [cond-mat.mtrl-sci]
49. Y.M. Lu, Y. Choi, C.M. Ortega, X.M. Cheng, J.W. Cai, S.Y. Huang, L. Sun, C.L. Chien, Pt magnetic polarization on Y₃Fe₅O₁₂ and magnetotransport characteristics. *Phys. Rev. Lett.* **110**, 147207 (2013). <https://doi.org/10.1103/PhysRevLett.110.147207>
50. A.V. Kolobov, F. Wilhelm, A. Rogalev, T. Shima, J. Tominaga, Thermal decomposition of sputtered thin PtO_x layers used in super-resolution optical disks. *Appl. Phys. Lett.* **86**, 121909 (2005). <https://doi.org/10.1063/1.1886255>
51. H.B. Vasili, M. Gamino, J. Gàzquez, F. Sánchez, M. Valvidares, P. Gargiani, E. Pellegrin, J. Fontcuberta, Magnetoresistance in hybrid Pt/CoFe₂O₄ bilayers controlled by competing spin accumulation and interfacial chemical reconstruction. *ACS Appl. Mater. Interfaces* **10**, 12031 (2018). <https://doi.org/10.1021/acsami.8b00384>
52. Static magnetic proximity effect in Pt/NiFe₂O₄ and Pt/Fe bilayers investigated by X-ray resonant magnetic reflectivity. *Phys. Rev. Lett.* **115**, 097401 (2015). <https://doi.org/10.1103/PhysRevLett.115.097401>
53. M. Collet, R. Mattana, J.-B. Moussy, K. Ollefs, S. Collin, C. Deranlot, A. Anane, V. Cros, F. Petroff, F. Wilhelm, A. Rogalev, Investigating magnetic proximity effects at ferrite/Pt interfaces. *Appl. Phys. Lett.* **111**, 202401 (2017). <https://doi.org/10.1063/1.4987145>
54. S.S.P. Parkin, M. Hayashi, L. Thomas, Magnetic domain-wall racetrack memory. *Science* **320**, 190 (2008). <https://doi.org/10.1126/science.1145799>

55. A. Fert, V. Cros, J. Sampaio, Skyrmions on the track. *Nat. Nanotechnol.* **8**, 152 (2013). <https://doi.org/10.1038/nnano.2013.29>
56. R. Ramaswamy, J.M. Lee, K. Cai, H. Yang, Recent advances in spin-orbit torques: moving towards device applications. *Appl. Phys. Rev.* **5**, 031107 (2018). <https://doi.org/10.1063/1.5041793>
57. M. Baumgartner, K. Garello, J. Mendil, C.O. Avci, E. Grimaldi, C. Murer, J. Feng, M. Gabureac, C. Stamm, Y. Acremann, S. Finizio, S. Wintz, J. Raabe, P. Gambardella, Spatially and time-resolved magnetization dynamics driven by spin-orbit torques. *Nat. Nanotechnol.* **12**, 980 (2017). <https://doi.org/10.1038/nnano.2017.151>
58. J.-Y. Chaudreau, W. Legrand, N. Reyren, D. Maccariello, S. Collin, H. Popescu, K. Bouzehouane, V. Cros, N. Jaouen, A. Fert, Chirality in magnetic multilayers probed by the symmetry and the amplitude of dichroism in X-ray resonant magnetic scattering. *Phys. Rev. Lett.* **120**, 037202 (2018). <https://doi.org/10.1103/PhysRevLett.120.037202>
59. T. Taniuchi, M. Oshima, H. Akinaga, K. Ono, Vortex-chirality control in mesoscopic disk magnets observed by photoelectron emission microscopy. *J. Appl. Phys.* **97**, 10J904 (2005). <https://doi.org/10.1063/1.1862032>
60. C. Moreau-Luchaire, C. Moutafis, N. Reyren, J. Sampaio, C.A.F. Vaz, N. Van Horne, K. Bouzehouane, K. Garcia, C. Deranlot, P. Warnicke, P. Wohlhüter, J.-M. George, M. Weigand, J. Raabe, V. Cros, A. Fert, Additive interfacial chiral interaction in multilayers for stabilization of small individual skyrmions at room temperature. *Nat. Nanotechnol.* **11**, 444 (2016). <https://doi.org/10.1038/nnano.2015.313>
61. O. Boulle, J. Vogel, H. Yang, S. Pizzini, D. de Souza Chaves, A. Locatelli, T.O. Mendes, A. Sala, L.D. Buda-Prejbeanu, O. Klein, M. Belmeguenai, Y. Roussigné, A. Stashkevich, S.M. Chérif, L. Aballe, M. Foerster, M. Chshiev, S. Auffret, I.M. Miron, G. Gaudin, Room-temperature chiral magnetic skyrmions in ultrathin magnetic nanostructures. *Nat. Nanotechnol.* **11**, 449 (2016). <https://doi.org/10.1038/nnano.2015.315>
62. X. Cheng, D. Keavney, Studies of nanomagnetism using synchrotron-based X-ray photoemission electron microscopy (X-PEEM). *Rep. Prog. Phys.* **75**, 026501 (2012). <https://doi.org/10.1088/0034-4885/75/2/026501>
63. V. Uhlíř, S. Pizzini, N. Rougemaille, J. Novotný, V. Cros, E. Jiménez, G. Faini, L. Heyne, F. Sirotti, C. Tieg, A. Bendounan, F. Maccherozzi, R. Belkhou, J. Grollier, A. Anane, J. Vogel, Current-induced motion and pinning of domain walls in spin-valve nanowires studied by XMCD-PEEM. *Phys. Rev. B* **81**, 224418 (2010). <https://doi.org/10.1103/PhysRevB.81.224418>
64. M.J. Grzybowski, P. Wadley, K.W. Edmonds, R. Beardsley, V. Hills, R.P. Campion, B.L. Gallagher, J.S. Chauhan, V. Novak, T. Jungwirth, F. Maccherozzi, S.S. Dhesi, Imaging current-induced switching of antiferromagnetic domains in CuMnAs. *Phys. Rev. Lett.* **118**, 057701 (2017). <https://doi.org/10.1103/PhysRevLett.118.057701>
65. C. Donnelly, M. Guizar-Sicairos, V. Scagnoli, S. Gliga, M. Holler, J. Raabe, L.J. Heyderman, Three-dimensional magnetization structures revealed with X-ray vector nanotomography. *Nature* **547**, 328 (2017). <https://doi.org/10.1038/nature23006>

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