Atomic configurations at InAs partial dislocation cores associated with Z-shape faulted dipoles

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The atomic arrangements of two types of InAs dislocation cores associated by a Z-shape faulted dipole are observed directly by aberration-corrected high-angle annular-dark-field imaging. Single unpaired columns of different atoms in a matrix of dumbbells are clearly resolved, with observable variations of bonding lengths due to excess Coulomb force from bare ions at the dislocation core. The corresponding geometric phase analysis provides confirmation that the dislocation cores serve as origins of strain field inversion while stacking faults maintain the existing strain status.

Polytype superlattices consisting of chemically identical but structurally different materials have attracted intense study in recent years, both for controlled growth of twinning and polytypic superlattices in nanos-structural materials1–4, and for exploring possible electronic and optoelectronic applications5–8. Because of the low stacking fault energy of cubic III-V compound semiconductors9, it is relatively easy to produce stacking faults on {111} planes, which then facilitate the formation of parallel twins and polytypic superlattices. Under certain conditions, twins and stacking faults on different {111} planes intersect and form different types of dislocation cores. Knowledge of the dislocation core structure can play a vital role in explaining the electronic and dynamical properties of defects in semiconductors. In this study, the atomic structures of two InAs dislocation cores, associated by a Z-shape faulted dipole, are directly observed, and single columns of unpaired atoms embedded in a matrix of atomic dumbbells are clearly resolved using aberration-corrected high-angle annular-dark-field (HAADF) imaging. Experimental measurement of the strain fields around the two dislocation cores are obtained based on geometric phase analysis (GPA).

Aberration-corrected HAADF imaging is a very powerful technique for determining structural as well as chemical information at atomic resolution. It has been applied, for example, to segregation and dislocation in a Ga-terminated 30° partial dislocation core in GaAs10, antisite defects in LiFePO411, the chemical identity of individual atoms in hexagonal BN monolayers that contain substitutional defects12, and the interfaces of InGaAs-InAlAs superlattices13. A recent paper describes applications to studying dislocation core structures in CdTe solar-cell material14. The utilization of aberration-corrected HAADF imaging in the current study has made it possible to determine the atomic structure of a Z-shape faulted dipole within the central area of an InAs nanopillar, that is comparatively thick (thickness of ~ 70 nm according to electron holography measurement). The overall structural features of the faulted dipole as well as atomic arrangements at the dislocation core were simultaneously obtained, and variations of bond length for pairs of atomic columns (“dumbbells”) close to the dislocation cores were also determined.

The Z-shape faulted dipole was first reported by Mader as faint straight lines along <110> directions in strong-beam electron micrographs of deformed f.c.c. metals15. This defect was suggested by Hirsch to be a long Frank loop16, and it was later discussed in detail by Seeger who proposed that the Frank dislocations were dissociated on the primary glide plane17. The Z-shape faulted dipole includes three stacking faults and four partial dislocations (i.e., two stair rods and two Shockley partials). The determination of stacking fault energy from the faulted dipole geometry using lattice imaging was reported18–19, and the atomic structure of the central
distortions that affect the images formed. The shear distortions in a scanning transmission electron microscope (STEM) scanning system and the CCD camera often have been corrected. HAADF images recorded at higher magnification are clearly resolved. The two dislocation cores indicated by blue and red arrows. (c) Aberration-corrected HAADF image of the faulted dipole with atomic dumbbells of the lattices and stacking sequences of the stacking faults clearly resolved. The dislocation cores are labeled with colored arrows similar to (b).

The locations of two dislocation cores are indicated by blue (core 1) and red (core 2) arrows. Stacking sequence analysis reveals that core 1 is formed by the interaction of one intrinsic stacking fault and two growth twins with staggered twinning planes, while core 2 is at the intersection of one extrinsic stacking fault and one intrinsic stacking fault. The area contained in the white box including both dislocation cores is filtered using the annular mask tool in Digital Micrograph to remove high frequency noise, and presented in Fig. 2b at higher magnification. The cation and anion atomic columns are clearly resolved with one column having much brighter contrast than the other. Energy-dispersive X-ray spectra (EDS) from the dislocation core as well as from a region free of defects were obtained and overlapped (Fig. S1). Both spectra show no sign of Sb, and the corresponding In and As peaks are almost coincident, with just some peak intensities of In and As at the dislocation core being slightly lower. Thus, it could be concluded based on the EDS results that there is no accumulation of Sb dopants at the dislocation core or else trace amounts only. Considering the thickness of the material (~70 nm), the introduction of Sb would not be expected to have marked impact on the image intensity. Thus, for simplicity we have labeled the observed dumbbells as In-As with brighter image contrast corresponding to In$^{3+}$ ($Z_{\text{In}} = 49$) and darker image contrast corresponding to As$^{5+}$ ($Z_{\text{As}} = 33$).

