High-mobility AlAs quantum wells with out-of-plane valley occupation

K. Vakili, Y.P. Shkolnikov, E. Tutuc, E.P. De Poortere, M. Padmanabhan, and M. Shayegan
Department of Electrical Engineering, Princeton University, Princeton, NJ 08544
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Employing state-of-the-art molecular beam epitaxy techniques to grow thin, modulation-doped AlAs quantum wells, we have achieved a low temperature mobility of 5.5 m²/V·s with out-of-plane occupation, an order of magnitude improvement over previous studies. However, due to the narrow well width, mobilities are still limited by scattering due to interface roughness disorder. We demonstrate the successful implementation of a novel technique utilizing thermally-induced, biaxial, tensile strain that forces electrons to occupy the out-of-plane valley in thicker quantum wells, reducing interface roughness scattering and allowing us to achieve mobilities as high as 8.8 m²/V·s.

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Two-dimensional electron systems (2DESs) confined to AlAs quantum wells (QWs) have revealed a variety of novel phenomena in recent years. This has been made possible, in large part, by advances in semiconductor growth techniques that have dramatically improved the quality of these systems [1, 2], allowing the observation of behaviors that had previously been obscured by disorder effects. These improvements have, however, only been implemented in thick AlAs QWs in which electrons occupy the valleys oriented in the QW plane. Recent work has demonstrated the value of studying thinner AlAs QWs with the out-of-plane valley occupied [3, 4], and there is therefore a desire to reduce the obfuscating effects of disorder in these systems as well.

Bulk AlAs has an indirect band gap with conduction band minima at the six equivalent X-points of the Brillouin zone. The Fermi surface for electrons consists of three full (six half) prolate ellipsoids, or valleys, with major axes parallel to each of the crystal axes (Fig. 1). The conduction mass, in units of the bare electron mass, is \( m_c = 1.04 \) along each ellipsoid’s major (longitudinal) axis and \( m_t = 0.21 \) along the minor (transverse) axes [4]. Confining the charges along a crystal direction breaks the bulk valley degeneracy as a consequence of the valleys’ mass anisotropy. Our QWs are grown on (001) GaAs substrates, and the valley oriented along this axis (which we will heretofore refer to as the Z valley) is lower in confinement energy with respect to the two in-plane valleys (X,Y) due to its larger mass along [001]. There is a competing effect, however, due to the lattice mismatch between AlAs and the GaAs substrate that induces a biaxial, compressive strain, lowering the X and Y valleys in energy with respect to the Z valley by approximately 23 meV [2, 3, 4]. Consequently, there is a crossover between Z and (X,Y) valley occupation at a particular QW thickness, \( t_c \sim 55 \) Å, with thinner wells favoring the Z valley and thicker wells favoring the X and Y.

The main difference between electrons occupying the Z or (X,Y) valleys is the shape of the Fermi contour projected in the 2DES plane. For the Z valley, the contour is isotropic, while the X and Y each have anisotropic contours. Though such anisotropy can be interesting, the Z valley’s isotropic profile provides a simpler system for the study of general properties of 2DESs [3, 4]. Furthermore, electrons in the Z valley are very similar to electrons in Si metal-oxide-semiconductor field-effect transistors (MOS-FETs) (e.g. similar band effective mass, Landé g-factor, etc.) except for the lack of valley degeneracy, thereby providing a control system for the study of effects related to that degeneracy. Though there are previous reports of occupation of the Z valley in AlAs [5, 6, 7], these early samples had low mobilities (\( \sim 0.37 \) m²/V·s at 0.5 K), and subsequent experiments in AlAs have almost exclusively involved X and Y valley occupation.

We have studied several different AlAs QWs with thicknesses less than or close to \( t_c \). All samples were patterned in an L-shaped Hall bar configuration with arms parallel to the in-plane crystal axes. Ohmic contacts were made by depositing AuGeNi and alloying in a reducing environment. Metallic front and back gates allow for in situ control of the electron density, \( n \), and facilitate populating the QW via a recently developed field-effect persistent photoconductivity (PPC) technique [8]. We made measurements in a pumped \(^3\)He refrigerator with a base temperature of 0.3 K and employed standard low-frequency lock-in techniques.

