

## Effect of surrounding environment on atomic structure and equilibrium shape of growing nanocrystals: gold in/on SiO<sub>2</sub>

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**Abstract** We report on the equilibrium shape and atomic structure of thermally-processed Au nanocrystals (NCs) as determined by high resolution transmission electron microscopy (TEM). The NCs were either deposited on SiO<sub>2</sub> surface or embedded in SiO<sub>2</sub> layer. Quantitative data on the NCs surface free energy were obtained via the inverse Wulff construction. Nanocrystals inside the SiO<sub>2</sub> layer are defect-free and maintain a symmetrical equilibrium shape during the growth. Nanocrystals on SiO<sub>2</sub> surface exhibit asymmetrical equilibrium shape that is characterized by the introduction of twins and more complex atomic defects above a critical size. The observed differences in the equilibrium shape and atomic structure evolution of growing NCs in and on SiO<sub>2</sub> is explained in terms of evolution in isotropic/anisotropic environment making the surface free energy function angular and/or radial symmetric/asymmetric affecting the rotational/translational invariance of the surface stress tensor.

**Keywords** Nanocrystal · Crystal growth · Transmission electron microscopy · Wulff plot · Surface energy · Gold · SiO<sub>2</sub>

### Introduction

The discovery of nanostructures (quantum dots, nanoparticles, quantum wires, nanowires, etc.) have stimulated research in many different areas, becoming the basis for the “Nano-electronic and Nano-technology revolution” [1–3]. The fundamental issue of such a revolution is the effort towards the fabrication of innovative nano-structured functional materials and, in particular, the acquisition of the capability to control all their characteristics (electrical, optical, mechanical, etc.) by the atomic-level manipulation of the structural ones.

In particular, nano-structured materials fabricated by metal nanocrystals (NCs) dispersed in or deposited on insulator matrices have been extensively studied [4–7] because their unique optical, magnetic and electrical properties and thanks to the possibility of control such properties in a wide range simply controlling the NCs structural properties (atomic structure, size, density, etc.). It is well-known that the NCs characteristics are size-dependent and intermediate between the properties of the atomic and bulk matter [4, 8]: such a peculiar property results from the NCs nanometric dimensions and in particular from the enhanced surface-area-to-volume ratio (with respect to the bulk matter) and thus vary as a function of NC size (as a consequence of the limited number of atoms incorporated therein and the considerable influence of surface atoms) [9]. For these reasons, nano-structured materials based on metal NCs acquired a relevant role for the design and fabrication of nano-devices with numerous applications ranging from micro- and nano-electronics to catalysis, gas sensor, biology and many others. On the other hand, in view of the fabrication of innovative nano-devices using metal NCs in/on insulator substrates with pre-determined tunable (electrical, optical,...) properties, it

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is fundamental to know the details of the chemical and physical interaction between the NCs and the surrounding environment to understand (and to control) the structural ones and their evolution as a function of the NCs size.

In the present work, we study the influence of the surrounding environment on the atomic structure and equilibrium shape of growing NCs focusing on Au NCs/SiO<sub>2</sub> system. We choose such a system because its potential application in the fabrication of nano-electronic, optoelectronic (plasmonics) and gas-sensor innovative devices [10–12]. On the other hand, exists from 80s a vast experimental and theoretical literature concerning the study of free and supported Au NCs [9, 13–24]. The experimental works are primarily based on the study of the Au NCs structure and equilibrium shape by transmission electron microscopy (TEM), while the theoretical ones on simulation of the atomic structures and equilibrium shapes by several models (Monte-Carlo, Density functional theory). The present work, gives a contribution to the experimental investigation evidencing clear differences (by TEM analyses) in the atomic structure and equilibrium shape of growing Au NCs embedded in SiO<sub>2</sub> layers from those supported on SiO<sub>2</sub> and can furnish new theoretical basis to include the effect of the environment surrounding the NCs in the simulation models. An explanation of the observed differences in terms of thermodynamic concepts and symmetry considerations is given. In particular, we demonstrate how to extract quantitative information about the surface free energy of the Au NCs grown in and on SiO<sub>2</sub> using the inverse Wulff construction basing on the equilibrium shape inferred by high-resolution transmission electron microscopy (HR-TEM). Furthermore, the method to growth ideal NCs on surfaces come out and the conclusions can be extended to other crystalline nanoclusters of other materials.

