Heusler Compounds and Spintronics

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Abstract

Heusler compounds are a large group of intermetallic compounds with over 1000 members with similar crystal structures having a vast array of tunable properties. These properties depend on the number of valence electrons per formula unit allowing tuning of properties through composition and alloying. The Heusler lattice parameters span many metal oxides and semiconductors and their crystal structures are closely related. For spintronic applications, the magnetic and half-metallic properties, in particular, are of great interest. In this chapter the electronic and magnetic properties of Heusler compounds are discussed as well as the importance of composition and defect control on tailoring their properties. Examples of applications include the great success of Heusler magnetic tunnel junction in metallic spintronic devices. The potential of going beyond metallic spintronics to the integration of Heusler compounds with III-V semiconductors for semiconductor spintronics device physics and technology, the tuning of magnetic properties, and the fabrication of Heusler compound heterostructures and superlattices are also discussed

Key words:

Heusler, full-Heusler, half-Heusler, spintronics, semiconductors, half-metals, magnetism, spin polarization, Schottky barrier, Co_2MnSi , Co_2FeSi , Fe_3Ga , GaAs, NiMnSb, spin transfer torque, magnetic random access memory, MRAM, STT-RAM, MgO, GaAs, tunneling anisotropic magnetoresistance (TAMR), spin valve, magnetic tunnel junction, MTJ, spin polarization, valence electron count (VEC), Slater-Pauling, Spin-LED, thermodynamic stability, lattice mismatch, spin-FET, perpendicular magnetization, non-local voltage, molecular beam epitaxy, MBE, sputtering, density of states, point defects, stoichiometry, reflection high energy electron diffraction, RHEED, magnetic moment

Introduction:

In this chapter the many properties of Heusler compounds will first be introduced, then their successful application in metallic spintronic devices (magnetic tunnel junctions) will be described, followed by a discussion on their applications beyond metallic spintronics and their integration with the well established III-V compound semiconductors and devices. Emphasis will be made on developing a detailed understanding of Heusler properties and how to integrate them with other materials, such as III-V compound semiconductors.

Heusler compounds are a large group of intermetallic compounds (Fig. 1) with ~1500 members that have a wide range of material properties[1] and can be thought of as the metallic equivalents to the Perovskite complex metal oxides. With such a large number of combinations of elements, many compounds have only recently been predicted theoretically to be thermodynamically stable and have yet to be experimentally verified[2,3]. In general, the Heusler compounds have face centered cubic crystal structures, which are either the full-Heusler or the half-Heusler structures (Fig. 2) with lattice parameters ranging from ~5.6Å to 7.2Å. The full-Heusler compound structures (Fig. 2) can be viewed as ordered B2 or stuffed NaCl crystal structures of YZ with every tetrahedral site occupied by X. The half-Heusler compounds can be thought of as having only half of the NaCl tetrahedral sites filled by X or as an XZ zincblende lattice with the octahedral sites filled with Y. The half-Heusler compound CoTiSb, which is a semiconductor, may be thought of as a zincblende sp^3 bonded CoSb⁴⁻ lattice with Ti⁴⁺ in the octahedral sites. The Heusler compound electronic properties are predicted to depend on the number of valence electrons per formula unit (VEC) [1,4-9]. Half-Heuslers with 17 or 19 VEC are predicted to be ferromagnetic (CoTiSn, CoVSb,...)[10], with 22 VEC half-metallic (NiMnSb,...)[6,11], and 8 or 18 VEC semiconductors (CoTiSb, NiTiSn,...)[1,7,8,12]. For the 18, 19 and 22 VEC half-Heusler compounds, covalent hybridization between the lower-energy d-states of the high valence transition metal (X) and the higher-energy d states of the lower valence transition metal atom (Y), leads to the formation of bonding and anti-bonding bands with a gap ($X d^n$, $n=0, 1, 4$ for 18, 19 and 22 VEC, $Y d^{10}$). The sp-elements (Z), give rise to one s and three p bands below the center of the d-bands and can accommodate a total of 8 electrons (s^2p^6) [4,6].

Full-Heuslers with 24 VEC are predicted to be semiconductors[5,13,14] and 26-29 can be superconductors (Pd₂HfAl,...)[15-19]. Some semiconducting Heuslers have also been predicted to be piezoelectric[20]. The dependence on VEC suggests that their properties can be tuned through alloying with elements to increase or decrease the number of electrons so as to adjust the position of the Fermi level. However, their properties and potential applications go far beyond what band gap engineering has enabled for conventional compound semiconductors [1].

For Spintronic application, the magnetic Heusler compounds and half-metals (Fig. 3), in particular, are of must interest. A number of review articles and books have been devoted to the magnetic and half-metal Heusler compounds[21-23]. The free carrier spin polarization at the Fermi level is an important property and is given by:

$$P(E_f) = D^\uparrow(E_f) - D^\downarrow(E_f) / D^\uparrow(E_f) + D^\downarrow(E_f) \quad (1)$$

where $D^\uparrow$ and $D^\downarrow$ are the spin-up and spin-down carrier density of states at the Fermi level. In half-metals the Fermi level is in the band gap for one spin (semiconducting) and in the band for

the other (metallic). Hence the material will be 100% spin polarized at the Fermi level (Fig. 3). This makes half-metals ideal for spintronic applications, as will be discussed below.

Figure 4 shows the calculated density of states for a number of half-Heusler compounds with 17, 18 and 19 VEC[24]. By increasing the electron count from 17 to 19 the Fermi level moves from the valence band to the conduction band and the properties change from magnetic to non-magnetic and back to magnetic. The magnetization depends on the number of magnetic moments per formula unit. For 3d metals Slater-Pauling behavior indicates that the magnetic moment per atom can be estimated from the average number of electrons per atom[25,26]. The total number of valence electrons per formula unit, Z_t is given by

$$Z_t = N_{v\uparrow} + N_{v\downarrow} \quad (2)$$

where $N_{v\uparrow}$ and $N_{v\downarrow}$ are the number of spin-up and spin-down electrons, respectively. The magnetic moment is the difference in the number of spin-up and spin-down electrons:

$$M_t = N_{v\uparrow} - N_{v\downarrow} \quad (3)$$

Half-Heuslers with 9 minority bands filled with 2 electrons in each band gives the “rule of 18”[6]

$$M_t = Z_t - 18 \quad (4)$$

Figure 5 shows a plot of magnetic moment versus Z_t for a number of half-Heusler compounds. For the full-Heusler compounds, the “rule of 24” applies[5] (Fig. 6). For the inverse Heusler compounds, in which the X and Y atoms are switched[1], the rules of 18, 24 and 28 apply[27] (see Fig. 7).