The layer structure near the surface of the wafer containing our thinnest QW is similar to that in Ref. [2] except for a narrower QW width. A 45 Å AlAs QW is grown atop a thick layer of Al\(_0.39\)Ga\(_{0.61}\)As and separated...
from a Si $\delta$-doping layer by a 750 Å spacer also made of Al$_{0.35}$Ga$_{0.65}$As. This front-side modulation-doping, together with the aforementioned PPC technique, considerably improves the mobility in thick AlAs QWs with the (X,Y) valleys occupied \cite{2}. The QW layer is flanked by thin layers of pure GaAs, separating it from the AlGaAs barriers to reduce interface roughness at the QW edges and limit the penetration of the electronic wavefunction into the AlGaAs where alloy scattering can take place.

A typical magnetoresistivity ($\rho_{xx}$) trace for a sample (A) from this wafer is shown in Fig. 2. Shubnikov-de Haas (SdH) oscillations are well developed and persist to fields as low as $B = 0.6$ T and filling factors as high as $\nu = 25$. This is a considerable improvement over the only other transport measurement for Z valley electrons in AlAs of which we are aware \cite{7}, where SdH oscillations first become perceptible at $B = 5$ T. We have confirmed that the Z valley is occupied in our samples by extracting six times lower than the peak $\mu$ values reported for the Z valley sample in Ref. \cite{7} and comparable to the highest mobility achieved in Si-MOSFETs.

The dependence of mobility, $\mu$, on $n$ for sample A is exhibited in Fig. 3 (triangles). A peak mobility of 5.1 m$^2$/Vs is achieved near $n = 3.5 \times 10^{11}$ cm$^{-2}$, which is more than an order of magnitude higher than the peak $\mu$ reported for the Z valley sample in Ref. \cite{7} and comparable to the highest mobility achieved in Si-MOSFETs. Though this is a dramatic improvement, it is still about six times lower than the peak $\mu$ values reported for thicker AlAs QWs with (X,Y) valley occupation \cite{2}, despite essentially identical growth conditions. Additionally, there is a saturation and even slight decline of $\mu$ as $n$ increases past the peak value. Both facts are consistent with the hypothesis that $\mu$ is limited by interface-roughness scattering (IRS) at high $n$ in the AlAs QWs that are thin enough to allow Z valley occupation.

Since the effects of IRS rapidly weaken with increasing QW thickness \cite{10}, we measured a second sample (B) that contains a slightly thicker AlAs QW (50 Å) but is otherwise identical to A. $\mu$ vs. $n$ for this QW is shown in Fig. 3 (circles). Indeed, $\mu$ attains a peak value of 5.5 m$^2$/Vs, higher than sample A, but then drops rapidly for higher $n$. This drop begins near the same $n$ where we observe Landau level crossings in $\rho_{xx}$ that we can attribute to the (X,Y) valley(s). The rapid drop in $\mu$, therefore, is likely associated with inter-valley scattering at the onset of (X,Y) valley population.

In order to side-step this mobility limiting dilemma, we devised a technique that allows preferential population of the Z valley in QWs thicker than $t_c$ (see Fig. 1). First, after the conventional processing described above (patterning, formation of the contacts, and front gate deposition) the sample is thinned to $\sim 150$ μm. We then deposit a back gate and glue the sample to a smooth block of fused silica (f-SiO$_2$) using a two-component epoxy and curing technique described previously \cite{10}. We used f-SiO$_2$ blocks cut from optical-quality mirrors with their metallization removed to ensure smoothness of the surface on which the sample is glued. The purpose of gluing the samples to f-SiO$_2$ is related to this material’s low thermal expansion coefficient ($\alpha_{f-SiO_2}$). As an ordinary sample is cooled, it contracts along all three directions in proportion to the thermal expansion coefficient of GaAs, $\alpha_{GaAs}$, which is 6.4 $\times 10^{-6}$/K at room temperature ($T_r$) \cite{11}. When thinned and affixed to f-SiO$_2$, however, the sample’s contraction follows the f-SiO$_2$ along the in-plane axes. Since $\alpha_{f-SiO_2} = 0.55 \times 10^{-6}$/K at $T_r$ \cite{12} is lower than $\alpha_{GaAs}$, this causes a temperature induced, biaxial, tensile strain in the plane of the QW and a compressive strain in the [001] direction along which the sample is free (Fig. 1). As a result, the (X,Y) valleys are raised in energy with respect to the Z valley.