Both dislocation cores are composed of two columns of single atoms and four columns of dumbbells, where three dumbbells point towards the core with light atoms and one with heavy atoms. The results of intensity line profiles across two atomic columns outside the defect, and two atomic columns at the dislocation cores, as labeled by the white dashed arrows in Fig. 2b, are presented in Figs. 2c and 2d, respectively. The profile of core 1 has two peaks with average intensity level of $2.4 \times 10^5$ and two of $1.5 \times 10^5$, while all four peaks in the line profile of core 2 have intensity level of $2.4 \times 10^5$. Thus, the two columns of single atoms in core 1 should be As$^{5+}$, while the two columns of single atoms in core 2 should be In$^{3+}$. Close inspection of the atomic structure of core 1 reveals that the In-As bonding lengths of the dumbbells closest to the single As$^{5+}$ column are slightly shorter than normal (the average variation of In-As bonding length: $\Delta l \approx 0.05$ nm), as indicated by the black arrow heads in core 1, which could result from excess Coulomb repulsive force from the nearby single As$^{5+}$ columns. On the other hand, the In$^{3+}$ and As$^{5+}$ atomic columns of the dumbbells closest to the single In$^{3+}$ columns are clearly resolved, as indicated by the black arrow heads in core 2, even though the In-As bond lengths do not appear to be longer than normal. The error bar in the determination of the actual atomic positions is around 0.015 nm (see supplemental information for more details).

Atomic models for the regions outlined in white boxes in Fig. 2b are presented in Figs. 2e–2f; these are based on using the peak-finding algorithm to locate the atomic column positions and the relationship between intensity and elemental number: $Z_1 \sim Z_2$ (where $n$ is mainly determined by the geometry of the annular detection angles). The formation of core 1 can be attributed to the reaction of 30° Shockley partial dislocations bordering the stacking faults. Rotational twins with staggered twinning planes form a 30° Shockley partial dislocation with Burgers vector $b_{11} = (a/6)$ [121] (C8) lying in a (111) plane, which is indicated in Fig. 2e by a green triangle. This partial dislocation reacts with the other 30° partial with Burgers vector $b_{12} = (a/6)$ [211] (rC8) lying in a (111) plane indicated by an orange triangle. The overall Burgers vector of core 1 is obtained as $b_1 = b_{11} + b_{12} = (a/6)$ [110] (z0), which is perpendicular to the stacking faults of a Z-shape faulted dipole in deformed GaAs was determined based on HREM imaging and image simulations.

**Results**

The conventional TEM image in Fig. 1a shows many parallel dark lines perpendicular to [111] direction intersecting the other dark line perpendicular to [111] direction at 70.5°, forming the so-called Z-shape faulted dipole. Fig. 1b is an HREM image projected along [110] direction which indicates that the dark lines correspond to rotational twins and stacking faults. Since the twins and stacking faults which lie on (111) planes extend to the nanopillar surfaces, there are no partial dislocations at the far ends of the faulted dipole. Several stair-rod dislocations are pinned along the stacking faults on (111) planes, which is different from the classic Z-shape faulted dipole. Due to limited resolution (point resolution: 0.19 nm) of the microscope used for imaging, the atomic dumbbells of the lattice cannot be resolved (dumbbell of InAs crystal: 0.15 nm), and the stacking sequences of the stacking faults and atomic arrangements at the dislocation cores are not clear due to local strain as well as some image delocalization that occurs in a microscope without aberration correction. Fig. 1c is an aberration-corrected HAADF image with hardly any diffraction contrast and delocalization: the atomic dumbbells of the crystal lattice and stacking sequences of the stacking faults are clearly resolved. The two dislocation cores indicated by blue and red arrows are discussed in more detail below based on aberration-corrected HAADF images recorded at higher magnification.

It is well known that the scanning transmission electron microscopy (STEM) scanning system and the CCD camera often have distortions that affect the images formed. The shear distortions present in the aberration-corrected HAADF image in Fig. 2a have been removed, according to the perfect zincblende structure in [110] projection. In order to check to what extent the obtained image might have been affected by environmental instabilities, the Jitterbug software was applied to the image. However, it was found that the resulting STEM-HAADF image was almost unchanged from the original, which indicated that it was recorded under conditions of high instrument stability.
dislocation line in [110] direction, and verifies the formation of an edge dislocation. The geometric relationship of the above Burgers vectors and the dislocation line are sketched in Fig. 2g according to Thompson’s notation. The mechanism of dislocation formation for core 1 can be expressed as: αC + Cδ → αδ, which resembles part of the reported double-V shaped dislocation in diamond thin films, which has a core structure that is not as complex as this case. The formation of core 2 can be explained as follows. The leading fast partial dislocation Cδ, which is dissociated from a perfect 60° dislocation lying in a (111) plane, reacts with the stacking fault lying in a (111) plane. The incoming partial Cδ would dissociate into a stair-rod αδ and another partial dislocation Cα. While gliding on the (111) stacking-fault plane, the partial Cα starts to transform the intrinsic stacking fault into an extrinsic stacking fault, and finally join the partials at the bottom end. Thus, the formation mechanism of core 2 could be expressed as: Cδ → Cα + αδ, which has been discussed in detail as an example of the reaction between a dislocation and a stacking fault in nanocrystalline Al. Our study provides direct experimental evidence for the theoretical study but with much more information about the dislocation core.