For a half-metal the minority spin bands are completely filled and the additional electrons must be in the majority band. This means that for a half-metal the magnetic moment per formula unit must be an integer number of Bohr magnetons, μ_B . The solid points in Figs. 5-7 indicate predicted half metallic compounds. The Slater-Pauling behavior of Heusler compounds allows one to make a direct comparison to conventional magnetic materials. Figure 8 shows the Slater-Pauling curve for conventional magnetic alloys together with a number of full-Heusler compounds[1] and Fig. 9 a plot of Curie temperature versus VEC. An extensive review of half-metal Heusler compounds can be found in reference [21].

A number of half-Heusler[6,11,28], full-Heusler[29-35] and inverse full-Heusler compounds are predicted to be half-metals[1,22,36,37]. Figure 10 shows the calculated band structure for the half-metal full-Heusler Co_2MnSi [38]. As can be seen in this figure, there are no minority spin states at the Fermi level. The majority spin band is metallic and the Fermi level is predicted to be in the middle of the minority spin band gap. However, defects can have a dramatic influence on the half-metallic properties. For the half-metallic half-Heusler compound NiMnSb Orgassa et al.[39] predicted that disorder (interchange of Ni and Mn or vacancies) would result in minority states at the Fermi level, dramatically decreasing the spin polarization. Similarly for Co_2MnSi and Co_2MnGe , Picozzi et al.[40,41] found that the Mn antisite had the lowest formation energy, but still resulted in half-metallicity. On the other hand, Co antisites, which have slightly higher formation energies, have a dramatic effect on the half-metallicity. Mn-Si and Mn-Co swaps have higher formation energy and keep the half-metallicity. Similar results were obtained by Özdogan and Galanakis[42] for Co_2MnSi and Co_2MnAl . Mn-Si

disorder retains the half-metallicity, but changes the crystal structure from the $L2_1$ ordered Heusler structure to the disordered B2 crystal structure (Fig. 2). In order to enhance $L2_1$ ordering, Heusler films are often annealed following deposition.

The surface properties of Heusler films may be very different from their bulk properties and will depend on the atomic termination[36 ,43-46]. Using argon ion sputter cleaning of the surface of 600°C annealed room temperature magnetron sputter deposited $\text{Co}_2\text{MnSi}(001)$ films on $\text{MgO}(001)$ and spin-resolved photoemission spectroscopy, Wüstenberg et al.[47] reported that for Mn-poor films ($\text{Co}_2\text{Mn}_{0.69}\text{Si}_{1.01}$) the spin polarization at the Fermi energy (E_f) was negative due to the presence of Co_{Mn} antisites and minority surface state contributions. In Mn-rich compounds, $\text{Co}_2\text{Mn}_{1.19}\text{Si}_{0.88}$, they reported that the suppression of Co_{Mn} antisites leads to a positive spin polarization at E_f , and the influence of minority surface states on the photoelectron spin polarization is reduced. An in-situ spin-resolved UV photoemission study of radio frequency (RF) sputter deposited at room temperature and in-situ 550°C annealed $\text{Co}_2\text{MnSi}/\text{MgO}(001)$ films yielded a 93% surface spin polarization at room temperature[48]. A more detailed in-situ spin polarized photoemission study of molecular beam epitaxy (MBE) grown and ~750°C annealed $\text{Co}_2\text{Mn}_{1+\delta}\text{Si}/\text{MgO}(001)$ films by Andrieu et al.[46] showed that ~100% spin polarization occurred below the Fermi level for stoichiometric films ($\delta=0$) (Figs. 11 and 12). This indicates that the Fermi level is above the optimum position in the gap for the minority states in the bulk as well as at the surface (Fig. 11). They demonstrated that the Fermi level could be tuned to enhance the spin polarization at the Fermi level (Fig. 11) by increasing the Mn concentration ($\delta>0$) (VEC tuning) or by one monolayer deposition of Mn, MnSi, or MgO (Fig. 12). These results demonstrate that bulk composition and surface termination have a strong influence on the spin polarization.

In addition to the surface, the spin polarization at an interface is also predicted to depend on the atomic structure of the interface[41,49]

Heusler Device Structures:

Metal Spintronics: Magnetic Tunnel Junction and Magnetic Spin Valve

The concept of a spintronic device is to utilize the spin rather than the charge of the electron. The magnetic tunnel junction (MTJ) and magnetic spin valve (SV) are examples of metallic spintronic devices. These consist of ferromagnet/insulator tunnel barrier/ferromagnet (MTJ) and ferromagnet/normal metal/ferromagnet (SV) structures and are shown in Fig. 13. For the MTJ, transport of electrons from one ferromagnetic contact to the other occurs via tunneling through the insulating barrier. The resistance of spin valves and MTJ devices depends on the relative alignment for the two ferromagnetic layers. The insert in Fig. 13 shows the schematic of the density of states for two half-metallic ferromagnet contacts for the parallel and antiparallel magnetic states. When the two ferromagnetic layers are aligned parallel, there are states for the spin down electrons at the Fermi level to go from the one ferromagnetic layer into the second. When they are antiparallel, the states at the Fermi level have opposite spin for the two ferromagnets; although there are states available in the normal metal, there are no states for the spin down electrons to enter the second ferromagnet without flipping their spin. Hence, in this configuration, the spin valve will be in its high resistance state. This is schematically shown in Fig. 13. In the case of the MTJ, the transport would be by tunneling through the insulating tunnel

barrier. Following the approach of Julliere [50], the tunneling magnetoresistance (TMR) is given by:

$$TMR/V=0=G_P-G_{AP}/G_{AP}=R_{AP}-R_{PP}/R_{PP}=2P'_1P'_2/(1-P'_1P'_2) \quad (5)$$

where G_P , G_{AP} , R_P , and R_{AP} are the tunneling conductance (dI/dV) and tunneling resistance for the parallel and antiparallel configurations. P'_1 and P'_2 are the tunneling spin polarization for the two ferromagnetic contacts, respectively. Note that P'_1 and P'_2 are not the carrier spin polarizations P_1 and P_2 in the ferromagnetic contacts. P'_1 and P'_2 also depend on the electron effective mass, interface bonding and scattering and the spin dependent attenuation in the tunnel barrier. For perfect half-metal contacts $P_1(E_F)$ and $P_2(E_F)$ would be 1 and, assuming no spin flip scattering, this would result in P'_1 and P'_2 also being 1 and the TMR would be infinity.