## Experiment

To avoid any possible difference in the Au NCs structures and shapes as a consequence of differences in samples preparation, we followed a unique procedure to prepare the samples with NCs inside and over amorphous SiO<sub>2</sub>. CZ- <100> silicon substrates with resistivity  $\rho \approx 6 \times 10^{-3} \Omega \text{ cm}$  were etched in a 10% aqueous HF solution to remove the native oxide, and subsequently annealed at 1223 K for 15 min in O<sub>2</sub> in order to grow an uniform, 10 nm thick, amorphous SiO<sub>2</sub> layer (as determined by ellipsometric measurements and subsequent cross-sectional TEM). A 2 nm thick Au film was deposited (at room temperature) by sputtering on the SiO<sub>2</sub> layer using an Emitech K550x Sputter coater apparatus (Ar plasma, 10<sup>-4</sup> Pa pressure). In some samples the Au layer was

covered by a 3 nm thick SiO<sub>2</sub> layer deposited (at room temperature) by sputtering using an AJA RF Magnetron sputtering apparatus (Ar plasma,  $1.3 \times 10^{-8}$  Pa pressure). Then, the Au/SiO<sub>2</sub> and SiO<sub>2</sub>/Au/SiO<sub>2</sub> samples were contemporary annealed in Ar ambient in the 873–1073 K temperature range and in the 5–60 min time range to obtain the NCs self-organization and growth.

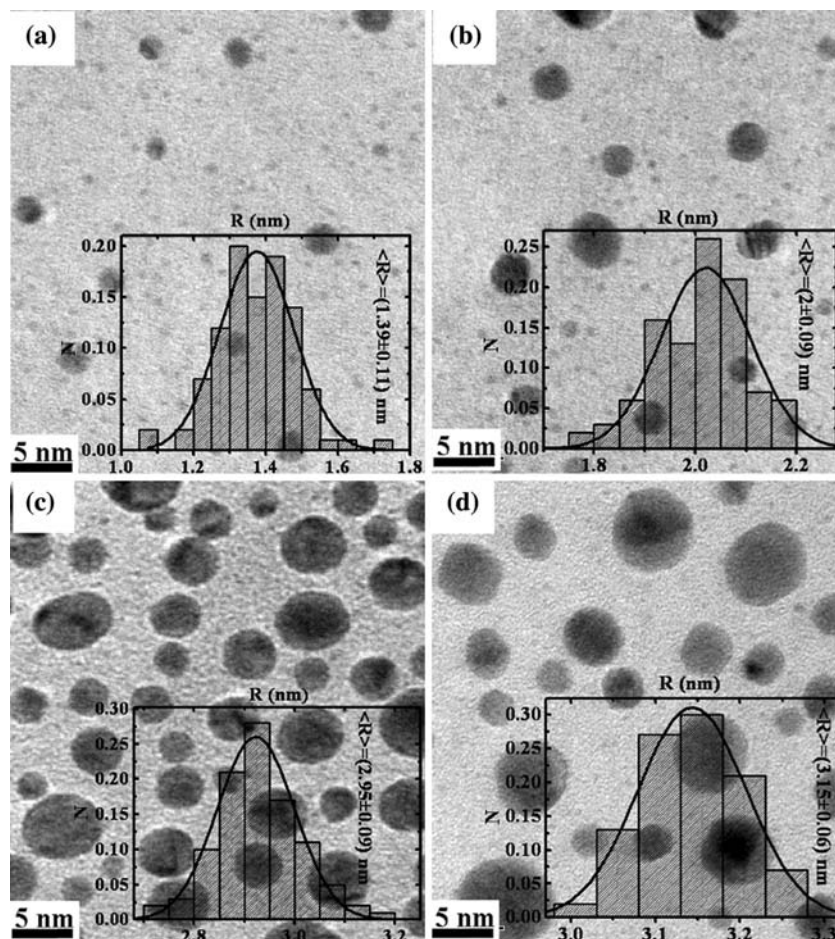
The samples were analyzed by Rutherford backscattering spectrometry (RBS), TEM and HR-TEM, after mechanical polishing and final Ar ion milling. The RBS analyses were performed using 2 MeV <sup>4</sup>He<sup>+</sup> backscattered ions at 165°. The TEM analyses were performed by a Jeol 2010F energy filtered transmission microscope (EFTEM), operating at 200 kV and equipped with a Gatan image filtering apparatus. EFTEM allowed to map of the elemental composition of the samples with spatial resolution typical of TEM analyses, by selecting electrons with a well defined energy from the transmitted beam. Before the cross-sectional TEM analyses of the samples with Au on SiO<sub>2</sub>, a thin SiO<sub>2</sub> cap-layer (~3 nm) was deposited (by sputtering with the AJA RF Magnetron apparatus) on the samples in order to protect the NCs during the samples preparation. It is worth to note that this last deposition was performed at room temperature, so that we can assume that the Au NCs do not change their atomic structure and equilibrium shape in consequence of the SiO<sub>2</sub> protective-layer deposition because it is quite improbably that they release energy during such a process.