Non-stoichiometric dislocation cores terminated by different atomic columns and associated by stacking faults were reported in α-Al2O3 based on bright-field STEM imaging technique, but later considered by Lagerlöf et al. as questionable. Heuer et al. did similar experiments, using the negative spherical aberration imaging technique combined with image simulations. It was found that the dislocation appeared to be stoichiometric and uncharged, and the dislocation cores exhibited 50% occupancy of the Al cations in order to achieve electrical neutrality. Thus, the determination of non-stoichiometric dislocation core is not straightforward, and further image simulation is necessary to verify the core stoichiometry. The models of the dislocation cores in this present study have been proposed according to the relative intensities of the atomic columns, and these intensities could be affected by the strain in the dislocation core. However, the stress fields of the edge-type dislocation cores should have no component along their dislocation lines: there might be collective displacements of atomic columns in the plane with the dislocation end-on, but no misalignment within single atomic columns at the dislocation core. Thus, the overall intensities of the atomic columns at the dislocation cores of edge-type should not be

Figure 2 | Dislocation cores with opposite ions at atomic resolution. (a) Aberration-corrected HAADF image of dislocation cores 1 and 2 as projected along the [110] direction. The location of dislocation cores are indicated by blue and red arrows, and the corresponding stacking sequences are overlapped. (b) Area contained in the white box in (a) shown at higher magnification. The atomic arrangements at dislocation cores and stacking faults are clearly resolved. The dumbbells nearest to the single atomic columns at dislocation cores with bond length variations are indicated by black arrow heads. The intensity line profiles at dislocation cores are labeled by white dashed arrows and the results are presented in (c) and (d), respectively. (e) and (f) Corresponding atomic models of the two regions framed by white boxes in (b). (g) Sketch of geometric relationship of the Burgers vectors and the dislocation line according to Thompson’s notation.
greatly affected by strain. Considering the large difference between the atomic number of In⁴⁺ (Z_In = 49) and As³⁻ (Z_As = 33), and the obvious intensity differences at the dislocation cores of the experimental image, it seems to be reasonable to determine the compositions of the single atomic columns in the dislocation cores in this way. Moreover, similar observations of unpaired Cd and Te atomic columns have recently been reported elsewhere³⁴.

Defects and associated long-range strain fields are of considerable importance in affecting the electronic properties of semiconductors, since the interaction of defects with charge carriers can occur far from defect cores³⁵. Geometric phase analysis (GPA) is a method of obtaining local phase which can be directly related to any displacement field. Moreover, similar observations of unpaired Cd and Te atomic columns have recently been reported elsewhere³⁴.

Discussion

Overall, our results have demonstrated that aberration-corrected HAADF imaging is a very powerful technique for determining atomic arrangements in and around dislocation cores. This approach has allowed us to distinguish between two types of dislocation cores which are part of a Z-shape faulted dipole, and variations of bond
lengths at the dislocation core are clearly resolved. Subsequent GPA analysis of the same area provides confirmation that the dislocation core serves as the origin of strain field inversion whereas the stacking fault acts to retain the existing strain. This combination of atomic structural information and strain distribution can be used as a starting point for studies of electronic properties of such defects within semiconductors, and these techniques should likewise be applicable to other types of defects in nanocrystals.

**Methods**

The InAs nanopillars, with nominal Sb concentration of 10%, were grown on nanopatterned substrates by metal-organic chemical vapor deposition in the catalyst-free growth mode (Detailed growth information will be published elsewhere). The nanopillars were deposited on copper grids with carbon support films for electron microscopy observation. Conventional transmission electron microscopy (TEM) and high resolution electron microscopy (HREM) were carried out using a Philips CM200-HEMI transmission electron microscope. The aberration-corrected HAADF observations were performed using a JEOL ARM200F TEM (camera length: 6 cm, convergence angle: 20 mrad, collection angle: 90–170 mrad). All microscopes were operated at an accelerating voltage of 200 kV.

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Author contributions

L.L. carried out the TEM experiments and wrote the paper, H.L. and H.Y. assisted the material growth, L.L., J.W. and D.S. designed the experiment. L.L., Z.G., M.M., H.Y., Y.G., J.W. and D.S. contributed to the data analyses and paper revisions.

Additional information

Supplementary information accompanies this paper at http://www.nature.com/scientificreports

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