From the discussion above, it is clear that the measured spin polarization depends upon the measurement technique used and care must be taken when comparing different spin polarization measurements. In general, the spin polarization is given by:

$$P(E_f)n=vf\uparrow nD(E_f)\uparrow -vf\downarrow nD(E_f)\downarrow /vf\uparrow nD(E_f)\uparrow +vf\downarrow nD(E_f)\downarrow \quad (6)$$

where $vf\uparrow n$ and $vf\downarrow n$ are the Fermi velocities for spin up and spin down electrons, respectively. The density of states is weighted by the Fermi velocity raised to the power of n , which is 0 for electrons ejected in a spin-polarized photoemission experiment (as in Figs. 11 and 12), 1 for ballistic transport (i.e. tunneling, as in a TMR measurement) and 2 for diffusive transport or tunneling at low bias[51].

In the MTJ, Butler et al. [52] and Mathon and Umersky[53] predicted that the tunneling conductance depends strongly on the band structure of the barrier as this would have a dramatic effect on how the states in the ferromagnetic contacts decay in the barrier. In the case of Fe/MgO/Fe they predicted a very strong spin filtering effect of the barrier. In 3d ferromagnetic metals the Bloch states with Δ_1 symmetry (spd hybridized states) tend to have large majority spin polarization at the Fermi level. Bloch states with Δ_5 (pd states) and Δ_2 symmetry (d states) have smaller or minority (opposite) spin polarization at the Fermi level. In MgO the attenuation is much larger for states with Δ_5 and Δ_2 than Δ_1 symmetries (see Fig. 14), resulting in preferential tunneling of majority spins. This spin filtering effect can substantially increase the measured TMR. The use of epitaxial MgO(001) tunnel barriers rather than AlO_x amorphous barriers resulted in a dramatic increase in TMR[54-56].

MgO(001) tunnel barriers have also resulted in a dramatic increase over AlO_x tunnel barriers for Heusler compound based MTJs. Figure 15 shows the improvement in TMR with change in tunnel barrier material, time and variations in Heusler composition (data obtained from references [57-67]). These are all structures made using sputter deposition and post deposition annealing rather than MBE. The TMR is substantially higher at low temperature with the record TMR ratios of 2610% at 4.2K and 429% at 290K for a $Co_2Mn_{1.24}Fe_{0.16}Si_{0.84}$ composition film[67]. From the figure, the films with excess Mn, Fe and deficient Si have the highest TMR. From the arguments above, the excess Mn should minimize the number of Co_{Mn} antisite defects that would reduce the spin polarization. An alternate explanation would be that the Fermi level is being tuned into the middle of the gap for minority spin density of states by VEC tuning[46]. Liu

et al. [67] have done one of the most extensive studies on the influence of composition on the TMR and conclude that both mechanisms are responsible for their record TMR results. This conclusion can be seen from Fig. 16, which shows the TMR ratio data from reference [67] plotted versus the Y-element concentration (Mn+Fe) (Fig. 16(a) and (b)) and VEC (Fig. 16(c) and (d)). It is clear that for $\text{Co}_2\text{Mn}_{0.73+\delta}\text{Si}_{0.84}$ the TMR is strongly dependent upon VEC. The decrease in the TMR ratio with increasing VEC for $\text{Co}_2\text{Mn}_{0.73+\delta}\text{Si}_{0.84}$ is consistent with the recent spin polarization measurements of $\text{Co}_2\text{Mn}_{1+\delta}\text{Si}$ of Andrieu et al.[46], which suggest that the spin polarization at the Fermi level decreases as the Fermi level increases in energy (Fig. 11), i.e. as VEC increases. However as discussed above, it is difficult to make a quantitative comparison between the spin polarization measured by photoemission, which measures the density of states, and the MTJ, which measures the spin dependent transport through a tunnel barrier. Furthermore, the details of the surfaces and interface are important.

Figure 17 shows a high angle annular dark field scanning transmission electron microscopy (HAADF-STEM) image of an MTJ with $\text{Co}_2\text{Mn}_{1.24}\text{Fe}_{0.16}\text{Si}_{0.84}$ ferromagnetic layers. This corresponds to the MTJ with the highest TMR ratio shown in Fig. 15. Note that although the interface appears abrupt, it is not atomically smooth. There are clear atomic variations in the MgO thickness. Furthermore, Figs. 17(b) and (c) show that the $\text{Co}_2\text{Mn}_{1.24}\text{Fe}_{0.16}\text{Si}_{0.84}$ is mixed phase with both the disordered B2 (region b) and ordered L_{21} (region c) crystal structures, respectively. The fact that these MTJ have shown the highest TMR ratios to date suggests that the B2 disorder does not have a strong influence on the spin polarization, which is in agreement with theory[40-42].

Beyond MTJs and Metal Spintronics: Integration with III-V Semiconductor Technology

Metallic MTJ and GMR devices have been highly successful in applications such as hard discs read heads and magnetic random access memory (MRAM) devices. Integration with semiconductors has the potential for developing devices with new function. The spin-field effect transistor (spin-FET)[68] (Fig. 18), spin-LEDs and spin-laser[69-71], as well as spin logic devices[72] are examples. In the spin-FET, ferromagnetic contacts are used for spin injection and detection into a two-dimensional electron gas (2DEG) in a semiconductor with large spin-orbit coupling. The spin precession due to spin-orbit coupling is controlled by the applied voltage on the Schottky gate, which produces a perpendicular electric field that results in an effective magnetic field (Rashba) that makes the spins precess in the channel [68]. Spin-induced gain anisotropy on output polarization and reduced threshold current have been demonstrated for quantum dot vertical cavity surface emitting lasers (VCSELs) using ferromagnetic spin injecting contacts[69,70]. The high spin polarization of Heusler compounds would make these ideal potential candidates for spin contacts to semiconductors. However, spin polarization is not the only significant factor; lattice matching, thermodynamic stability, interfacial bonding and electronic properties are all important[73]. MgO is remarkably stable against reactions with transition metals, whereas III-V semiconductors are much more reactive. Hence the approach of high temperature annealing to achieve ordering of the Heusler films for MTJs is unlikely to be possible for growth on III-V semiconductors.

