

Title	Fractal patterning of nanoparticles on polymer films and their SERS capabilities
Authors	Amarandei, George;O'Dwyer, Colm;Arshak, Arousian;Corcoran, David
Publication date	2013-08-27
Original Citation	Amarandei, G., O'Dwyer, C., Arshak, A. and Corcoran, D. (2013) 'Fractal Patterning of Nanoparticles on Polymer Films and Their SERS Capabilities', ACS Applied Materials & Interfaces, 5(17), pp. 8655-8662. doi: 10.1021/am402285e
Type of publication	Article (peer-reviewed)
Link to publisher's version	<a href="https://pubs.acs.org/doi/abs/10.1021/am402285e">https://pubs.acs.org/doi/abs/10.1021/am402285e</a> - <a href="https://pubs.acs.org/doi/abs/10.1021/am402285e">10.1021/am402285e</a>
Rights	© 2013 American Chemical Society. This document is the Accepted Manuscript version of a Published Work that appeared in final form in ACS Applied Materials & Interfaces, copyright © American Chemical Society after peer review and technical editing by the publisher. To access the final edited and published work see <a href="https://pubs.acs.org/doi/pdf/10.1021/am402285e">https://pubs.acs.org/doi/pdf/10.1021/am402285e</a>
Download date	2024-04-27 04:42:29
Item downloaded from	<a href="https://hdl.handle.net/10468/6140">https://hdl.handle.net/10468/6140</a>



# UCC

**University College Cork, Ireland**  
Coláiste na hOllscoile Corcaigh

# Fractal Patterning of Nanoparticles on Polymer Films and their SERS Capabilities

*George Amarandei<sup>\*1</sup>, Colm O'Dwyer<sup>2,3,4</sup>, Aroushian Arshak<sup>1,4</sup> and David Corcoran<sup>1</sup>*

AUTHOR ADDRESS.

<sup>1</sup>Department of Physics and Energy, University of Limerick, Ireland

<sup>2</sup>Department of Chemistry, University College Cork, Ireland

<sup>3</sup>Tyndall National Institute, Lee Maltings, Cork, Ireland

<sup>4</sup>Materials and Surface Science Institute, University of Limerick, Ireland

ABSTRACT. We demonstrate control, via electro-hydrodynamic (EHD) induced polymer instabilities and nanoparticle mobility, of hierarchical fractal arrangements of gold nanoparticles on patterned thin polymer films. The induced changes in the film curvature enhance fractal formation for high and not for low mobility nanoparticles. The high mobility nanoparticles cluster in circular fractal networks on the rims of a hexagonally ordered array of EHD-induced polymer peaks. These arrangements exhibit plasmonic properties for surface enhanced Raman scattering (SERS) spectroscopy.

1  
2  
3  
4  
5  
6  
7  
8  
9  
10  
11  
12  
13  
14  
15  
16  
17  
18  
19  
20  
21  
22  
23  
24  
25  
26  
27  
28  
29  
30  
31  
32  
33  
34  
35  
36  
37  
38  
39  
40  
41  
42  
43  
44  
45  
46  
47  
48  
49  
50  
51  
52  
53  
54  
55  
56  
57  
58  
59  
60

KEYWORDS. nanoparticles, thin polymer films, electro-hydrodynamic instabilities, patterning, SERS

## Introduction

The patterning of nanoparticles (NPs) on surfaces is a prerequisite for nanotechnological applications in photonics, sensing and biomedical science.<sup>1-4</sup> Patterned polymer films containing dielectric NPs have been proven to be viable candidates for a new class of solar cell.<sup>1</sup> Pioneering expertise has been developed in the functional application of nanoparticle polymer film composites<sup>5-9</sup> with potential applications ranging from novel optical and magnetic materials to sensors and even antibacterial coatings.<sup>8</sup> A large variety of reproducible fabrication methods are available for adding nanoparticles to thin-polymer film surfaces e.g. drop casting from solutions,<sup>6</sup> spray coating,<sup>7</sup> vapor,<sup>8</sup> thermal or sputter deposition<sup>5</sup>. In particular, solvent evaporation arising from drop casting or spray coating can be used to pattern in a controllable manner the nanoparticles on a polymer film.<sup>6,7</sup> These low cost fabrication methods use evaporation induced self-assembly for achieving gold nanoparticle ensembles and patterning of the thin film simultaneously.

When arranged periodically and/or regularly, metallic NPs are useful for surface enhanced Raman scattering (SERS) that combines non-invasive and non-destructive molecular fingerprint specificity<sup>10</sup> and single-molecule sensitivity.<sup>11</sup> The coupling of surface plasmon polaritons on proximal Au NP surfaces enhances the electromagnetic (EM) field<sup>12</sup> and creates ‘hot spots’ that allow significantly enhanced scattering cross-sections and very sensitive detection limits.<sup>13-16</sup>

The ultra-sensitivity and potential label-free analysis associated with SERS has made it an attractive and powerful analytical technique to identify and quantify analytes in applications ranging from biosensing and explosive detection to art conservation.<sup>15,16</sup> The main problem in