## Results

Rutherford backscattering spectrometry analyses allowed to determine the deposited Au atomic surface density  $Q$ : for all the samples (Au in and on SiO<sub>2</sub>, as-deposited and annealed) it is the same (within a statistical error of 5%) and equal to  $Q \approx 9 \times 10^{15} \text{ Au/cm}^2$ . So, we can conclude that no Au loss occurs during thermal treatments (evaporation, out-diffusion).

Plan- and cross-sectional TEM analyses allowed to follow the evolution of the NCs size distribution as a function of annealing temperature and time. As an example, Fig. 1 shows plan-view TEM images for some annealed samples. Figure 1a and b refer to the samples annealed at 873 K for 20 min but with Au NCs within and over the SiO<sub>2</sub> layer, respectively. Figure 1c and d show the images for the Au NCs in and on SiO<sub>2</sub> annealed at 1073 K for 60 min. The inset of each figure shows the derived NCs size distribution, reproduced by a Gaussian fit (continuous line), and the relative mean radius  $\langle R \rangle$ , derived as the position of the Gaussian peak. The error on  $\langle R \rangle$  was evaluated as the standard deviation  $\sigma$  (1/2 of the full width at half maximum) of the Gaussian fit. The systematic

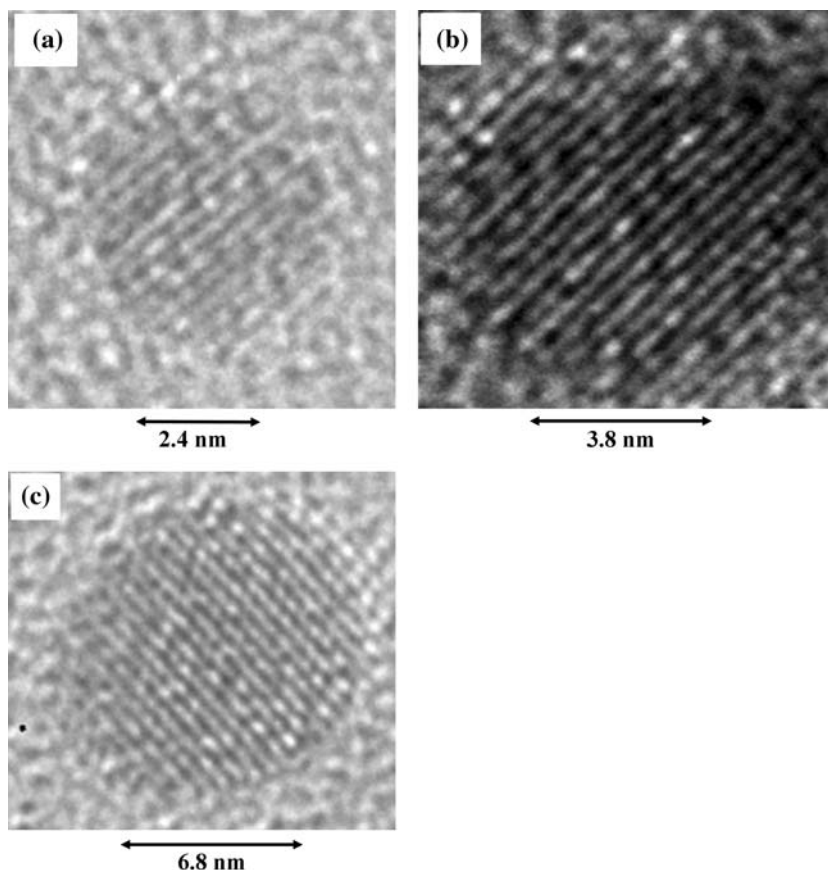
**Fig. 1** Plan-TEM images of the samples (a) with Au NCs inside the SiO<sub>2</sub> layer and annealed 873 K–20 min, (b) with Au NCs over the SiO<sub>2</sub> layer and annealed 873 K–20 min, (c) with Au NCs inside the SiO<sub>2</sub> layer and annealed 1073 K–60 min, (d) with Au NCs over the SiO<sub>2</sub> layer and annealed 1073 K–60 min. The inset of each figures show the obtained NCs radius distribution (N is the normalized number of NCs and R the NCs radius,  $\langle R \rangle$  the mean NC radius)