Lattice Matching

As shown in Fig. 2, the Heusler compounds have fcc crystal structures ($L2_1$, $C1_b$, and DO_3) that are closely related to the III-V semiconductor zincblende structure. Fig. 19 shows a plot of lattice parameter versus element for a number of Heusler compounds. The plot indicates the lattice parameters of a number of common III-V semiconductors. As can be seen from this figure, Co_2MnSi and Co_2FeSi are closely lattice matched to GaAs, suggesting that these may be ideal spin contacts to GaAs. As discussed above, these are expected to have high spin polarization.

Thermodynamic stability

The Heusler compound/III-V semiconductor heterostructure is unlikely to be thermodynamically stable as this would typically involve 5 elements with the potential for the formation of a very large number of different compounds through multiple combinations of the elements, such as between Co, Mn, Si, Ga and As for a Co_2MnSi /GaAs heterostructure. Although a number of metal-III-V ternary phase diagram predictions and measurements have been made[74,75], quaternary and pentenary phase diagrams have not been made. There have been numerous studies of metal/III-V semiconductor contact reactions[76-79]. In general, most elements will react with GaAs and other III-V semiconductors. Si and Ge are two of the elements that are predicted to be thermodynamically stable in contact with GaAs[74]. Although the individual elements are not thermodynamically stable in contact with III-V semiconductors, metallic compounds of the elements can be [78,79]. Although the relative stability of each Heusler compound and III-V semiconductor is not known, one can make some intuitive predictions on which systems may be more stable than others using knowledge of the reaction kinetics of the individual elements with the III-V semiconductor. In the case of transition elements and GaAs, the interfacial reaction kinetics are fastest for Ni[80], then Mn[81], Co[82-84], and the slowest is Fe[85,86]. Hence, one may expect Ni based Heusler compounds to be more likely to have higher interfacial reaction kinetics than Fe based ones.

From Fe/GaAs thin film reaction studies Schultz et al. predicted the Fe-Ga-As phase diagram shown in Fig. 20[86] which indicates that Fe is not thermodynamically stable in contact with GaAs, but Fe-Ga and Fe-As phases are and, in particular, Fe_3Ga . In this study they found that the Fe_3Ga grew in the metastable DO_3 structure, possibly due to epitaxial stabilization by the GaAs substrate. As discussed above, this DO_3 structure is essentially the Heusler $L2_1$ crystal structure with elements X and Y being the same (i.e. Fe_2FeGa) (Fig. 2).

Heusler Compound Growth on III-V Semiconductors

The prediction of half-metallicity for the half-Heusler compounds $NiMnSb$ and $PtMnSb$ [11] and for the full-Heusler compounds Co_2MnGe and Co_2MnSi [30] precipitated strong interest in pursuing experimental verification of half-metallicity by epitaxial Heusler compound growth. Unlike metallic MTJs, where sputtering has been the technique of choice, most Heusler compound growth on semiconductors has generally been by molecular beam epitaxy (MBE) using individual elemental sources. Initial Heusler growth studies on III-V semiconductors were aimed at studies of the film properties, particularly the magnetic and electronic properties, rather than the Heusler compound/III-V heterostructure properties. A number of Heusler compounds were successfully grown on III-V semiconductors. These include: $Fe_3(Al,Si)$ [87-89], Ni_2MnGa , Ni_2MnAl , Ni_2MnIn and Ni_2MnGe [90-96], Co_2MnGe [97-100], Co_2MnSi [101,102], Co_2FeSi [103-

106], Co₂MnGa[107], Co₂NiGa[108] and the half-Heusler NiMnSb[109-114]. These studies clearly demonstrated high crystal quality epitaxial growth. Point Contact Andreev Reflection (PCAR) measurements of the spin polarization of Co₂FeSi films suggest a polarization $\leq 60\%$ [106,115], similar to other Heusler compounds[116,117] including NiMnSb and Co₂MnSi[118,119]. PCAR measurements were performed on air exposed films. In-situ spin-polarized photoemission measurements of Co₂MnSi epitaxial films[46,48], discussed above, yielded much higher spin polarization ($>90\%$). This suggests that either the in-situ prepared films were of higher crystal quality or air exposure significantly alters the film properties. An alternate explanation for the large difference in transport spin polarization (PCAR) and the density of states spin polarizations is a difference in the Fermi velocity for the two spins, as discussed above (equation (5)). Discussions above also indicated that Y-Z atomic disorder should, in general, not have a large influence on the spin polarization. However, this is not always the case. For instance, Ni₂MnAl is ferromagnetic in the ordered L₂₁ structure and is antiferromagnetic in the B2 [120,121]. Dong et al.[10] found that by controlling the substrate temperature during MBE growth, antiferromagnetic or ferromagnetic Ni₂MnAl films could be grown. Antiferromagnetic epitaxial Ni₂MnAl in the B2 structure would grow at 180°C while growing at 400°C resulted in L₂₁ ordering and being ferromagnetic.

Growth of Heusler Compound Contacts for Semiconductor Spintronics

In the preceding sections the focus has been on the properties of the Heusler compounds; however, when integrating Heusler compounds with III-V semiconductors, the interfacial properties are critical. The large impedance mismatch between a metallic Heusler compound and a semiconductor requires the transport to be ballistic (i.e. tunneling) rather than diffusive in order to achieve significant spin polarization[122]. Fortunately, Fermi level pinning results in metal contacts to GaAs being Schottky barriers rather than ohmic contacts. Hence, in order to ensure sufficient tunneling through the Schottky barrier, the semiconductor immediately beneath the contact must be heavily doped so as to have a shallow depletion width and narrow barrier[123,124]. Furthermore, the details of the narrow degenerate doping region in the vicinity of the ferromagnet/semiconductor interface plays an essential role in the spin injection and detection[125]. In addition, interfacial reactions can have a dramatic effect on the spin transport[126-128]. These results indicate that the properties of Heusler/III-V semiconductors are likely to be very sensitive to both the details of the III-V semiconductor interfacial doping profile, the semiconductor surface preparation and the heteroepitaxial growth process itself.

The surface composition and reconstruction of III-V semiconductors depends on temperature and group-V overpressure. Figure 21 shows the surface phase diagram for GaAs(001)[129]. During MBE growth, the typical surface reconstruction is the (2x4)/c(2x8) β 2 with ~ 0.75 monolayer As coverage[130]. Upon cooling under an As-flux, this surface adsorbs more As and converts to a c(4x4) reconstruction with ~ 1.75 monolayer As coverage. Other reconstructions and terminations can be achieved through annealing and controlling of the Ga and As fluxes[130]. It is important to remember that although GaAs is a cubic crystal, the (001) surface has two fold surface symmetry and therefore the [110] and [1-10] directions are not equivalent due to the sp³ bonding. This tends to result in interfacial induced uniaxial magnetic anisotropy of epitaxial magnetic films[131].