1  
2  
3 creating “nanoplasmonic devices” is to impregnate/pattern rigid or flexible substrates over large  
4  
5 areas with regular arrays of nanoparticles that can exhibit SERS effects.<sup>4,15,17</sup> The patterning has  
6  
7 to be done in a simple and repeatable fashion to ensure low manufacturing costs for the  
8  
9 development of low cost plasmonic sensors for daily life applications. In the current state-of-the-  
10  
11 art methods patterning is obtained by means of expensive lithographic techniques (nano-transfer  
12  
13 printing, shadow overlap of ion-beam lithography etc.) or by the development of sophisticated  
14  
15 synthesis colloidal techniques.<sup>14–20</sup> Both classes of methods can lead to the desired regular  
16  
17 patterns but they are expensive and have low-throughput. A possible alternative is the use of  
18  
19 soft-lithographic methods that employ thin polymer nanocomposite films to produce hierarchical  
20  
21 patterns.<sup>2–4</sup> By controlling the self-assembly and self-organization properties of matter, regular  
22  
23 patterns can be created with limited input from the outside.  
24  
25  
26  
27  
28

29  
30 At temperatures larger than the glass transition temperature  $T_g$  and in the absence of additional  
31  
32 external fields, the stability of the films is controlled by the van der Waals forces between the  
33  
34 air-polymer and polymer-substrate interfaces.<sup>21–24</sup> For pristine films in a capacitor configuration  
35  
36 (with applied electric fields of  $\sim 10 - 50$  V/ $\mu\text{m}$ ), at  $T > T_g$ , the electric pressure exerted on the  
37  
38 dielectric liquid usually dominates and destabilizes the initially flat air-polymer interface. A  
39  
40 process of self-organization that leads to pattern formation is triggered.<sup>25–28</sup>  
41  
42

43  
44 The high electric fields in a capacitor configuration generate interfacial pressures that are  
45  
46 strong enough to destabilize the film by overcoming the stabilizing action of the surface  
47  
48 tension.<sup>25,28,29</sup> The amplitude of the selected linear mode grows first exponentially, then  
49  
50 nonlinear effects set in and the growth continues until the film maxima touch the other substrate  
51  
52 of the capacitor. In this way an array of liquid columns that connect the two substrates is  
53  
54 formed.<sup>25</sup> The nonlinear growth process and potentially also the electrostatic repulsion between  
55  
56  
57  
58  
59  
60

1  
2  
3 the columns may lead to their arrangement in lattices with different symmetries (e.g. square,  
4 hexagonal). Their geometry, however, does not always remain the same over the entire area, and  
5 a coexistence of different patterns is sometimes observed.<sup>25,29–31</sup> The packing order of the  
6 columns is controlled by the film thickness, and an almost perfect hexagonal distribution is  
7 observed for a specific ratio of the film thickness with respect to the air gap in the capacitor.<sup>25,29–</sup>  
8  
9  
10  
11  
12  
13  
14  
15  
16  
17  
18  
19  
20  
21  
22  
23  
24  
25  
26  
27  
28  
29  
30  
31  
32  
33  
34  
35  
36  
37  
38  
39  
40  
41  
42  
43  
44  
45  
46  
47  
48  
49  
50  
51  
52  
53  
54  
55  
56  
57  
58  
59  
60

<sup>31</sup> The use of electric fields can create hierarchical 3D structures in a single step if multilayers of thin polymer films are employed.<sup>4,29,32–36</sup>

The incorporation of nanoparticles with high dielectric constants as fillers in a thin polymer film leads to a systematic reduction in length scales of the induced pattern.<sup>3</sup> Independently of the type of filler, no migration or altering of the structure generated by the electric field is observed.<sup>2,3</sup>

Here, we show that electro-hydrodynamic (EHD) induced instabilities form an array of polymer patterns with unique Au NP fractal distributions. The polymer structures decorated by NP fractal networks exhibit significant SERS for probe organic molecules. The method has the advantage of patterning the polymer and Au NPs simultaneously. This establishes the possibility of creating and controlling Au NP polymer hierarchical structures for use in nanotechnological applications.

## Experimental

*Sample preparation.* Si wafers (native oxide,  $h_{\text{SiO}_2} \approx 2.0$  nm) with a resistivity of 2 – 3  $\Omega$  cm were cleaned in a jet of CO<sub>2</sub> ice crystals, and were used as substrates. For good electrical contact, 5 nm Cr and 40 nm Au films were deposited on the backside of the wafers. Thin poly (methyl - methacrylate) [PMMA10,  $M_w = 10$  kg/mol,  $R_g = 2.76$  nm,  $M_w/M_n = 1.05$ , Sigma-Aldrich, UK]

1  
2  
3 and polystyrene [PS10,  $M_w = 10$  kg/mol,  $R_g = 2.59$  nm,  $M_w/M_n = 1.05$ , Sigma-Aldrich, UK] films  
4  
5 were obtained by spin coating onto the substrates using 2 – 3% (w/w) polymer solution in  
6  
7 toluene.<sup>29</sup>  
8  
9

