Ion irradiation effects on metallic nanocrystals

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(Received 6 April 2007; in final form 8 May 2007)

We have investigated structural and morphological properties of metallic nanocrystals (NCs) exposed to ion irradiation. NCs were characterized by transmission electron microscopy in combination with advanced synchrotron-based analytical techniques, in particular X-ray absorption spectroscopy and small-angle X-ray scattering. A number of different effects were observed depending on the irradiation conditions. At energies where nuclear stopping is predominant, structural disorder/amorphization followed by inverse Ostwald ripening/dissolution due to ion beam mixing was observed for Au and Cu NCs embedded in SiO₂. The ion-irradiation-induced crystalline to amorphous transition in the NCs, which cannot be achieved in the corresponding bulk metals, was attributed to their initially higher structural energy as compared to bulk material and possibly preferential nucleation of the amorphous phase at the NC/SiO₂ interface. At very high irradiation energies (swift heavy ion irradiation), where the energy loss is nearly entirely due to electronic stopping, a size-dependent shape transformation of the NCs from spheres to rod like shapes was apparent in Au NCs. Our preliminary results are in good agreement with considerations on melting of the NCs in the ion track as one mechanism involved in the shape transformation.

Keywords: Nanocrystals; Ion irradiation; EXAFS; SAXS; Swift heavy ions

1. Introduction

Metallic nanocrystals (NCs) have demonstrated enormous potential for applications in a wide variety of scientific areas including medicine, catalysis, and materials engineering of electronic and optical devices. Their interesting properties, which can differ significantly from those of the corresponding bulk material, are generally related to the limited number of atoms within the NCs and the considerable surface-to-bulk ratio (SBR). This can lead to significant structural deviations as compared to ‘infinite’ bulk material and a concomitant alteration of the NC physical properties [1].

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Ion implantation is a very versatile technique and has widely been used for NC fabrication (ion beam synthesis) as well as for a structural modification of semiconductor-based materials (ion irradiation). While the influence of ion irradiation on the structure of bulk materials has been studied extensively and some understanding of the underlying mechanisms have been established [2], the influence of ion irradiation on NCs has only recently been investigated revealing a number of interesting phenomena. Simulations show that ion irradiation can significantly alter the evolution of the *size* distribution of NCs embedded in a matrix material, *i.e.* lead to narrowing due to inverse Ostwald ripening [3]. This was confirmed by several experiments on irradiated Au NCs using transmission electron microscopy (TEM) [4, 5]. At very high ion energies, where electronic stopping is dominant, a shape transformation from spheres to elongated rods was observed for irradiated Co NCs [6]. The influence of ion irradiation on the *structure* of metallic NCs, however, has remained a widely unexplored area.

In this paper, the effect of ion irradiation on *size*, *shape*, and *structure* of metallic NCs is investigated using a combination of synchrotron-based analytical techniques, in particular extended X-ray absorption fine structure (EXAFS) spectroscopy and small angle X-ray scattering (SAXS), with conventional laboratory-based techniques such as TEM. EXAFS provides a powerful tool for analysis of NCs giving accurate information about the average local atomic environment of the absorbing atom [7]. It is particularly sensitive to inter-atomic distances and local disorder and has successfully been utilized to resolve NC structural details [8]. SAXS and TEM give additional information about the NC shape and size distribution and thus enables a size-dependent interpretation of the structural data obtained by EXAFS.

2. Experimental

2.1 NC formation

Embedded Au and Cu NCs were produced by ion beam synthesis in 0.6–2 μm thin SiO_2 layers grown by thermal oxidation on Si (100) substrates. The amorphous SiO_2 layers were implanted with ^{197}Au or ^{64}Cu ions, respectively, to peak concentrations between 3 and 10 at.%. Subsequently, samples were annealed in a conventional quartz furnace to promote further precipitation and reduce irradiation-induced damage of the host material. Annealing temperatures were typically around 650 °C for Cu NC formation and 1000 °C–1100 °C for Au NC formation. While Cu implantations were annealed in forming gas atmosphere (95% N_2 and 5% H_2) to prevent Cu oxidation, Au implantations were annealed in O_2 (Au NCs do not oxidize when annealed in O_2). It has previously been shown that the annealing atmosphere has a profound influence on the resulting NC size [9] and the use of different atmospheres thus provides a means of tailoring the resulting NC size distribution. For the irradiation experiments presented here, Cu and Au NCs with average diameters between 2.5 and 12 nm were fabricated. TEM analysis reveals that NCs are spherical in shape. Details about the implant and annealing conditions can be found in previous publications [10–12].