In the experimental results described below, a multi chamber ultra-high vacuum (UHV) growth and in-situ characterization system was used to grow, characterize and study the growth process of Heusler compound/GaAs heterostructures.

Measurement of spin transport at Heusler/III-V semiconductor interfaces

Spin-LED

Initial demonstration of spin injection from a ferromagnetic contact into III-V semiconductor heterostructures relied on the use of a spin-LED[123,124,132-134]. Figure 22 shows a schematic of a spin-LED structure. The ferromagnetic-/semiconductor Schottky barrier is reverse biased to inject electrons into the n-type semiconductor and the p-n junction is forward biased. The injected electrons recombine with holes from the p-type layer in the p-n junction depletion region, which in this case is an undoped region with a GaAs quantum well (QW). A magnetic field is applied perpendicular to the contact to ensure that the ferromagnetic contact magnetization is pulled out of plane so the injected electron spin orientation is perpendicular to the surface. Measurement of the left and right circularly polarized light gives, through the optical selection rules, the spin polarization of the electrons recombining with holes in the quantum well. The spin polarization should track with the out of plane magnetization of the ferromagnetic contact. Although this is a direct measure of the spin polarization in the semiconductor quantum well, it has several drawbacks. It can only be used for semiconductors with a direct band gap and it is essentially a back-to-back diode, which makes it difficult to determine the potential profile within the device. The first demonstration of spin injection from a Heusler compound into GaAs was reported by Dong et al. for Co_2MnGe [135] and later for Co_2FeSi by Ramsteiner et al.[105] using spin-LEDs.

Non-local lateral Spin Valve measurements

Measurement of spin transport in lateral ferromagnet/semiconductor devices offers many advantages over spin-LEDs as spin injection, transport and detection can all be measured in a single device[136,137]. Figure 23 shows a schematic of the contacts and the non-local measurement. The doping profile in the GaAs is similar to that used in the spin-LED. However, the n^+ contact layer is etched away between the ferromagnetic contacts to ensure that the lateral transport is in the n-type GaAs channel. Spin-polarized electrons are injected into the GaAs channel at contact C and flow towards contact A, while the voltage, $V_{D,E}$, is measured between contacts D and E. Although current flows between contacts C and A, non-equilibrium spin polarization in the GaAs can diffuse in either direction from the source. The spin polarization results in an electrochemical potential difference, $\Delta\mu$, for the two spin states in the channel, leading to a change in $V_{D,E}$ when the magnetization of contact D is switched antiparallel to contact C, to the parallel configuration. The GaAs spin polarization at contact E is always zero as it is many spin diffusion lengths away. The spin-valve measurement is carried out by sweeping the magnetic field along the magnetic easy axis of the contacts and looking for a change in voltage over the narrow field range in which contacts C and D are antiparallel. The top curve in Fig. 23(d) is the raw signal and the bottom is the signal after the linear background has been subtracted. By setting the contacts C and D in parallel and antiparallel states and applying a small varying out of plane magnetic field, Hanle measurements can be made. Fig 23(e) shows

the raw signal (top curve) and after background subtraction (bottom curve). By varying the current and bias of contact C, the GaAs spin polarization can be measured as a function of forward (spin extraction) and reverse (spin injection) bias. Magnetoptic Kerr effect (MOKE) can also be used to measure the spin polarization directly in the Ga As channel[137-139]. Furthermore, by biasing of the detector the spin detection sensitivity can be increased significantly[137,139].

Fe₃Ga/GaAs(001)

Thin film Fe/GaAs reaction studies by Schultz et al.[86] found that Fe₃Ga with the DO₃ related Heusler structure formed after 450°C annealing and that it should be thermodynamically stable in contact with GaAs as is shown in the phase diagram in Fig. 19. The mismatch between Fe₃Ga in the DO₃ structure (5.812 Å) and GaAs (5.6533 Å) is ~3%. Figure 24 shows x-ray diffraction scans for a 30nm thick Fe₃Ga film grown by codeposition at 325°C on GaAs(001). The film grows in the DO₃ structure and the reciprocal lattice scan around the GaAs(224) peak indicates that the film is growing pseudomorphic on the GaAs(001) surface with an out of plane lattice parameter of 6.12 Å. The finite thickness fringes indicate a relatively smooth film and sharp interface.

When considering the growth of Fe₃Ga on GaAs(001) one has to consider the atomic details of the growth process as shown in Fig. 25. The first question to be answered is to determine which GaAs surface reconstruction should be used. As-rich or Ga-rich reconstructions can promote specific interfacial phase formations. For example, an As-rich surface may promote the formation of Fe-As compounds, which can be antiferromagnetic. A second consideration is the stacking sequence in Fe₃Ga itself, which in the [001] direction involves sequential layers of Fe+Ga and Fe. The Fe₃Ga growth can then be initiated with a layer of Fe or Fe+Ga followed by sequential atomic layers, by codeposition or a mixture of sequential layers and codeposition. To test the influence of the growth sequence, GaAs(001) heterostructures for lateral transport measurements as shown in Fig. 23 were grown and then different surface reconstructions were prepared immediately prior to Fe₃Ga growth. The Fe₃Ga initial growth sequence was also varied. Fig. 26 shows the procedures used to create different interfaces.

Figure 27 shows high angle annular dark field scanning transmission electron microscopy (HAADF-STEM) images of the Fe₃Ga/GaAs(001)As-rich and Ga-rich interface samples[140]. The interfaces appear atomically abrupt suggesting minimal interfacial reactions. However, the interfaces are clearly different between the [1-10] and [110] cross-sections as would be expected for the two fold surface symmetry. In addition, there is a clear difference in interface structure between the two samples. The [110] cross section of the Ga-rich interface sample shows additional atomic structures every three Ga-As dumbbells at the interface. This is interpreted as remanence of the Ga-rich GaAs(001) c(8x2) reconstruction[130].

Lateral transport spin valves were fabricated and measured for the differently prepared interfaces and the bias dependence of the spin polarization electrochemical potential difference (spin valve signal) is shown in Fig. 28. Clearly, the Ga initiated growth on GaAs(001) (4x6)/c(8x2) Ga-rich surface interface sample (sample A) has the highest spin valve signal.