10 Gold layers with different nominal thicknesses (as specified for each experiment in the figure  
11 captions) were sputtered at a low rate ( $0.09$  nm s<sup>-1</sup>) onto the polymer thin films creating a  
12 Si/SiO<sub>x</sub>/polymer/AuNP/air configuration.<sup>5,14,23</sup> The deposition led to initial uniformly random  
13 distributed NPs on the polymer surface in the central region of the samples allowing a direct  
14 comparison between Si/SiO<sub>x</sub>/polymer/air and Si/SiO<sub>x</sub>/polymer/AuNP/air regions (see Refs. [23,  
15 24] and schematics drawn in the figures). The films were used as cast and deposited, and no  
16 annealing procedure was performed prior to heating. All experiments were made in a convection  
17 oven at 170 °C. The polymer-air and polymer-air-polymer capacitor configurations (as shown in  
18 the different schematic diagrams) were obtained using SiO<sub>2</sub> microspheres of 1 μm diameter that  
19 act as spacers.<sup>29</sup> A voltage  $U = 50$  V was applied to perturb the air-polymer interfaces. During  
20 the experiments the current through the capacitor was monitored. After an initial transient period  
21 during heating, a small constant current was observed (1 – 60 mA). If significant large changes  
22 in the current or short-circuits occurred, the sample was disregarded.<sup>29</sup> The samples were  
23 removed from the oven and quenched at different time intervals in the presence of the electric  
24 field. At the end of the experiment the samples were mechanically separated. For the polymer-  
25 air-polymer configuration the PS was washed from the PMMA substrate using cyclohexane. The  
26 SiO<sub>2</sub> spheres were removed by using PDMS stamps.  
27  
28  
29  
30  
31  
32  
33  
34  
35  
36  
37  
38  
39  
40  
41  
42  
43  
44  
45  
46  
47  
48  
49

50 *Electron Microscopy.* The polymer films and the Au NP distributions were imaged by electron  
51 microscopy using a Hitachi S4800 FESEM operating at 4 kV and a FEI Orion Focused Ion Beam  
52 at 5 kV. The mean radius of the particles  $R_p$  and the edge-to-edge distance  $d_{ee}$  between the  
53  
54  
55  
56  
57  
58  
59  
60

1  
2  
3 nearest neighbours were measured using SPIP and calculated as described in detail  
4 elsewhere.<sup>23,24</sup> The initial deposition of a nominal 1 nm thickness of Au led to NPs with  $R_p \approx$   
5 1.74 nm and  $d_{ee} \approx 1.76$  nm respectively (see also Refs. [23, 24]). For a nominal 2 nm Au  
6 deposition, presented in Figure 5, the particles were characterized by  $R_p = 3.48$  nm and  $d_{ee} = 1.79$   
7 nm. The fractal dimension  $D$  was calculated using the box counting method available in FIJI (see  
8 Ref. [24] for details).  
9

10  
11  
12  
13  
14  
15  
16  
17 *Surface Enhanced Raman Spectroscopy.* SERS measurements were carried out using a Horiba  
18 Jobin-Yvon T64000 Triple Raman spectrometer attached to an Olympus confocal optical  
19 microscope. Excitation was provided by a He-Ne laser with a wavelength  $\lambda = 633$  nm. The  
20 spectra were typically acquired with a 5 s exposure time and a laser power of 5 mW (at 633 nm)  
21 to avoid organic dye burning. A 50 $\times$  objective with a numerical aperture N.A. of 0.75 was used  
22 for all Raman scattering measurements. This led to a laser spot size  $A_{\text{spot}} = 1.22 \lambda/\text{N.A.} \approx 1.1$   
23  $\mu\text{m}^2$ . Methyl Violet (MV) 10B was used as the probe molecule. The Methyl Violet 10B reagents  
24 were dissolved into pure deionized water (18.3 M $\Omega$  cm) and ethanol. The patterned sample was  
25 immersed for 1 h in the dye solution at a concentration of  $1.0 \times 10^{-6}$  mol dm $^{-3}$  for sufficient  
26 molecule adsorption before SERS measurements, and subsequently rinsed with deionized water.  
27  
28  
29  
30  
31  
32  
33  
34  
35  
36  
37  
38  
39  
40  
41  
42

## 43 Results

44 Thin polymer films at temperatures larger than their glass transition temperature become liquid  
45 and the Au NPs present at the polymer-air interface become free to diffuse and aggregate.<sup>23,24</sup> To  
46 distinguish between the effect of heating with and without the electric field, only a part of the  
47 thin PMMA-NP covered system is placed in the capacitor configuration as indicated in the  
48 schematic diagram in **Figure 1**. This allows a direct comparison of the instabilities developed by  
49  
50  
51  
52  
53  
54  
55  
56  
57  
58  
59  
60

1  
2  
3 the NP coverage in Region I (**Figure 1.a**) and by the electric field in Region II (**Figure 1.b-d**)  
4 and the influence these instabilities have on the Au NP distribution.  
5  
6

7  
8 In Region I the initial stages of the spinodal instability (**Figure 1.a**) are seen. A previously  
9 reported fractal network ( $D = 1.68 \pm 0.09$ ) of Au NPs (seen at  $t > 45$  min) that forms as a result  
10 of a cluster-cluster limited aggregation process<sup>24</sup> is not observed here. Only its incipient stages  
11 i.e. the formation of quasi-linear NP clusters, which are the main constituents of the eventual NP  
12 fractal network, are observed as the experimental time is short ( $t_l \approx 21$  min). The Au clusters  
13 formed during aggregation are uniformly distributed on the peaks and within the troughs of the  
14 polymer instability, showing that the development of the spinodal instability does not influence  
15 NP movement.  
16  
17

18  
19 The transition between Region I and Region II is seen in **Figure 1.b**. In Region II the typical  
20 EHD pattern develops by the end of the experiment as observed by the shadow features in the  
21 SEM observations (**Figure 1.b-d**). The top electrode of the capacitor is placed at a small angle  
22 (exaggerated in the schematic diagram for illustration purposes) and this allows the observation  
23 of the EHD instability development at different times without measurable change to the  
24 instability wavelength.<sup>4,29</sup> In Region IIa where the electric field is larger, the instability develops  
25 faster and it is observed in its final development stages after touching the upper electrode  
26 (**Figure 1.b**, right), at an intermediate developed stage but not touching the electrode (**Figure**  
27 **1.c**, left), and in its initial stages in Region IIb (**Figure 1.c**, right). The black arrows in **Figure 1.c**  
28 reveal the initial stages of the EHD development and associated Au NP distribution. When the  
29 amplitude of the instability is small, the particle distribution on the peaks is almost the same as in  
30 the instability troughs. However, once the instability grows in amplitude the particle density on  
31 the peaks starts to increase and the quasi-linear clusters begin forming fractal networks (white  
32  
33  
34  
35  
36  
37  
38  
39  
40  
41  
42  
43  
44  
45  
46  
47  
48  
49  
50  
51  
52  
53  
54  
55  
56  
57  
58  
59  
60