behavior observed consists in an increase of the mean NCs radius (translation of the radius distribution towards higher sizes) and a decrease of  $\sigma$  (narrowing of radius distribution) when annealing temperature and/or time are increased. The systematic analyses of the evolution of such a distribution as a function of annealing temperature and time allowed to elucidate the kinetic growth mechanism (in the analyzed annealing temperature and time range) of the Au NCs in and on SiO<sub>2</sub> in a ripening process of three-dimensional structures limited by diffusion but with different growth rates. In particular, a lower growth rate for the NCs embedded in SiO<sub>2</sub> was found with respect to the growing NCs on SiO<sub>2</sub> as evident from the TEM images and distributions in Fig. 1 (and considering that the starting distributions for the NCs in and on SiO<sub>2</sub> are identical). In particular, we determined the increase of the mean radius  $\langle R \rangle$  from  $(1.24 \pm 0.16)$  nm to  $(2.95 \pm 0.09)$  nm for the Au NCs in SiO<sub>2</sub> and to  $(3.15 \pm 0.06)$  nm for those supported on SiO<sub>2</sub> (in the same range of annealing temperature and time). The detailed study of the kinetic growth mechanisms of Au NCs on SiO<sub>2</sub> are reported by Ruffino et al. [25] and for the Au NCs in SiO<sub>2</sub> by De Marchi et al. [26] and Miotello et al. [27].

NCs internal structures and morphologies were examined using Fresnel contrast and HR-TEM analyses. The first result was obtained by analyses on significant statistical population of NCs (were analyzed 100 NCs for each sample): their internal atomic structures and shapes are determined only by their size and not by the particular annealing process performed. This fact means that each sample is characterized by a distribution of NCs size (which mean value increases when annealing temperature and time increase) but in all the samples with Au NCs inside SiO<sub>2</sub>, independently from the temperature, the smaller NCs have a similar structure between them, the greater NCs are a similar structure between them but different with respect to the smaller ones. The same situation occurs for the NCs over SiO<sub>2</sub> but in a different manner from those grown inside the SiO<sub>2</sub> layer. In particular, for the Au NCs embedded in SiO<sub>2</sub>, the HR-TEM images indicate that, in the range of radius  $1 \text{ nm} < R < 3.5 \text{ nm}$  the NCs always exhibit a single-crystal-icosahedral structure (close-packing of atoms structure) [14] free of internal defects as indicated by the representative HR-TEM images in Fig. 2a, b and c (more complex different structures, such as Marks decahedra or cuboctahedra, are evident for higher

**Fig. 2** HR-TEM images of Au NCs in SiO<sub>2</sub> with the same size as the NCs on SiO<sub>2</sub> forming the three determined groups



sizes). These images refer to the samples as-deposited (a), annealed 873 K–60 min (b) and 1073 K–60 min (c), respectively. They show a good resemblance with the proposed model for Au NCs structure viewed along (110) [13]. The spacing between the crossing (111) planes is 0.235 nm in this projection and the Au-Au distance is 0.25 nm. The fact that all the NCs in SiO<sub>2</sub>, with size in this range, exhibit single-crystal-icosahedral structure, free of internal defects reflects a situation in which their equilibrium shape symmetry is conserved when they grow.

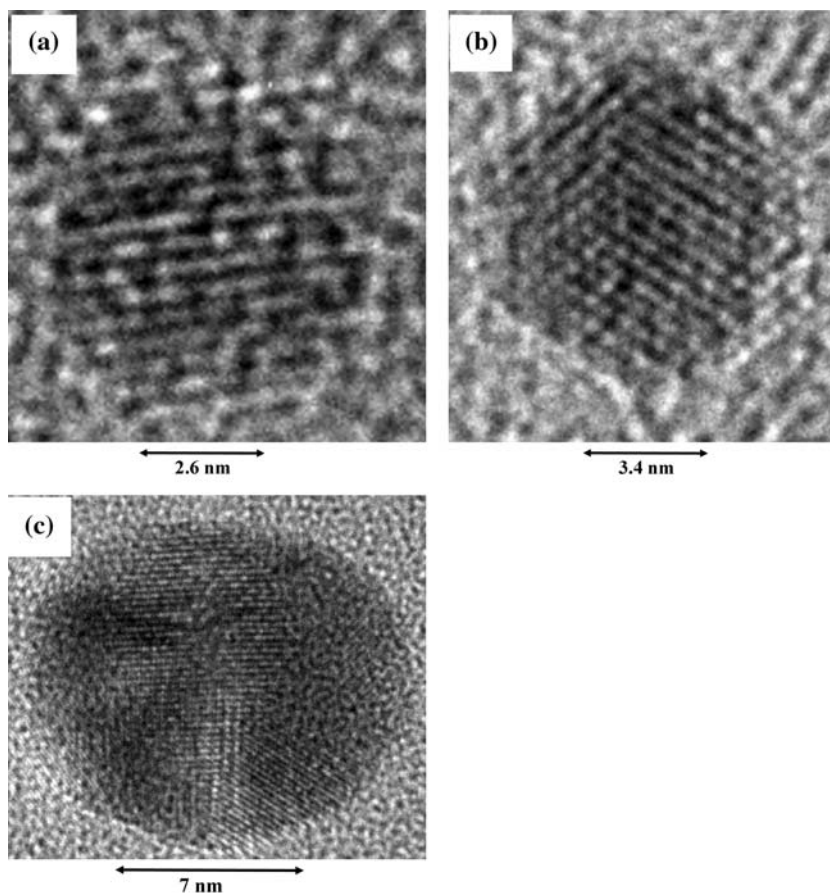
Also the HR-TEM images of the NCs deposited and annealed over SiO<sub>2</sub> show a good resemblance with the proposed model for Au NCs structure viewed along (110) but, differently from the NCs growing in SiO<sub>2</sub>, the data seem to delineate three large groups of structures, determined by the NCs size (in the same size range 1 nm < R < 3.5 nm):