2.2 NC irradiation

Ion irradiation was performed at different energies to study the influence of nuclear and electronic stopping on the NC morphology and structure. In the region where nuclear stopping

is predominant, here referred to as *low-energy* region, NCs were irradiated with ^{119}Sn ions at liquid nitrogen temperature at 5 MeV for Cu NCs and 2.3 MeV for Au NCs. Ion fluences ranged between 1×10^{14} and 1×10^{16} ions/cm². The energies were chosen such that essentially all Sn ions were stopped within the Si substrate, *i.e.* Sn-related impurity effects were negligible. Sn was used as its absorption edges are well separated from those of the NC metals (Au and Cu) and as such interference effects in the EXAFS measurements are avoided. However, other ion/energy combinations yielding similar energy losses are expected to produce the same results. Rutherford backscattering spectrometry of Au NC samples reveals that even at 1×10^{16} ions/cm², the Au concentration profile remains essentially unchanged upon irradiation [10]. Similar results have been observed for all irradiation fluences. To provide a reference for our structural analysis, metal thin films were evaporated on SiO₂ and subsequently covered with 100 nm of SiO₂ using plasma enhanced chemical vapor deposition. These standards were irradiated simultaneously with the NCs. To study the possible influence of impurities on the atomic structure of ion irradiated Au and Cu, we have implanted Si and O into the metal films to a concentration of ~ 8 at.% at liquid nitrogen temperature. Subsequently, those films were irradiated with Sn under identical conditions as the NCs.

In the here termed *high-energy* (swift heavy ion) region, Au NCs were irradiated at room temperature with 185 MeV Au ions to fluences between 2×10^{13} and 6.6×10^{14} ions/cm². At these irradiation conditions, the energy transfer of ~ 17 keV/nm to the SiO₂/NC layer is nearly entirely do to electronic stopping. Furthermore, it exceeds by far the threshold value of 2 keV/nm for latent track formation in amorphous SiO₂ [13].

2.3 Synchrotron measurements

To prepare the NC samples for the EXAFS and SAXS measurements, the thin SiO₂ layer containing the NCs was isolated by removing the Si substrate via mechanical polishing and selective wet chemical etching in a KOH solution. To obtain an improved fluorescence signal in the EXAFS measurements, multiple layers of the thin implanted SiO₂ were mounted together on the sample holder between two layers of kapton. SAXS and EXAFS measurements were performed on the same samples to ensure comparability.

Fluorescence EXAFS measurements at the Au L3 edge (11.919 keV) and Cu K edge (8.979 keV) were performed at the Photon Factory, Japan (beamline 20-B). Data were collected using a 10-element Ge solid-state detector. The fluorescence signal comprised between 10% and 50% of the incoming count rate, with the latter maintained well within the linear region of the detector. Measurements were performed at a temperature of 15 K to reduce thermal vibrations. The spectra were analyzed using the FEFFIT program package [14]. Structural parameters were extracted from the first coordination shell including the first three moments of the inter-atomic distance distribution [7]. Details about the data analysis can be found elsewhere [10–12].

Transmission SAXS measurements were carried out at the ChemMatCARS beamline 15ID-D at the Advanced Photon Source, USA using X-rays of 1.5 and 1.1 Å wavelength at a camera length of 1870 mm. Scattering images were recorded with a Bruker 6000 CCD detector at an exposure time of 1 s. Multiple images were recorded for each sample and averaged before analysis. An un-implanted SiO₂ film, prepared as described above, was measured to obtain the scattering contribution from the SiO₂ and kapton. For quantitative analysis, a maximum entropy method [15] was used to determine the size distributions from the scattering spectra.

3. Results and discussion

3.1 Structural properties of metallic NCs

Metallic NCs show structural differences to their corresponding bulk materials as a result of their finite size. We have previously studied such properties of Au and Cu NCs using EXAFS [12, 16–21]. Generally, we find that the bulk face-centered-cubic (FCC) structure is retained in the NCs, however, subtle structural differences are apparent from the EXAFS measurements. Common features of both NC systems as compared to bulk material are bond-length (BL) contraction, increased disorder, reduced coordination number (CN), and a BL distribution skewed towards shorter BLs. Figure 1 shows as an example the radial distribution function (RDF) of the first nearest neighbor reconstructed from EXAFS measurements [22] of a Au film (bulk) and 3.6 nm Au NCs [12]. Both increased disorder and reduced CN are a result of the significant SBR inherent to NCs. For example, in a 2.5-nm diameter NC, half of the atoms

Figure 1. Radial distribution functions reconstructed from EXAFS measurements showing the first nearest neighbor shell of a Au film (bulk) and 3.6 nm Au NCs.