1  
2  
3 arrows in **Figure 1.c**). Finally, prior to contact, the fractal network is fully formed and covers the  
4 entire peak of the EHD instability as seen to the left of **Figure 1.c** and magnified in **Figure 1.d**  
5 and its inset. The network is formed by Au NPs with a mean radius  $R_p \approx 5.91$  nm and a mean  
6 edge-to-edge distance  $d_{ee} \approx 4.51$  nm and it has a fractal dimension of  $D = 1.69 \pm 0.11$ . Fractal  
7 networks of Au NPs are known to arise when the deposited films are near the percolation  
8 threshold.<sup>9</sup> The values of fractal dimension compare well with those seen after 135 min of  
9 heating in a sample where no electric field was present (see Ref. [24]). Here, however, these  
10 values are obtained after shorter time ( $t_1 \approx 21$  min).

11  
12 Thus the EHD induced polymer instability with a characteristic wavelength can be patterned  
13 with high and low density distributions of Au NPs and it seems to enhance the cluster-cluster  
14 aggregation process that leads to fractal formation.

15  
16 The experiment is repeated using thin polystyrene films covered by Au NPs where the NP  
17 mobility is smaller.<sup>23,24,37,38</sup> In the absence of any other external fields such films also develop a  
18 spinodal instability,<sup>23</sup> but the time to reach full development is much longer than for PMMA. In  
19 PMMA films a spinodal instability can develop in the first  $\sim 45$  min, reaching its maximum  
20 amplitude by  $\sim 315$  min,<sup>24</sup> in PS films a spinodal instability requires over 24 hours to reach its  
21 maximum amplitude.<sup>23</sup> Therefore, the electric field is applied over the entire surface area  
22 covered by Au NPs, as schematically described in **Figure 2**. As in the PMMA study (**Figure 1**),  
23 a small angle (exaggerated in the schematics) exists between the capacitor plates allowing  
24 observation of different evolution stages of the EHD instability.

25  
26 In Region I, where no Au NPs or electric field are present (**Figure 2.a**), the thin polymer film  
27 exhibits nucleation dewetting and beads up in a Voronoi pattern as expected.<sup>23</sup> In Region IIa the  
28 instability is fully formed and its peaks already contact the upper electrode at a few locations (the  
29  
30  
31  
32  
33  
34  
35  
36  
37  
38  
39  
40  
41  
42  
43  
44  
45  
46  
47  
48  
49  
50  
51  
52  
53  
54  
55  
56  
57  
58  
59  
60

1  
2  
3 black points indicated by white arrows in **Figure 2.b**). In Region IIb, the early stage of the  
4  
5 instability is seen (**Figure 2.c**). Independently of the development stage of the EHD instability or  
6  
7 their position (i.e. on the peaks or within the troughs) the Au NPs maintain their initial uniform  
8  
9 distribution as seen in **Figure 2.d**. In this regard, as noted earlier Au NP aggregation occurs at a  
10  
11 slower rate on PS than on PMMA, taking place either through Ostwald ripening or a slow  
12  
13 coalescence process.<sup>23,39</sup>  
14  
15

16  
17 In a double layer PS/air/AuNP/PMMA configuration (**Figure 3**), a single mode instability is  
18  
19 predicted<sup>29</sup> and indeed is experimentally seen to develop in both layers. Due to the higher bulk  
20  
21 viscosity of the PMMA film combined with the presence of Au NPs on its surface, its mobility is  
22  
23 significantly lower than the mobility of the PS film, and consequently the instability grows faster  
24  
25 in the more mobile PS film.<sup>29</sup> Indeed, the topography of the PS-PMMA double layer is consistent  
26  
27 with the type II topography previously seen for double layer systems where the mobility of one  
28  
29 film far exceeds that of the other. The more mobile film exhibits a labyrinthine pattern while the  
30  
31 less mobile film exhibits features that appear to be extruded from a flat background.<sup>29</sup>  
32  
33

34  
35 As PS and PMMA are used, columns with a PS core and PMMA shell are formed<sup>29,32,33</sup> when  
36  
37 the corrugated peaks of the two polymer films meet (**Figure 3.a**). In this case the densification  
38  
39 of the Au NPs occurs on and/or around the rims created by the PMMA shells (**Figure 3.b,c**).  
40  
41 Where PS peaks contact the AuNP/PMMA layer, the Au NP aggregation appears to freeze-in  
42  
43 (**Figure 3.b,c**) as the Au NPs adsorb to PS.<sup>23,40</sup> As has been shown above, the NP distribution on  
44  
45 PS is effectively retarded from significant changes. The result is the generation of a rich  
46  
47 hierarchical Au NP – polymer structure (**Figure 3.b**) consisting of a polymer film plane of  
48  
49 uniform Au NP cluster distribution, on which are superimposed polymer rims with high density  
50  
51 Au NPs and inner troughs of Au NP clusters (**Figure 3.a**).  
52  
53  
54  
55  
56  
57  
58  
59  
60