- Group 1 (a representative NC in Fig. 3a): it is formed by NCs with a radius  $R < 1.5$  nm; they have single-crystal-icosahedra structure [14] free of internal defects;
- Group 2 (representative NC in Fig. 3b) it is formed by NCs with a radius  $1.5 \text{ nm} < R < 2$  nm; they have a (111)-twinned icosahedral structure [14];
- Group 3 (representative NC in Fig. 3c): it is formed by NCs with a radius  $2 \text{ nm} < R < 3.5$  nm; they have a

complicate decahedral multi-twinned (and lamellar) structure [14].

From the representative images reported in Fig. 3 (that refer to the samples as-deposited (a), annealed 873 K–60 min (b) and 1073 K–60 min (c), respectively) it is clear that the evolution in size of the supported NCs on SiO<sub>2</sub> corresponds to an evolution of the equilibrium shape with a progressive loss of symmetry and a corresponding introduction of internal defects. In this figure, were reported representative images of NCs with size similar to those reported in Fig. 2 to make a direct comparison: the NC in Fig. 3a (representative of the group 1 of growing NCs on SiO<sub>2</sub>) has an atomic structure (single icosahedral) coincident with those showed in Fig. 2a, b and c (growing NCs in SiO<sub>2</sub>); images in Fig. 3b and c show NCs representative of the growing NCs on SiO<sub>2</sub> belonging to the group 2 and 3, respectively: it is evident the different atomic structure in comparison with the NCs grown in SiO<sub>2</sub> and with similar sizes (Fig. 2b and c). The NCs of group 2, despite the similar single-crystal-icosahedral structure as the NCs of group 1, are characterized by the introduction of twins of (111) planes corresponding to introduction of asymmetry in the NC shape. For the NCs of group 3 occurs a transition from the icosahedral structure to the decahedral

**Fig. 3** HR-TEM images of Au NCs on SiO<sub>2</sub> for the three determined groups ((a) relative to the group 1, (b) to the group 2 and (c) to the group 3)



one, characterized by the introduction of multi-twins and completely lost shape symmetry.

## Discussion

To explain the differences observed in the atomic structure and equilibrium shape of Au NCs with similar size but grown inside and over the amorphous SiO<sub>2</sub> layer, in this section we illustrate a discussion based on thermodynamic and symmetry considerations. In particular, we base our discussion on the concept of angular and radial symmetry/asymmetry of the NC surface free energy corresponding to a symmetrical/asymmetrical spatial situation of the environment surrounding the NC. Furthermore, we correlate the angular/radial symmetric/asymmetric behaviour of the NC surface free energy with the rotational/translational invariance of the surface stress tensor of the NC.

In the bulk form Au has the face centred cubic (FCC) structure which is closest packed. As a consequence Au macro-crystals exhibit the symmetric cubic, octahedral or rhombododecahedral crystal form associated with the FCC structure. For the Au NCs the situation is more complicated because the strongly influence of the surface atoms. The