Figure 2. Bond-length contraction of Au and Cu NCs as a function of inverse NC diameter. The solid lines show the corresponding fits using a simple liquid drop model.

reside at the surface. These atoms are under-coordinated, and as a consequence are expected to relax and/or reconstruct. The former yields a reduction in average CN and the latter leads to an increase in disorder expressed in the EXAFS Debye–Waller (D–W) factor. Furthermore, while the bulk RDF resembles a nearly Gaussian distribution, the NC RDF is asymmetric and skewed to shorter BLs. This is consistent with the general observation that relaxed/reconstructed metal surfaces contract. An asymmetric deviation from a Gaussian BL distribution is manifested as a non-zero third cumulant C_3 . The apparent (size dependent) BL contraction relative to bulk material can be explained surprisingly well in terms of capillary pressure due to the high SBR using a simple liquid drop (LD) model given by $\Delta R = -(4/3)K R_b f(1/D)$ [23], where K is the bulk compressibility, R_b the bulk metal BL, and f the surface tension. Figure 2 shows the BL contraction of Au and Cu NCs extracted from our EXAFS measurements as a function of inverse NC diameter [17, 18]. From these measurements, we have obtained values of 3.4 ± 0.5 [18] and $3.8 \pm 0.4 \text{ J/m}^2$ [17] for the surface tension of Au and Cu NCs, respectively. The value for the Au NCs is in very good agreement with previously measured values extracted from TEM and EXAFS [24, 25], whereas no comparable measurements have been reported for Cu NCs. We note that the macroscopic LD model can only be considered a crude approximation, however, molecular dynamics simulations [26] support the validity of this approach and are in good qualitative agreement with our results.

3.2 Low-energy ion irradiation of Cu NCs

The irradiation of metallic NCs in the low-energy regime is governed by nuclear collisions as the predominant energy loss mechanism. This can lead to structural and morphological changes in the NC composite material. At lower irradiation fluences, where mixing due to collision cascades is small, the structure of 2.5 nm Cu NCs irradiated with $1 \times 10^{14} \text{ ions/cm}^2$ Sn ions resembles striking similarities with that expected from amorphous material [11]. In contrast, bulk elemental metals cannot be rendered amorphous using ion irradiation (with one reported exception for Ga [27]). Figure 3 shows the RDFs of a (bulk) Cu film, unirradiated, and irradiated Cu NCs. The unirradiated Cu NCs show similar features when compared with the bulk standard as the Au NCs discussed above, *i.e.* BL contraction, increased D–W factor, negative C_3 , and reduced CN. When comparing unirradiated and irradiated NCs, a further decreased Cu–Cu CN, an increased BL and D–W factor and a non-zero, positive value of C_3 (the asymmetric deviation from a Gaussian inter-atomic distance distribution skewed toward

Figure 3. Radial distribution functions reconstructed from EXAFS measurements showing the first nearest neighbor shell of Cu thin film (bulk), and 2.5 nm Cu NCs before and after irradiation.

longer BLs) are apparent. *All* such changes are consistent with theoretical predictions from a dense random packing model for an elemental amorphous metal [28]. The size distribution of the NCs, however, remains nearly unchanged as apparent from SAXS measurements. Size distributions are shown in figure 4a and reveal a slightly increased volume-weighted average diameter after irradiation (~ 2.8 nm). This is consistent with our ion-fluence-dependent binary-collision simulations [29] of Sn-induced atomic mixing for a Cu layer of width 2.5 nm in SiO_2 as shown in figure 4b. Recoiling Cu, Si and O atoms altered the Cu depth distribution from step-like to Gaussian-like yielding the less-well-defined NC/matrix interfaces. The net effect of irradiation, viewed simplistically, was thus a reduction in diameter of the elemental Cu core from ~ 2.5 to ~ 2.2 nm surrounded by a Cu_2O shell of thickness ~ 0.6 nm. The fraction of Cu atoms in an oxidized state was quantified with X-ray absorption near-edge spectroscopy from which only Cu metal and Cu_2O environments were observed. As anticipated, the structural parameters of the Cu film extracted from EXAFS were unchanged upon irradiation. Furthermore, no structural changes were observed for the Cu film implanted with Si and O ions prior to and following irradiation. We thus conclude, that for these concentrations, Si and O impurities, potentially present as recoiled matrix atoms, do not initiate or enable amorphization of the bulk. Thermal annealing of the amorphous NCs under conditions insufficient for NC growth (350°C/h) induces recrystallization of the NCs indicated by nearly identical structural parameters observed for unirradiated and annealed samples [11].