1  
2  
3 Different stages of evolution of the Au NPs on polymer patterns are observed due to the slight  
4 tilt in the electrode spacing mentioned earlier (see also Ref. [29]). In the more developed  
5 situation presented in **Figure 3**, the fractal network is located on the outside base of the rim. At  
6 an earlier stage of development (**Figure 4.a**), the Au NP fractal network ( $D = 1.73 \pm 0.09$ ) is  
7 distributed along the entire contour and on the top of the rim (**Figure 4.b**). The particles have a  
8 mean  $R_p \approx 6.86$  nm and  $d_{ee} \approx 4.8$  nm. The particles outside of the rim (see also **Figure S.1** in the  
9 Supporting Information.) form clusters that are not connected to the NP network. The particles  
10 within such clusters are characterized by an  $R_p \approx 7.26$  nm and  $d_{ee} \approx 5.2$  nm. As seen in **Figure 1**,  
11 the cluster-cluster aggregation process is enhanced by the presence of a larger curvature (in this  
12 case on the rim).  
13  
14  
15  
16  
17  
18  
19  
20  
21  
22  
23  
24  
25  
26

27 A similar ridge with Au NP densification and an inner trough with uniformly distributed NPs  
28 can also be produced by placing a SiO<sub>2</sub> sphere on a heated AuNP/PMMA configuration, as  
29 depicted in **Figure 5**. In this instance, the Au NPs outside the rim form the typical fractal  
30 network of Au NP clusters previously observed.<sup>24</sup> The fractal is also present on the rim, but  
31 underneath the spheres, *i.e.* where the mobility of the NPs is severely impaired, the distribution  
32 remains uniformly random. These results together with the observations from polymer-air and  
33 the polymer-air-polymer configurations above suggest that the local increase in the curvature of  
34 the polymer extrusions enhances the Au NP organization process to form fractal NP  
35 arrangements, while the factors that reduce Au NP mobility act to retard this process.  
36  
37  
38  
39  
40  
41  
42  
43  
44  
45  
46  
47

48 Site specific SERS measurements are carried out on the early stage PMMA structures from the  
49 polymer-air-polymer experiment shown in **Figures 4.a** and **4.b**. The small quasi-linear Au NP  
50 aggregates located inside of the rim (position 2), and the large fractal network on the rim  
51 (position 3) are potentially suitable arrangements for SERS<sup>13,18,41-43</sup> of organic moieties such as  
52  
53  
54  
55  
56  
57  
58  
59  
60

MV 10B used here. In principle, an organic molecule with a molecular footprint that covers proximal NPs can have their vibrational characteristics enhanced. The higher particle density in these arrangements allows for the presence of the so called hot spots that are defined by electric field enhancement between closely spaced Au NPs. The hot spots are necessary for SERS given that Raman scattering cross sections for MV 10B are extremely small ( $\sim 3.6 \times 10^{-26} \text{ cm}^2 \text{ sr}^{-1}$ ).<sup>13,44</sup> The coverage, arrangement of the Au NP structures, the planar surface distribution of Au NPs (**Figure 4.b**) and the curvature of the polymer are unaffected by the liquid-based dye during functionalization as shown in **Figure S.1**. The quasi-linear Au NP clusters and the NP networks remain spatially pinned in their original positions.

Raman scattering measurements of the MV 10B are acquired from the fractal NP network covering the rim (position 3) and the central region inside the structure (position 2) and compared to those from the planar area between the polymer peaks (position 1) in **Figure 4.c**. The spectra reveal that the fractal networks exhibit significant enhancement of the scattering cross-section as the specific Raman-shifts undergo intensity increases and markedly improved resolution (**Figure 4.c**, spectrum 3). The NP arrangement in the central region also exhibits an enhancement (**Figure 4.c**, spectrum 2) but the effect is generally weaker. The primary signatures of the MV i.e. the main bands enhanced, are at  $\sim 1175$  (in plane aromatic C–H bending),  $\sim 1373$  (phenyl C–N stretching and in-plane C–H bending),  $\sim 1584$  and  $1617 \text{ cm}^{-1}$  (out-of-phase in plane stretching of the carbon ring atoms).<sup>45</sup> The ratios of the integrated intensities (i.e. the area under each spectrum) are  $I_2/I_1 \approx 3$  and  $I_3/I_1 \approx 10^2$ .

The enhancement factor,  $EF$ , for each region can be calculated<sup>44,46</sup> using:

$$EF = \frac{I_{SERS}}{I_{RS}^{single}} \cdot \frac{1}{N_{CV} \cdot t} \quad (1)$$

For MV dye the contribution of a single dye molecule without surface enhancement to the Raman signal rate is  $I_{RS}^{single} = 6.3 \times 10^{-7} s^{-1}$ .<sup>46</sup> In the present study a monolayer of dye molecules is assumed to be adsorbed onto the substrate (see Supporting Information). Considering that a MV dye molecule occupies an area of 3.51 nm<sup>2</sup> per molecule<sup>46</sup> implies  $\sim 3.12 \times 10^5$  molecules are adsorbed in the laser spot area of  $\sim 1.1 \mu\text{m}^2$ . Using the  $I_{SERS}$  for the 1617 cm<sup>-1</sup> peak and considering a uniform monomolecular coverage of the dye the enhancement factors for the regions are found to be  $EF_1 \approx 1$ ,  $EF_2 \approx 3$  and  $EF_3 \approx 2 \times 10^2$ . The enhancement factor calculated for  $EF_3$  is an underestimate as  $A_{spot} = 1.1 \mu\text{m}^2$  is larger than the rim area  $A_{rim} \approx 0.36 \mu\text{m}^2$  and so contains regions of lower NP density and SERS enhancement (see **Figure S.1**). Accounting for this and considering uniform coverage,  $EF_3$  can be written as the linear contribution of the two regions:<sup>46</sup>

$$EF_3 = \frac{A_{rim}}{A_{spot}} \times EF_{rim} + \frac{A_{spot} - A_{rim}}{A_{spot}} \times EF_1 \quad (2)$$