FCC Au based NCs generally present to the environment as a combination of crystal facets of the {111} and {100} planes in icosahedral or decahedral structure that yield to an efficient compromise between surface area and packing [28]. It is known, furthermore, that the surface of Au shows a tendency to reconstruct so that the surface atoms are closely packed together [29]. This behaviour could be due to the relief of large tensile surface stress in the unreconstructed surface. The most striking reconstruction was observed in the (111) plane [30] (that is expected to be the closest packing plane, the plane with the lowest surface free energy). Such a plane forms a  $23 \times \sqrt{3}$  reconstruction which appears to involve the insertion of an extra row of atoms into the surface every 23 rows [31]. This reconstruction clearly decreases the average coordination number of the atoms because to form it atoms must be taken from 12-fold-coordinated bulk sites placed in 9-fold-coordinated surface sites. Such a reconstruction could only be lower in energy than the unreconstructed surface if the shorter bond-lengths that are formed in the plane of the surface are energetically favourable. Then, the unreconstructed Au (111) surface exhibit a considerable tensile surface stress and the driving force for the  $23 \times \sqrt{3}$  reconstruction is the relief of this tensile surface stress.

In our discussion, we, according to Shuttleworth [32] and Cammarata [33], define the second rank surface stress tensor  $f_{ij}$  of the NC by:

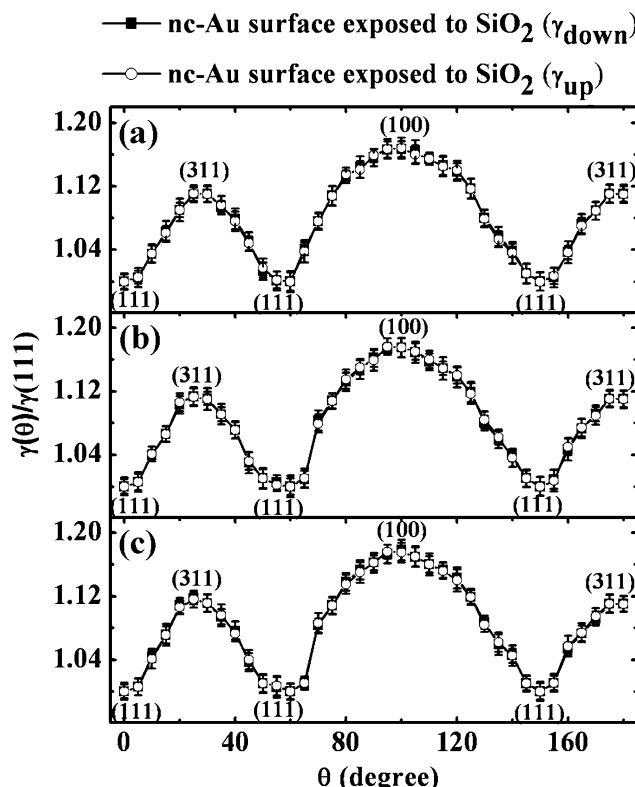
$$f_{ij} = \gamma \delta_{ij} + \frac{\partial \gamma}{\partial \varepsilon_{ij}} \quad (1)$$

being  $\gamma$  the surface free energy (per unit area) of the NC,  $\varepsilon_{ij}$  the surface elastic strain tensor,  $\delta_{ij}$  the Kronecker delta.

Then, we extracted quantitative data on the surface free energy from the NCs shapes via the inverse Wulff construction [34]: first, we identified the ‘‘Wulff point’’ of the NC and then applied the Wulff construction in reverse order to obtain the surface energy plot to within a constant scale factor. The Wulff point (‘‘center of mass’’) of a NC was defined as the intersection point between the perpendicular bisectors of two (111) facets. This point was taken as the centre of a polar plot, with angular and radial coordinates  $\theta$  and  $r$ . For each  $\theta_i$  (with a 5° spacing), we measured the distance  $r_i$  between the Wulff point and the NC surface (determined by the accurate analyses of the HR-TEM images). According to the Wulff relation  $\gamma_i / r_i = \lambda = const$ , the value of the surface free energy  $\gamma_i \equiv \gamma(\theta_i)$  for those orientations was obtained. In this way the  $\gamma(\theta)$  plot was determined. In particular, to make a direct comparison between the NCs on and in SiO<sub>2</sub>, we distinguished the NC surface in two parts. We draw the line passing through the Wulff point and parallel to the substrate plane (using, in particular, cross-TEM images). For the NCs on SiO<sub>2</sub>, the NCs surface above this line was defined as the surface exposed to the Ar (the gas environment in which were performed the annealing processes), while the below one as the surface exposed to SiO<sub>2</sub>. In the same way for the NCs in SiO<sub>2</sub>, but considering that the upper and lower parts of surface are anywhere exposed to SiO<sub>2</sub>. According to this consideration, we named  $\gamma_{up}(\theta)$  the NC surface energy relative to the upper surface part and  $\gamma_{down}(\theta)$  the NC surface energy relative to the bottom part, so that for the NCs on SiO<sub>2</sub> is  $\gamma_{up}(\theta) \equiv \gamma_{Au/Ar}(\theta)$  and  $\gamma_{down}(\theta) \equiv \gamma_{Au/SiO_2}(\theta)$ , while for the NCs in SiO<sub>2</sub> is  $\gamma_{up}(\theta) \equiv \gamma_{Au/SiO_2}(\theta)$  and  $\gamma_{down}(\theta) \equiv \gamma_{Au/SiO_2}(\theta)$ .