We suggest several factors enabled the ion-irradiation-induced crystalline-to-amorphous phase transformation reported here for elemental Cu NCs. Before irradiation, the NCs were

Figure 4. (a) Size distributions of the Cu NCs before and after irradiation extracted from SAXS. (b) Binary-collision simulations of Sn-induced atomic mixing for a Cu layer of width 2.5 nm in SiO_2 .

clearly perturbed relative to the polycrystalline standard. Although the FCC structure was retained, the inherent structural disorder (D–W factor) of the NCs was twice that of the polycrystalline phase potentially lowering the energy barrier for the amorphization process. During irradiation, disorder at the NC/matrix interface can serve as a preferential site for the nucleation of the amorphous phase and/or stabilize such a phase subsequent to formation. Given half the Cu atoms reside at the interface, the influence thereof is necessarily great. This is in contrast to the irradiation of ~ 8 nm NCs where $<15\%$ of the atoms reside at the interface and amorphization was not observed [30]. Finally, the recombination of vacancies and displaced atoms would necessarily be inhibited for Cu atoms recoiled beyond the NC bounds yielding an increased defect concentration for the NCs relative to bulk polycrystalline material. Our observation of metallic NC amorphization follows that of an elemental semiconductor where we reported Ge NCs could be amorphized with an irradiation fluence nearly 100 times less than that required for its polycrystalline counterpart [31]. Evidently, finite-size effects render both semiconductor and metallic NCs significantly more sensitive to ion irradiation than the bulk phase.

3.3 Low-energy ion irradiation of Au NCs

For 5 nm, Au NCs structural parameters were measured for irradiation at fluences between 1×10^{15} and 1×10^{16} ions/cm² [10]. At these fluences, mixing effects become important and have to be taken into account. We have not only observed a change in the structural parameters measured by EXAFS, but also a significant change in the size distribution as apparent from figure 5. The evolution of the Au NC size distribution measured by SAXS is shown as a function of irradiation fluence. We note that TEM analysis confirms that NCs retain their spherical shape

Figure 5. Size distributions of Au NCs as a function of irradiation fluence extracted from SAXS.