Thus, the enhancement factor from the rim alone is estimated to be  $EF_{rim} \approx 6 \times 10^2$ . As the largest enhancement occurs from the fractal network of NPs on the rim (position 3), this confirms that the rim has the highest density of NPs with proximities that allow electromagnetic field enhancement causing plasmonic coupling and therefore SERS. Conversely those regions with lower enhancement (regions 2 and 1) correspond, in order of decreasing value, to lower NP density.

## Discussion

In this study the Au NPs are sputtered onto the surface of the polymer film. Once the polymer films become liquid the particles are free to diffuse and aggregate. The mobility of the NPs on the polymer film influences their aggregation process. Thus, for lower mobility the NP diffusion

1  
2  
3 and aggregation occurs by Ostwald ripening and/or slow coalescence.<sup>23,39</sup> These processes do not  
4  
5 seem to be affected by the growth of an electric field induced instability in the polymer film (see  
6  
7 **Figure 2**).

8  
9  
10 When the NPs are highly mobile, the aggregation process proceeds by initially forming small  
11  
12 linear clusters that continue to aggregate and eventually form a fractal network through a  
13  
14 diffusion limited process (see Ref. [24] and the references therein). The presence of an electric  
15  
16 field that induces faster changes in polymer surface corrugation seems to enhance this process.  
17  
18 This leads to changes in particle density and spatial distribution resulting in fractal clusters  
19  
20 (network) on the peaks of the corrugation. The localised fractal formation seems to be associated  
21  
22 with an increase in the localised polymer curvature. Thus, in **Figure 1** and **3** we note that the  
23  
24 fractal is formed either on the peak of the EHD instability (clusters) or on the rims (circular  
25  
26 network), while in the flat regions between structures the Au NPs remain organized in quasi-  
27  
28 linear clusters. When compared with previous studies presented in Ref. [24] the values  
29  
30 characterizing the particles arrangements (i.e.  $R_p$ ,  $d_{ee}$  and  $D$ ) seen on the curved features after  $t \approx$   
31  
32 21 or 28 min are similar to those obtained for larger heating times (i.e.  $t \geq 135$  min).  
33  
34  
35  
36  
37  
38

39 The SERS measurements directly confirm the visual changes to the NP density observed in the  
40  
41 polymer-air-polymer experiment. The scattering data are consistent with the highest particle  
42  
43 density being on the rims, and a greater particle density being inside these structures (position 2)  
44  
45 compared to the flat regions (position 1).  
46  
47

48 In the polymer-air-polymer system, the EHD starts to develop in both films at the same time.  
49  
50 As the PS has a smaller viscosity than PMMA, the instability growth is accelerated for PS.<sup>29</sup>  
51  
52 Therefore, the instability in PMMA is less developed and its amplitude and curvature is small.  
53  
54 The contact with the mirrored features from PS freeze-in the linear Au NPs clusters at an early  
55  
56  
57  
58  
59  
60

1  
2  
3 stage i.e. well before the fractal is formed (compare **Figure 3.c** with **Figure 1.d**). Once the  
4  
5 contact between the mirrored features is formed the PMMA tends to form the shell of the  
6  
7 columnar structures (see schematic in **Figure 3**). The aggregation process continues on the rims  
8  
9 where the Au NP movement is not restricted (compare **Figure 4.b** with **Figure 3**). The presence  
10  
11 of SiO<sub>2</sub> sphere also impairs the movement of the NPs from the outset in **Figure 5**, the NP radius  
12  
13 as deposited ( $R_p = 3.48$  nm) being above the critical radius (3.3 nm) that allows for Au clusters to  
14  
15 undergo fusion processes on Si substrates.<sup>9</sup> Therefore, NPs maintain their initial randomly  
16  
17 uniform distribution (**Figure 5.a**). Essentially, the contact with PS instabilities or SiO<sub>2</sub>  
18  
19 microspheres can retard Au NP aggregation on PMMA, and thus control it.  
20  
21  
22  
23

24  
25 The NP mobility is affected mainly by viscosity of the polymer and the adhesion work  
26  
27 between the Au NP and the film, while the NP cluster mobility is affected by possible bridges  
28  
29 created by the polymer chains between the NPs.<sup>24,47</sup> In the present case, the enhancement in NP  
30  
31 and/or cluster mobility in PMMA can be attributed only to a decrease in viscosity as the other  
32  
33 two factors that act to reduce the mobility have similar values across the samples. Moreover, the  
34  
35 reduced viscosity appears to depend on changes in the local curvature of the polymer film as the  
36  
37 aggregation is enhanced in these regions.  
38  
39