For each class of NCs shapes (Fig. 2 and Fig. 3), an averaged  $\gamma(\theta)$  plot is shown in Fig. 4 and Fig. 5, where the relative energies are normalized to the surface free energy for the (111) plane. For all the plots the zero in the angle scale was taken as the direction of one (111) plane of the NC. However, the  $r_i$  values for each  $\theta_i$  are averaged over 10 different NCs belonging to the same group (i.e. similar size). The NC-to-NC variation in  $\gamma(\theta)$  for a given orientation was indicated as the error bars reported in Fig. 4 and 5.

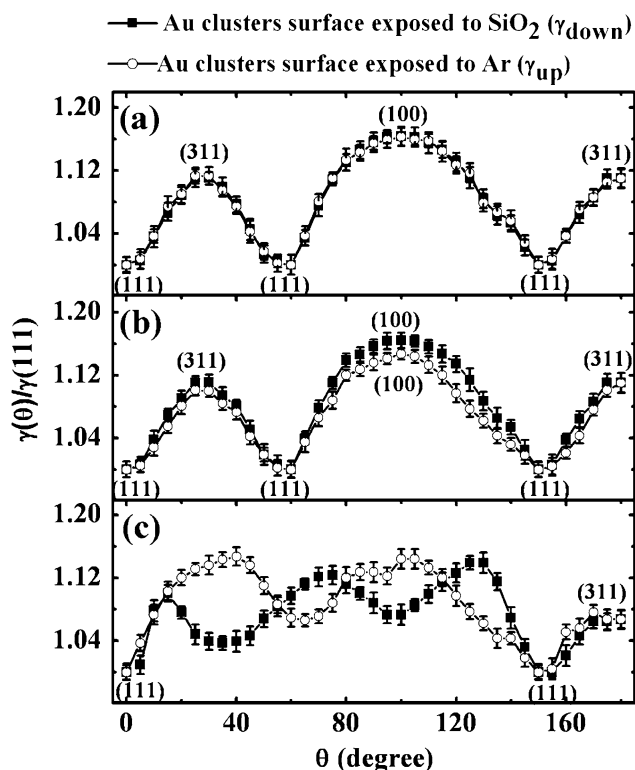
First we consider the case of Au NCs in SiO<sub>2</sub>: from Fig. 4a, it is clear that the peaks and valleys in the  $\gamma_{up}(\theta)$



**Fig. 4** Wulff plot ( $\gamma_{up}(\theta) = \gamma_{Au/SiO_2}(\theta)$  and  $\gamma_{down}(\theta) = \gamma_{Au/SiO_2}(\theta)$ ) for the NCs in SiO<sub>2</sub>, with the same size as the NCs on SiO<sub>2</sub> forming the three determined groups

and  $\gamma_{down}(\theta)$  lie in identical angular position (angular symmetry of  $\gamma(\theta)$ ) and assume identical values. The latter point means that the facets with the same Miller indexes in the upper and lower parts of the NC, respectively, have identical radial coordinates (radial symmetry of  $\gamma(\theta)$ ). Both the angular and radial symmetry of  $\gamma(\theta)$  are conserved with increasing the cluster size in the cases of NCs in SiO<sub>2</sub>, as it is evident from the comparison of Fig. 4a, b and c. This fact is reflected in the invariance of  $f_{ij}$  both for translational and rotational transformations and, as a consequence, in the symmetrical single crystal icosahedral structure for the equilibrium shape of NCs in SiO<sub>2</sub>. Those NCs grow subjected always only to the thermodynamic process determined by the Au/SiO<sub>2</sub> interface in any direction, i.e. in an ‘‘isotropic spatial situation’’.