upon irradiation consistent with rather isotropic collision cascades resulting from the nuclear stopping [32]. With increasing fluence, the NC size distribution becomes narrower and shift toward smaller average sizes. This is consistent with previous experimental results [4] and simulations on inverse Ostwald ripening, which involves smaller particles growing at the expense of larger ones [3]. As the NC structural properties are critically dependent on the NC size, we have to separate the effects of the irradiation on the NC structure from those only related to the reduction in size. Figure 6 summarizes the results extracted from EXAFS measurements for (a) the CN and (b) the BL for irradiated Au NCs as a function of average NC size [10]. As apparent from figure 5, the average NC size decreases with increasing irradiation fluence. The CNs of the unirradiated and irradiated NCs are shown in figure 6a along with an estimation given by $CN_{NC} = 12 \cdot (1 - 3/2D \cdot R_{nm})$ [33], where R_{nm} is the BL and D is the average NC diameter. Prior to irradiation both experiment and model are in good agreement. This has also previously been observed for a variety of unirradiated Au NCs [12, 18]. Upon irradiation a significant deviation from this model is apparent with a saturation of the CN at a value of ~ 4 . Figure 6b compares the BL variation of the Au NCs with that expected from the LD model described above. As for the CN, prior to irradiation measurements of the BL are in excellent agreement with the model. Upon irradiation (decreasing NC size), we observe an expansion of the BL with respect to the model for the lower irradiation fluences followed by a significant contraction for the higher fluences. Given the evaluated average NC sizes, the deviation of the BL and CN from the model assumptions yield information about irradiation induced effects. The BL expansion, reduction in CN, and increase in C_3 for the lower irradiation fluences show striking similarities to the behavior observed for the Cu NCs, indicating an amorphous phase is present in the Au NCs. We note, that similar to the Cu NCs, irradiated Au films with and without incorporation of O and Si impurities reveal no considerable structural changes with respect to the unirradiated film. The large decrease in the CN and BL for the NCs irradiated with higher fluences can be explained by the generation of monomers and small clusters. During the irradiation process, collision cascades lead to the observed evolution of the size distribution due to inverse Ostwald ripening. Au atoms are detached from larger NCs and then attach to smaller clusters or form nucleation sites for new clusters. This behavior has been predicted by kinetic Monte Carlo simulations, where the irradiation temperature is far below a critical temperature T_c [3] (T_c is defined as the temperature above which the NC size distribution widens due to conventional Ostwald ripening and below which it narrows due to inverse Ostwald ripening). The simulations predict a high solute concentration and the early stages of nucleation for temperatures far below T_c . Au monomers and small Au clusters (dimers, trimers, ...) are below the detection limit of TEM and SAXS. EXAFS, however, measures the average atomic environment around *all* Au atoms. Au monomers dissolved in the SiO₂ matrix will only affect the refined CN as they contribute to the absorption but not to the EXAFS oscillation originating from Au–Au bonding. The normalization of the EXAFS signal to the absorption will then cause an effective reduction in the EXAFS amplitude which appears as a reduction of the CN with respect to the bulk material. Small clusters with very low Au CNs (1 for dimers, 2 for trimers, ...) also yield a decrease in the measured average CN. The presence of both monomers and small clusters can explain the large decrease in the CN measured by EXAFS and the large difference compared to the model prediction. Small clusters additionally affect the simulated parameters for the first Au–Au coordination shell such as BL, D–W factor, and C_3 . The greater than expected reduction in BL (for the higher irradiation fluences) with the large deviation from the LD model is consistent with the formation of small Au clusters that were not detected by SAXS or TEM. A large number of small clusters with highly contracted BLs results in a BL distribution effectively skewed to small values. Using the difference between the CNs predicted by the equation above, CN_{NC} (using NC sizes extracted from SAXS) and the CN measured by EXAFS (CN_{EX}), we

Figure 6. (a) Au–Au CN and (b) BLs as a function of NC diameter as extracted from SAXS. The dotted line in (a) shows the expected CN for a spherical (FCC) NC. The solid line in (b) shows the BL contraction expected from a simple liquid-drop model using a surface tension value of $3.4 \pm 0.5 \text{ J/m}^2$ [18] (the dashed lines represent the associated error). For a given sample series, the NC size decreases as the irradiation fluence increases.

can estimate the volume fraction of Au present as monomers and small clusters by $V_{\text{rel}} = V_{\text{cl}}/V_{\text{tot}} = (\text{CN}_{\text{NC}} - \text{CN}_{\text{EX}})/(\text{CN}_{\text{NC}} - \text{CN}_{\text{cl}})$, where V_{cl} is the volume of Au in monomers and small clusters, V_{tot} is the total volume of Au atoms, and CN_{cl} the average Au CN of the small clusters (0 for monomers, 1 for dimers, . . .). Figure 7 shows the volume fraction of Au in monomers and small clusters for irradiation fluences of 1×10^{15} and 4×10^{15} ions/cm². For example, if only monomers are present ($\text{CN}_{\text{cl}} = 0$), $\sim 20\%$ of all Au atoms are dispersed

Figure 7. Estimated volume fraction of Au dispersed in the SiO₂ (as monomers and small clusters) as a function of the average CN of small clusters (monomers = 0) shown for selected irradiation conditions.

in the matrix for an irradiation fluence of 1×10^{15} ions/cm². If the dispersed Au was present as only dimers (CN_{cl} = 1) then this fraction would increase to ~25%. Naturally, we expect a mixture of cluster sizes together with a fraction of monomers. As apparent from figure 7, the fraction of dispersed Au increases with increasing irradiation fluence as expected due to the ongoing intermixing. For irradiation fluences $>4 \times 10^{15}$ ions/cm², where CN_{EX} saturates at ~4, the graph no longer changes significantly (not shown). However, the continuing BL contraction indicates a further increase in cluster formation and thus an ongoing increase in V_{rel} we note that these considerations are based on the assumption that the CN in the NPs is not altered due to structural changes induced by the irradiation.