40  
41 Such changes are reported in the literature<sup>27,29,48,49</sup> but are related to changes in the mobility of  
42  
43 pristine polymer films. For imprinted features with large aspect ratio and large curvature<sup>48</sup> such  
44  
45 enhancement in polymer mobility is seen only when the feature height is above a certain  
46  
47 threshold. A similar threshold is required for the enhancement in the growth rates of the  
48  
49 instabilities formed in a polymer-air-polymer system.<sup>29</sup> In the present case, however, the  
50  
51 enhancement of the NP aggregation is seen from the initial stages of the EHD instability where  
52  
53 extreme curvatures are not present and the system can be theoretically treated using the long  
54  
55  
56  
57  
58  
59  
60

1  
2  
3 wave approximation.<sup>29</sup> The presence of a surface layer with a lower viscosity is also reported for  
4  
5 high molecular weight polymers, where entanglement of the polymer chains occurs<sup>27</sup> or when the  
6  
7 film thickness is reduced.<sup>29,49</sup> If such an effect is present, it should affect the NP mobility  
8  
9 everywhere and would not be limited to the highly curved corrugations of the polymer film as  
10  
11 observed here.  
12  
13

14  
15 The reduction in the film viscosity is also reported to be caused by chain alignment under a  
16  
17 thermal field.<sup>50</sup> Such alignments increase with decreasing molecular weight and decreasing film  
18  
19 thickness and might be able to induce changes in film mobility in very thin films.<sup>29,51</sup> Here,  
20  
21 however, the system is isothermal. Therefore, if present, any alignment that could cause a  
22  
23 reduction in viscosity must increase with the curvature. This viscosity decrease could explain the  
24  
25 observed enhancement of the NP mobility as the structures are formed.  
26  
27

28  
29 Finally, it should be mentioned that the method reported here has the significant advantage of  
30  
31 patterning the polymer and Au NPs simultaneously. Classical approaches in thin polymer film  
32  
33 patterning include the use of top-down photolithography,<sup>52</sup> nanoimprinting<sup>53,54</sup> or e-beam  
34  
35 writing,<sup>4</sup> but these techniques are limited by their expensive multi-step processes, the  
36  
37 development of lithographic, or imprinting masks or by the beam drift and/or its stability.<sup>4</sup>  
38  
39 Alternative approaches to creating the required patterns are to incorporate NPs as fillers in thin  
40  
41 polymer films, and use either spinodal demixing,<sup>1</sup> induce spinodal<sup>23,24</sup> or EHD<sup>2,3</sup> instabilities to  
42  
43 corrugate the air interface of the thin nanocomposite film.  
44  
45  
46  
47

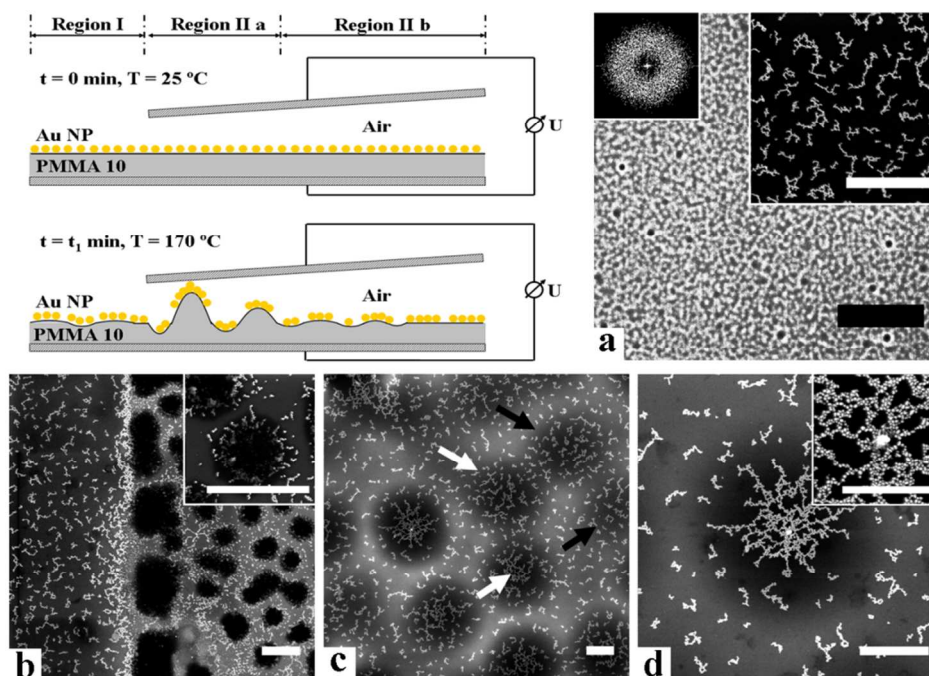
48  
49 SERS is observed here in the fractal networks of Au NPs formed on the rims of PMMA in a  
50  
51 polymer-air-polymer electric field configuration. The enhancement factors of the test molecule  
52  
53 spectral components are comparable to those recently reported in the literature for a silicon  
54  
55 substrate with evaporated carbon and sputtered gold layers<sup>55</sup> or for clusters in solutions with low  
56  
57  
58  
59  
60