Now we consider the case of Au NCs on SiO<sub>2</sub>. Fig. 5a shows, for the smaller sizes (group 1), a situation similar to the case of NCs in SiO<sub>2</sub>, i.e.  $\gamma_{up}(\theta)$  and  $\gamma_{down}(\theta)$  lying in identical angular position and assuming identical values ( $\gamma_{up}(\theta) \equiv \gamma_{Au/Ar}(\theta) \approx \gamma_{down}(\theta) \equiv \gamma_{Au/SiO_2}(\theta)$ ): both angular and radial symmetry of  $\gamma(\theta)$  is present ( $f_{ij}$  is again invariant both for translational and rotational transformations). However, differently from the NCs in SiO<sub>2</sub>, for sizes relative to the group 2, the angular symmetry of  $\gamma(\theta)$  is still conserved, but the radial one begins to be lost



**Fig. 5** Wulff plot ( $\gamma_{up}(\theta) = \gamma_{Au/Ar}(\theta)$  and  $\gamma_{down}(\theta) = \gamma_{Au/SiO_2}(\theta)$ ) for the three groups of NCs on  $SiO_2$  ((a) relative to the group 1 and representative NC in Fig. 3a, (b) to the group 2 and representative NC in Fig. 3b, (c) to the group 3 and representative NC in Fig. 3c)

( $\gamma_{up}(\theta) \equiv \gamma_{Au/Ar}(\theta) \neq \gamma_{down}(\theta) \equiv \gamma_{Au/SiO_2}(\theta)$ ), i.e.  $f_{ij}$  is invariant for rotational transformation but not for the translational ones. In fact, the NCs growing on  $SiO_2$  are subjected to two thermodynamic competitive processes, due to the different surface energies of the Au/Ar interface ( $\gamma_{up}(\theta) = \gamma_{Au/Ar}(\theta)$ ) and of the Au/ $SiO_2$  interface ( $\gamma_{down}(\theta) = \gamma_{Au/SiO_2}(\theta)$ ). This is reflected in a progressive loss of symmetry in the NCs equilibrium shape with increasing the cluster radius beyond a ‘critical radius’ of about 3 nm. In particular, both the angular and radial symmetry of  $\gamma(\theta)$  are lost for the NCs of group 3 (Fig. 3c), where any correspondence between  $\gamma_{up}(\theta)$  and  $\gamma_{down}(\theta)$  is not present (both the rotationally and translationally invariance of the  $f_{ij}$  are lost). Furthermore, the different surface energies at Au/Ar and Au/ $SiO_2$  interfaces determine during the NCs growth an internal strain accumulation, which is released in the formation of internal defects for larger sizes as indicated by Müller and Kern [35].

Obviously, our data are consistent with the expectation that FCC nano-structured Au present to the environment in icosahedral or decahedral structure with the {111} planes being the closest packing planes with the lowest surface free energy. On the other hand, previous calculations [35–37] showed that for noble metals, such as Au, the ratio of

the surface energies  $\gamma(100)/\gamma(111)$  is close to the value of the broken-bound rule considering only the next-neighbor broken bonds. For the (111) Au there are three broken bonds for each surface atom and four for each atom of the (100), so that should be  $\gamma(100)/\gamma(111) = 2/\sqrt{3} \approx 1.155$ . Our data (for example in Fig. 4a) give  $\gamma(100)/\gamma(111) \approx 1.167 \pm 0.018$ , in good agreement with this prediction.

Finally, it is worth to note that, simply by our TEM analyses, we are not able to correlate experimentally the NC’s surface/interface energy with its surface/interface atomic density. However, the link between the surface/interface energy and the surface/interface atomic density of the NCs is possible within the theoretical framework proposed by Qi et al. [38]. Such an approach request the determination of the size-dependent cohesive energy of the NCs. The data concerning this determination and the link between the surface/interface energy and the surface/interface atomic density of the NCs by this theoretical framework will be the topic of a next specific paper.

## Conclusion

We studied the effect of the surrounding environment on the atomic structure and equilibrium shape of growing NCs, considering the particular case of Au NCs in/on amorphous  $SiO_2$ . Starting from HR-TEM analyses, quantitative data on the surface free energy of the NCs were extracted by the inverse Wulff construction. The main result is that growing NCs surrounded by an ‘isotropic’ environment exhibit an angular and radial symmetrical surface free energy function  $\gamma(\theta)$  making rotationally and translationally invariant the surface stress tensor  $f_{ij}$  of the NC: this situation determines a symmetrical equilibrium shape of the NCs. Growing NCs in a ‘non-isotropic’ environment exhibit a  $\gamma(\theta)$  that lost its angular and radial symmetry for sufficiently large sizes determining the loss of the rotationally and translationally invariance of the  $f_{ij}$  and, as a consequence, a loss of symmetry in the equilibrium shape of the NC and the formation of internal defects.

The results suggest that ideal crystalline nanoclusters can be grown on surfaces simply controlling the surrounding environment even by deposition of thin layers of materials as the bulk surface.

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