3.4 High-energy irradiation of Au NCs

Irradiation of embedded metallic NCs with swift heavy ions can lead to a shape transformation from spherical to elongated rods first discovered by D'Orleans *et al.* in 2003 for Co NCs

Figure 8. Cross-sectional TEM images of Au NCs prior to and after irradiation with 185 MeV Au at different fluences.

[6]. In this section, we will present preliminary results on Au NCs and discuss aspects that yield further insight into the underlying mechanisms leading to the transformation. For our experiments, spherical Au NCs of 12 nm average diameter with a wide Gaussian-like size distribution with a full-width-half-maximum of 10 nm were irradiated with 185 MeV Au ions. Cross-sectional TEM images of unirradiated and irradiated NCs are shown in figure 8. We can clearly observe the shape transformation from spherical to ellipsoidal to rod-like shapes with increasing irradiation fluence. The elongation occurs in the direction of the incoming ion beam. Previously, a similar deformation of Au NCs was observed by Roorda *et al.* [34]. Figure 9 shows high-resolution TEM images of two NCs after irradiation with 2×10^{14} ions/cm². Apparent from both images is the polycrystalline nature of the NCs. The arrows on the right-hand side image indicate the direction of the lattice planes in two adjacent grains. Unirradiated NCs on the other hand are generally single crystalline. The images in figure 9 also clearly show the existence of small NCs (~ 1 nm) evenly distributed around the large ones. Consistent with previous considerations about the mechanisms responsible for the shape transformation [32], our results support the presumption that melting of the NCs in the ion track is involved in the deformation process. The expected rapid resolidification can explain the polycrystalline nature of the NC as well as the appearance of the smaller NCs. The latter can be caused by

Figure 9. High-resolution cross-sectional TEM images of Au NCs after irradiation with 185 MeV Au to a fluence of 2×10^{14} ions/cm².

Figure 10. Minor axis length of the NCs versus the major axis length for unirradiated NCs (open squares) and NCs irradiated with 6.6×10^{14} ions/cm² (open circles) extracted from TEM analysis. The dashed line indicates the diagonal through the diagram with $D_{\text{minor}} = D_{\text{major}}$, the solid line the elongation trend for the irradiated NCs.

the increased solubility of the metal in the liquid SiO₂ track and the formation of the small clusters upon resolidification. Our measurements also show a strong size dependence of the deformation process. This is displayed in figure 10, which shows the minor axis versus the major axis length for unirradiated NCs and NCs irradiated with 6.6×10^{14} ions/cm² extracted from TEM analysis. As expected from spherical particles, all unirradiated NCs (open squares) are located on the diagonal through the diagram (indicated by the dashed line). The irradiated particles (open circles) show no measurable deformation up to a threshold value of ~ 7 nm above which elongation is observed with aspect ratios up to ~ 5 for the given fluence. Our measurements also indicate that NCs below this critical diameter dissolve into the matrix upon irradiation. The observed minimum NC diameter coincides approximately with the ion track width calculated from the thermal spike model [35], which is also consistent with the above-mentioned considerations [32]. Further investigation is currently in progress.

4. Conclusions

In conclusion, we have studied the effect of ion irradiation on the size, shape, and structural evolution of ion beam synthesized Cu and Au NCs using a combination of TEM, EXAFS, and SAXS. We have demonstrated evidence for ion-irradiation-induced amorphization of Au and Cu NCs in the energy regime governed by nuclear stopping. This was attributed to structural perturbations in the NCs prior to irradiation and possible preferential nucleation of the amorphous phase at the NC/matrix boundaries. At higher irradiation fluences, the dissolution of Au NCs and the formation of small clusters was observed. At high energies (swift heavy ion irradiation), a shape transformation in Au NCs from spherical to rod-like was observed characterized by a minimum NC size required for the transformation and a change from single-crystalline to polycrystalline morphology of the NCs. Our results present potential applications in tuning the optical properties of the embedded NCs and as such they are interesting for utilization in advanced micro-optical device fabrication.

Acknowledgements

P.K. and M.C.R. thank the Australian Research Council for support. The authors' were supported by the Australian Synchrotron Research Program, which is funded by the Commonwealth of Australia under the Major National Research Facilities Program. ChemMatCARS Sector 15 at the Advanced Photon Source is principally supported by the National Science Foundation/Department of Energy under grant number CHE0087817 and by the Illinois Board of Higher Education. The Advanced Photon Source is supported by the US Department of Energy, Basic Energy Sciences, Office of Science, under Contract no. W-31-109-Eng-38.

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