Co₂Mn_{1-x}Fe_xSi/GaAs(001)

As discussed above, Co_2MnSi is predicted and recently confirmed by in-situ spin polarized photoemission measurements to be >90% spin polarized [46,48]. This together with the close lattice match to GaAs (0.06% mismatch) and high Curie temperature (985K), makes Co_2MnSi very attractive as a potential spin contact to GaAs.

Using effusion cells for Co and Mn and a single crystal Si strip heater cell for Si, Co_2MnSi films were grown on GaAs(001) c(4x4) As-rich surfaces[141]. Streaky RHEED patterns were observed during growth as were RHEED intensity oscillations indicative of a layer-by-layer growth mode at a substrate temperature of 270°C (Fig. 29). The effusion cell fluxes were set using a beam flux monitor that had been calibrated using Rutherford backscattering measurements of calibration samples grown on Si substrates. In-situ scanning tunneling microscopy (STM) images showed step heights corresponding to 1 bilayer (2.8 Å) of Co_2MnSi (Fig. 29(c)), consistent with the layer-by-layer growth inferred from the RHEED intensity oscillations Fig. 29(b)). The 2-fold surface symmetry of GaAs (001), combined with the thin film geometry, induces a strong in plane uniaxial magnetic anisotropy on the Co_2MnSi with an easy axis in the [110] direction (Fig. 29(d)).

The spin polarized photoemission measurements only measure the surface of the films and not the buried interface. For spin transport at the $\text{Co}_2\text{MnSi}/\text{GaAs}$ interface, the spin polarization at the interface is critical. There are many theoretical studies predicting the interfacial spin polarization for a number of different Heusler/III-V semiconductor heterostructures. These include $\text{Co}_2\text{MnGe}/\text{GaAs}$ [41,142,143], $\text{Co}_2\text{MnSi}/\text{GaAs}$ [143-146], $\text{Co}_2\text{FeSi}/\text{GaAs}$ [147], $\text{Co}_2\text{CrSi}/\text{GaAs}$ [143], $\text{Co}_2\text{CrAl}/\text{InP}$ [148], $\text{NiMnSb}/\text{GaAs}$ [149] and NiMnSb/InP [149]. In general, the predicted interfacial spin polarization is found to depend strongly on the interfacial atomic configuration and crystal orientation.

As in the case for $\text{Fe}_3\text{Ga}/\text{GaAs}(001)$ discussed above, different growth sequences may result in different interfacial structures. Looking at the Co_2MnSi L_{21} Heusler structure (Fig. 2) it is clear that the [001] stacking sequence consists of alternate layers of Mn+Si and Co. By performing sequential atomic layer depositions and in-situ analysis using RHEED, STM and x-ray photoelectron spectroscopy (XPS) after each layer, a picture of the different interface formations can be inferred. Starting with GaAs(001) c(4x4) As-rich surfaces, two different growth sequences were investigated. The first involved initiating the growth with 1 monolayer (ML) of Mn-Si followed by 1ML Co and then 1 ML Mn-Si in an atomic layer by atomic layer growth mode. The second used the same growth mode, but initiated the growth with 1ML Co. RHEED of a starting c(4x4) GaAs (001) surface as well as after the deposition of each monolayer in Co- and MnSi-initiated samples of Co_2MnSi is shown in Fig. 30. After the deposition of the first monolayer, there are no significant differences in RHEED, and after deposition of 15.4nm of film, the RHEED patterns are nearly identical, showing a slight 2x periodicity in both the [110] and $[\bar{1}\bar{1}0]$ directions. The corresponding XPS measurements (Fig. 31) of the Ga 3d peak intensity attenuates slower than expected indicating that Ga is either riding on the film surface or the Co_2MnSi is growing as islands and not completely covering the GaAs surface. From the in-situ STM images of the Co_2MnSi it is evident that the Co_2MnSi film after the first few ML is not growing as islands and the Ga must therefore be riding on the surface. The intensities of the Ga 3d, Co 2p and Mn 2p peaks as a function of monolayer deposition are shown in Fig. 31 for the two different growth sequences. Intensities are expected to increase upon deposition of the layer that contains the element, and are expected to attenuate after deposition of the next layer. The trend in the MnSi initiated sample is consistent with this description, but for Co initiations, the MnSi layer does not attenuate after the first layer,

suggesting that the Mn is going beneath the Co. Regardless of the nucleation sequence, the $\text{Co}_2\text{MnSi}/\text{GaAs}(001)$ interface is similar and Mn-rich (Fig. 32).

Electrical measurements are generally more sensitive than XPS. The room temperature current-voltage characteristics of $\text{Co}_2\text{MnSi}/\text{GaAs}(001)$ contacts with Co, Mn, Si, and Mn-Si nucleation layers were compared and are shown in Fig. 33(a). The interpretation of these data is that the Co and Si initiated contacts have similar effective Schottky barrier heights and that it is higher for Mn-Si and highest for Mn. This increase in the effective Schottky barrier height could either be a result of a true Schottky barrier height increase or due to a decrease in the n^+ type doping in the GaAs. Since Mn is a p-type dopant in GaAs and any Mn diffusion into the GaAs would result in compensation of the Si n-type doping. This would increase the Schottky barrier width, decreasing the tunneling probability, and therefore increase the effective Schottky barrier height. Fig. 33(b) shows a backside secondary ion mass spectrometry (SIMS) depth profile of the different elements for a structure with MBE-grown AlGaAs marker layers for sputter depth profile calibration[150]. The signal for Mn rises significantly before the Co and Si signals, indicating Mn diffusion into the GaAs. In addition, the profile is also consistent with a Mn rich interface, in agreement with the in-situ XPS studies. Ebina et al.[102] used conventional front side SIMS, which is more prone to ion beam mixing and knock-on resulting in decreased depth resolution, to detect Mn indiffusion from sputter deposited $\text{Co}_2\text{Mn}_{1.30}\text{Si}_{0.84}$ films and concluded that this was the cause of no clear spin signal. They found that the insertion of a thin FeCo layer suppressed the Mn indiffusion and observed high spin injection efficiency. Mn diffusion is also consistent with the difference in electrical properties of spin-LEDs with δ -MnGa[151] and Fe contacts.

Figure 34 shows [1-10] and [110] cross-section HAADF-STEM images of a $\text{Co}_2\text{MnSi}/\text{GaAs}(001)$ heterostructure. The images suggest that the interfaces look atomically abrupt, but detailed elemental atomic distribution at the interface may be complex, requiring a detailed comparison between theory and experiment. There is significant intensity modulation in the atomic rows on either side of the apparent interface. The Co_2MnSi also shows regions of both $L2_1$ and B2 ordering.