To acquire an estimate of this energy and to further elucidate the technique, we consider the plane stress boundary conditions. We assume that the relevant Pois-
son ratios and Young’s moduli are temperature independent. Since the temperature at which our experiments are performed is low (0.3 K), we take it to be zero. The strain along each direction is:

\[ \epsilon_{xx} = (\sigma_{xx} - \nu \sigma_{yy})/E + \int_0^{T_f} \alpha_{GaAs}dT \]  (1)

\[ \epsilon_{yy} = (\sigma_{yy} - \nu \sigma_{xx})/E + \int_0^{T_f} \alpha_{GaAs}dT \]  (2)

\[ \epsilon_{zz} = -\nu (\sigma_{xx}/E + \sigma_{yy}/E) + \int_0^{T_f} \alpha_{GaAs}dT \]  (3)

where \( E \) is the Young’s modulus of GaAs along the crystal axes, \( \nu = 0.32 \) is the Poisson ratio, \( \sigma \) is stress, \( \epsilon \) is strain, and we have used the free boundary condition in the \( z \) direction, \( \epsilon_{zz} = 0 \). We have ignored edge effects since our active area is far from the sample edges. Setting \( \epsilon_{xx} = \epsilon_{yy} = \int_0^{T_f} \alpha_{f-SiO_2}dT \) and solving for the net strain between the \( Z \) and \( (X,Y) \) directions gives,

\[ \epsilon_{net} = \epsilon_{zz} - \epsilon_{xx} = \kappa \int_0^{T_f} (\alpha_{GaAs} - \alpha_{f-SiO_2})dT \]  (4)

with \( \kappa = (1+\nu)/(1-\nu) \). Utilizing the temperature dependent \( \alpha_{GaAs} \) [11] and \( \alpha_{f-SiO_2} \) [12], and the X-point deformation potential of AlAs, \( E_2 = 5.8 \) eV [13], we can calculate the valley splitting associated with the thermally induced strain: \( E_{net} = E_2 \epsilon_{net} \approx 9 \) meV. Subtracting this from the 23 meV of lattice mismatch induced valley splitting, we acquire the final strain-induced splitting between \( Z \) and \( (X,Y) \) valleys. This would yield an upward shift of \( t_c \) by approximately 20 Å, based on the confinement energy calculations of Ref. [8].

We have tested this technique on a third sample (C), a 60 Å QW that is otherwise structurally identical to the other two samples, and have indeed observed occupation of only the \( Z \) valley at low \( n \). Control samples taken from nearby sample C on the same wafer but not affixed to f-SiO\(_2\) show only \((X,Y)\) valley occupation, with no indication of the \( Z \) valley even at the highest attainable \( n \). This implies that the thermally induced strain in the sample glued to f-SiO\(_2\) is forcing the \( Z \) valley occupation, as expected. The \( \mu \) limiting effects of IRS appear to be significantly reduced in the 60 Å QW; as shown in Fig. 3 (squares), a peak \( \mu \) of 8.8 m\(^2\)/Vs is achieved. The improved quality is also evident in \( \rho_{xx} \) (Fig. 2), with SdH oscillations visible down to \( B = 0.4 \) T (\( \nu = 37 \)), and spin splitting resolved at a 30 % lower field than in sample A. At higher \( n \), there is evidence for \((X,Y)\) valley occupation, and consequently \( \mu \) begins to drop once again.

Although our experimental results are in qualitative agreement with the simple calculation given above, there is one complicating factor that we have neglected. The epoxy we have used to glue our samples to the f-SiO\(_2\) is known to creep at higher temperatures, and this effect is believed to persist down to about 200 K [14]. This will reduce somewhat the effectiveness of our technique, as the strain will not be fully transmitted over the entire temperature range. It may be desirable to use a glue that transmits strain more effectively at higher temperatures, or even to glue the sample to any of the novel materials that exhibit negative coefficients of thermal expansion, though there may eventually be a risk of cracking the sample for too large a strain.

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