Non local spin valve measurements were performed on $\text{Co}_2\text{MnSi}/\text{GaAs}(001)$ heterostructures using similar procedures to those used for $\text{Fe}_3\text{Ga}/\text{GaAs}$ and $\text{Fe}/\text{GaAs}(001)$ [136]. The spin valve signals for a $\text{Co}_2\text{MnSi}/\text{GaAs}(001)$ and $\text{Fe}/\text{GaAs}(001)$ heterostructure are shown in Fig. 35(a) for both an unbiased and biased detector configuration. The temperature dependence of the spin signal for a biased detector (Fig. 35(b)) shows spin signal in excess of 2mV at low temperature. The non-local voltage is much larger for the Co_2MnSi sample. By modeling the spin injection efficiency and enhancement due to detector bias, the spin accumulation in the GaAs channel was calculated to be 60% for Co_2MnSi as compared to 30% for best Fe contacts[152].

The importance of the VEC to the properties of Heusler compounds was discussed in detail above. Fe has one more electron than Mn. Hence if Fe were substituted for Mn, the simple picture is that the Fermi level in the $\text{Co}_2\text{Mn}_{1-x}\text{Fe}_x\text{Si}$ would increase with x. This is schematically depicted in Fig. 36 (top). The spin polarization at the Fermi level should decrease and switch sign when increasing x from 0 to 1. This is what is experimentally observed (Fig. 36 (bottom)). The spin polarization has the opposite sign for Co_2FeSi contacts as compared to Co_2MnSi .

The large spin signals obtained electrically from lateral GaAs spin devices with Co_2MnSi and Co_2FeSi ferromagnetic contacts have enabled detection of electron spin accumulation, electrically in GaAs up to room temperature [102,153]. Figure 37 shows the non local spin valve

signal for $\text{Co}_2\text{MnSi}/\text{GaAs}(001)$ as a function of contact separation for temperatures ranging from 25 to 300K.

Co_2MnSi and Co_2FeSi contacts have also been used to demonstrate tunneling anisotropic magnetoresistance (TAMR) detection of ferromagnetic resonance at room temperature even under conditions where no spin accumulation is present and no spin-polarized current flows in the semiconductor[154].

Outlook

Heusler based MTJs have been very successful and are of great technological interest for developing spin transfer torque magnetic random access memory (STT-RAM) devices [155,156]. This requires ferromagnetic material with spin torque efficiency and high TMR as well as low magnetic damping [156]. Perpendicular magnetic anisotropy reduces the switching current while maintaining thermal stability. Half-metallic Heusler compounds would have high TMR and high spin torque efficiency. The lack of a minority spin density of states at the Fermi level would also result in low magnetic damping. NiMnSb and Co_2MnSi have very low magnetic damping coefficients ($\alpha = 0.002$ for NiMnSb [157] and 0.0007 for $\text{Co}_{1.9}\text{Mn}_{1.1}\text{Si}$ [46], respectively). Interface induced perpendicular magnetic anisotropy using $\text{MgO}(001)$ has been observed for Co_2FeAl [158,159] and Co_2FeSi [160] thin films. The challenge is to develop half-metal Heuslers with bulk magnetic anisotropy rather than interface induced[156,161]. Azadani et al. have predicted a number of different Heusler superlattices depending on crystallographic orientation that should be both half-metallic and have perpendicular magnetic anisotropy [162]. These predictions have yet to be realized experimentally.

The versatility of Heusler compounds is shown by the vast number of properties they can have, the demonstration of MBE growth of high quality semiconducting [163,164], topological Heusler compounds [165-167], and the ability to tune their properties by controlling the VEC. The possibility to form heterostructures of Heusler compounds with vastly different electronic properties but with the same basic crystal structure and their demonstrated ability to integrate with conventional semiconductors and metal oxides promises an exciting future for their use in novel devices with new function.

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Figures

- Fig. 1 Periodic table with color coding showing the typical elements for X (blue), Y (red) and Z (green) in the half-Heusler XYZ and full-Heusler X_2YZ compounds.
- Fig. 2 General ordered cubic crystal structures for the full-Heuslers X_2YZ , the $D0_3$, when X and Y are the same element (X_2XZ), the B2 when Y and Z are completely disordered, the half-Heuslers XYZ, the zincblende structure common to III-V semiconductors, and the NaCl. Note the strong similarities between these structures.
- Fig. 3 Electronic properties of full (FH) and half (HH) Heusler compounds as a function of the number of valence electrons per formula unit.
- Fig. 4 Calculated total densities-of-states of half-Heusler compounds with various numbers of valence electrons per formula unit (from ref. [24]). Note how the Fermi level moves across the gap in the density of states with increasing valence electron count.
- Fig. 5 Plot of spin magnetic moment versus valence electron count for a number half-Heusler compounds (from reference [36]).
- Fig. 6 Plot of total magnetic moment versus valence electron count for a number full-Heusler compounds (from reference [5]).
- Fig. 7 Slater-Pauling behavior for a number of inverse Heusler compounds. Note the rules of “18, 24 and 28” (from reference [27]).
- Fig. 8 Slater-Pauling curve for a number of conventional magnetic alloys and some full-Heusler compounds (from ref [1]).
- Fig. 9 Curie temperature versus VEC for a number of full Heusler compounds (from reference [1]).
- Fig. 10 Calculated spin dependent band structure of for the half metal Co_2MnSi : a) density of states and b) majority and minority spin band structure (from reference [38]).
- Fig. 11 Spin-resolved photoelectron spectroscopy spectra at 300 K for a series of $Co_2Mn_{1+\delta}Si$ layers with increasing Mn content. As shown in the schemes, Fermi level is moving towards the spin polarization maximum as the Mn content increases. (from reference [46]).
- Fig. 12 Influence of Mn and MgO coverage on Co_2MnSi electronic properties. In (a) and (b) are shown the electron diffraction patterns for the starting Co_2MnSi layers and when covered by Mn or MgO. The Spin-resolved photoelectron spectroscopy spectra and spin-polarization spectra for the starting Co_2MnSi layers and after the capping are shown in (c) Mn cap and (d) MgO cap. As the spin polarization is close to zero at Fermi level for both

Co₂MnSi uncoated surfaces, it is strongly increased when covering with Mn or MgO. (from reference [46]).

Fig. 13 Schematic of MTJ and (b) spin valve (GMR) structures and density of states for parallel and antiparallel alignment of the ferromagnetic layers. In the antiparallel state there are no states for the majority spins (spin up) to go into in the antiparallel aligned electrode without spin flip scattering.

Fig. 14 Tunneling density of states for $k_{\parallel} = 0$ for Fe(100)/MgO/Fe(100) for majority and minority spins with aligned (upper panels) and antialigned (lower panels) magnetic moments (from reference [52]).

Fig. 15 TMR versus date for full Heusler compound magnetic tunnel junctions. The data were obtained from references [57-67].

Fig. 16 TMR ratio for Co₂(Mn,Fe)_ySi_{0.84} dependence on the Y site concentration (Mn+Fe) (a) and (b) and VEC (c) and (d). (Data obtained from reference [67]).

Fig. 17 HAADF-STEM images of a Co₂Mn_{1.24}Fe_{0.16}Si_{0.84}/MgO/Co₂Mn_{1.24}Fe_{0.16}Si_{0.84} MTJ. Regions (b) and (c) correspond to areas with B2 and L2₁ crystal structures, respectively. Note that the lower MgO/Co₂Mn_{1.24}Fe_{0.16}Si_{0.84} interface is substantially smoother than the upper interface. (from reference [67]).

Fig. 18 Spin-FET. The Schottky gate is used to control the spin precession as electrons in a 2DEG are transported from the spin injector to spin detector. Semiconductors with large spin-orbit coupling will have more spin precession

Fig. 19 Plot of lattice parameter versus element for a number of Heusler compounds. Note how the lattice parameters span those of a number of III-V compound semiconductors. The effective lattice parameter for MgO assuming a 45 degree rotation ($\sqrt{2}a_{\text{MgO}}$) is also indicated.

Fig. 20 Predicted Fe-Ga-As ternary phase diagram. (adapted from reference [86]).

Fig. 21 GaAs(001) surface reconstruction phase diagram (from reference [129]).

Fig. 22 Schematic a spin-LED (a) band diagram, (b) typical heterostructure and (c) electroluminescence polarization as a function of out of plane magnetic field for a spin-LED with ferromagnetic Fe₃Ga spin injection contact.

Fig. 23 Schematic diagrams of the experiment and representative non-local spin-valve and Hanle effects. (a) typical layer structure beneath the ferromagnetic contacts. Note that the n⁺-GaAs layers are etched away in the transport channel regions. (b) band diagram of the Schottky contact with the n⁺-doped GaAs contact layers, (c) non-local spin-valve device structure, (d) and (e) are the non-local spin valve and Hanle effect measurements. The

upper panels show the raw data and the lower panels after background subtraction. The inserts show the measurement configurations.

Fig. 24 X-ray diffraction scan, rocking curve and reciprocal space map of a Al capped $\text{Fe}_3\text{Ga}(30\text{nm})/\text{GaAs}(001)$ heterostructure grown by codeposition at 325°C .

Fig. 25 Considerations on the growth procedure for Fe_3Ga on $\text{GaAs}(001)$.

Fig. 26 Procedures used to create different $\text{Fe}_3\text{Ga}/\text{GaAs}(001)$ interfaces.

Fig. 27 HAADF-STEM images of $\text{Fe}_3\text{Ga}/\text{GaAs}(001)$ heterostructure interfaces: top: Sample E with Fe/As interface grown at 175°C on $\text{GaAs}(2\times 4)/c(2\times 8)$ surface and bottom Sample A - FeGa/Ga interface grown at 300°C on $\text{GaAs}(4\times 6)/c(8\times 2)$ surface) (from [140])

Fig. 28 Bias dependent lateral spin valve signals for $\text{Fe}_3\text{Ga}/\text{GaAs}(001)$ heterostructures grown with different interface growth sequences. The Fe-Ga initiated growth on a Ga-rich $\text{GaAs}(001)$ surface shows the highest spin valve signal (from [140])

Fig. 29 5nm thick Co_2MnSi grown on $\text{GaAs}(001)$ (a) RHEED diffraction patterns, (b) RHEED intensity oscillations during growth, (c) in-situ STM image of the surface and (d) ex-situ magnetic hysteresis loops. (adapted from reference [141])

Fig. 30 RHEED of a starting $c(4\times 4)$ $\text{GaAs}(001)$ surface as well as after the deposition of each monolayer in Co- and MnSi-initiated samples of Co_2MnSi .

Fig. 31 Measured intensities of the Ga 3d, Co 2p and Mn 2p core levels as a function of sequential monolayer growth thickness. Top panels: MnSi initiated and bottom panel Co initiated films. Note that the Co 2p peak intensity does not decrease after the first MnSi monolayer deposition for the Co initiated film.

Fig. 32 Schematic of the results from the sequential monolayer growth studies. The $\text{Co}_2\text{MnSi}/\text{GaAs}$ is similar regardless of the nucleation sequence.

Fig. 33 $\text{Co}_2\text{MnSi}/\text{GaAs}(001)$ Schottky contacts (a) Current-voltage characteristics at room temperature for Si, Mn, Co and MnSi initiated growth. (b) Backside SIMS depth profile to detect preferential Mn indiffusion beneath the contact.

Fig. 34 Cross-sectional HAADF-STEM images of $\text{Co}_2\text{MnSi}/\text{GaAs}(001)$ interface. (Courtesy of P. Voyles).

Fig. 35 Lateral transport non local spin valve signals for $\text{Co}_2\text{MnSi}/\text{GaAs}(001)$ and $\text{Fe}/\text{GaAs}(001)$ for (a) unbiased and biased detector. (b) temperature dependence of the spin signal for a $\text{Co}_2\text{MnSi}/\text{GaAs}(001)$ with a biased detector (Courtesy P. Crowell).

Fig. 36 Top: Schematic of the density of states for $\text{Co}_2\text{Mn}_{1-x}\text{Fe}_x\text{Si}$ as a function of x . Bottom: Temperature dependence of spin polarization determined by biased detector technique (Courtesy P. Crowell).

Fig. 37 Non-local voltage versus Co_2MnSi contact separation for 25-300K (Courtesy P. Crowell).

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Fig.24

- **GaAs surface reconstruction**
- As-rich
- Ga-rich
- how As- or Ga- rich, multiple reconstructions are possible
- **Initiation of Fe_3Ga growth**
- Fe first
- Ga first
- Co-deposition of Fe+Ga

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