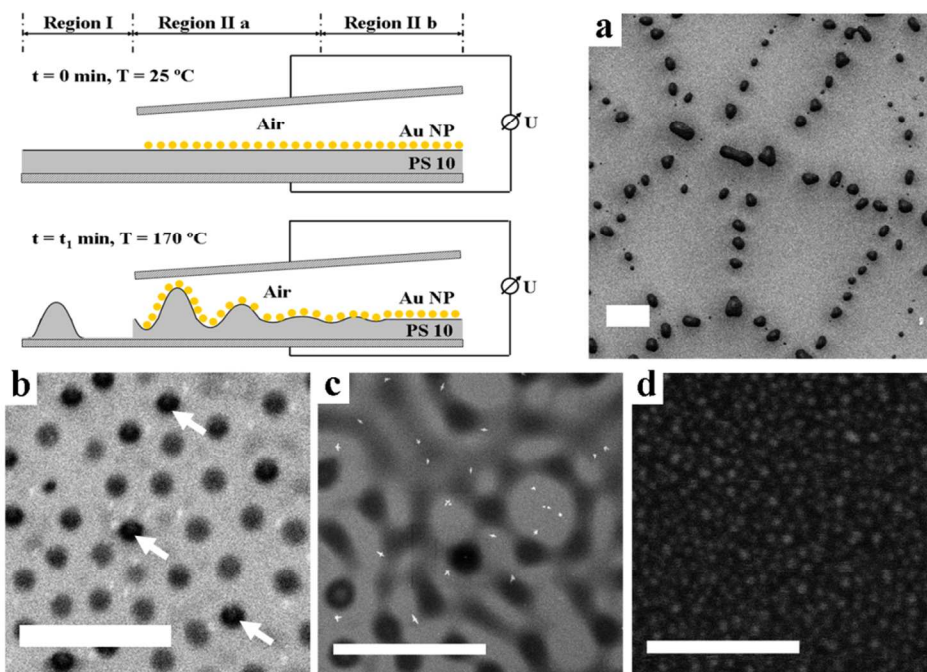
1  
2  
3 dye concentration.<sup>19</sup> Thus, when low dye concentration ( $1 \times 10^{-6}$  M) are used in solutions with  
4  
5 randomly distributed clusters formed by particles with small radii (7.5 – 15 nm) enhancement  
6  
7 factors of  $EF \sim 10^1 - 10^2$  are reported.<sup>19</sup> Here, however, the use of an electric field induced  
8  
9 instability allows for patterning of the thin polymer film <sup>25,32</sup> and, therefore, for spatial  
10  
11 positioning of the regions with higher Au NP density. Thus, polymer patterns and the high NP  
12  
13 density regions can be controlled and anticipated from the outset. This demonstrates that the  
14  
15 simple patterning of the polymer by an electric field induced instability associated with  
16  
17 densification of NPs around these features can be used as a soft-lithographic technique to create  
18  
19 patterned substrates suitable for SERS measurements.  
20  
21  
22  
23  
24  
25  
26

## 27 **Conclusions**

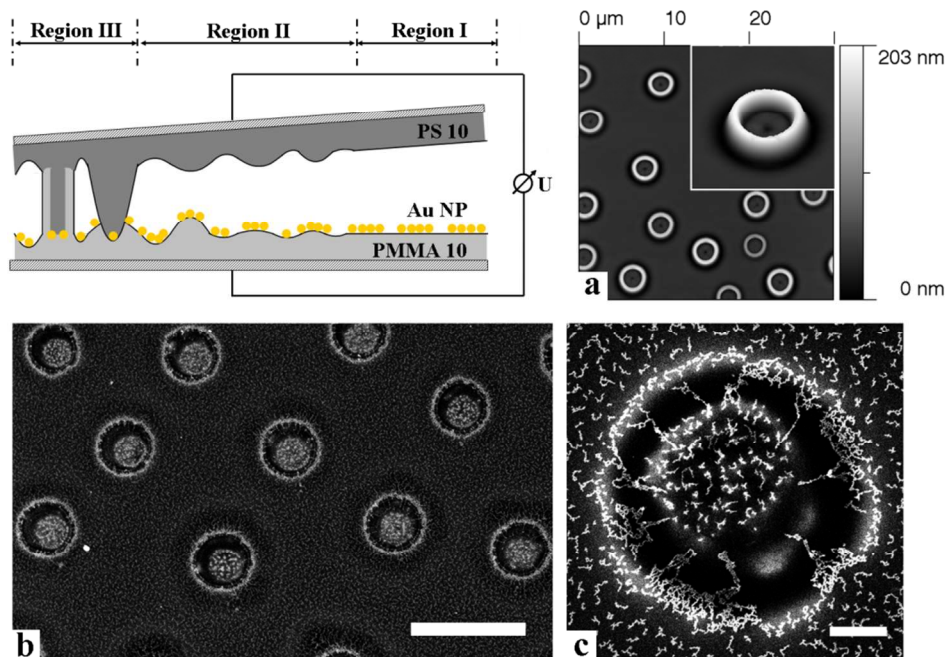
28 The present study shows that by inducing an electric field instability in a thin polymer film  
29  
30 covered by Au nanoparticles it is possible to form a panoply of patterns at different length scales  
31  
32 and spatial positions. It is observed that the rapid growth of a single instability mode from the  
33  
34 capillary spectrum due to the applied electric field can have a significant influence on the  
35  
36 nanoparticle aggregation process only if the NPs have large mobility on the polymer surface.  
37  
38 When they have low mobility, the NPs maintain their initial distribution and the electric field  
39  
40 leads to an instability in the polymer film. Higher mobility nanoparticles form fractal clusters or  
41  
42 networks spatially organized on the regular pattern in the polymer film induced by the electric  
43  
44 field instabilities. The potential for such systems in surface enhanced Raman spectroscopy is  
45  
46 demonstrated by exploiting the beneficial and controllable arrangement of the Au nanoparticles  
47  
48 at the same time as the polymer patterning.  
49  
50  
51  
52  
53  
54  
55  
56  
57  
58  
59  
60



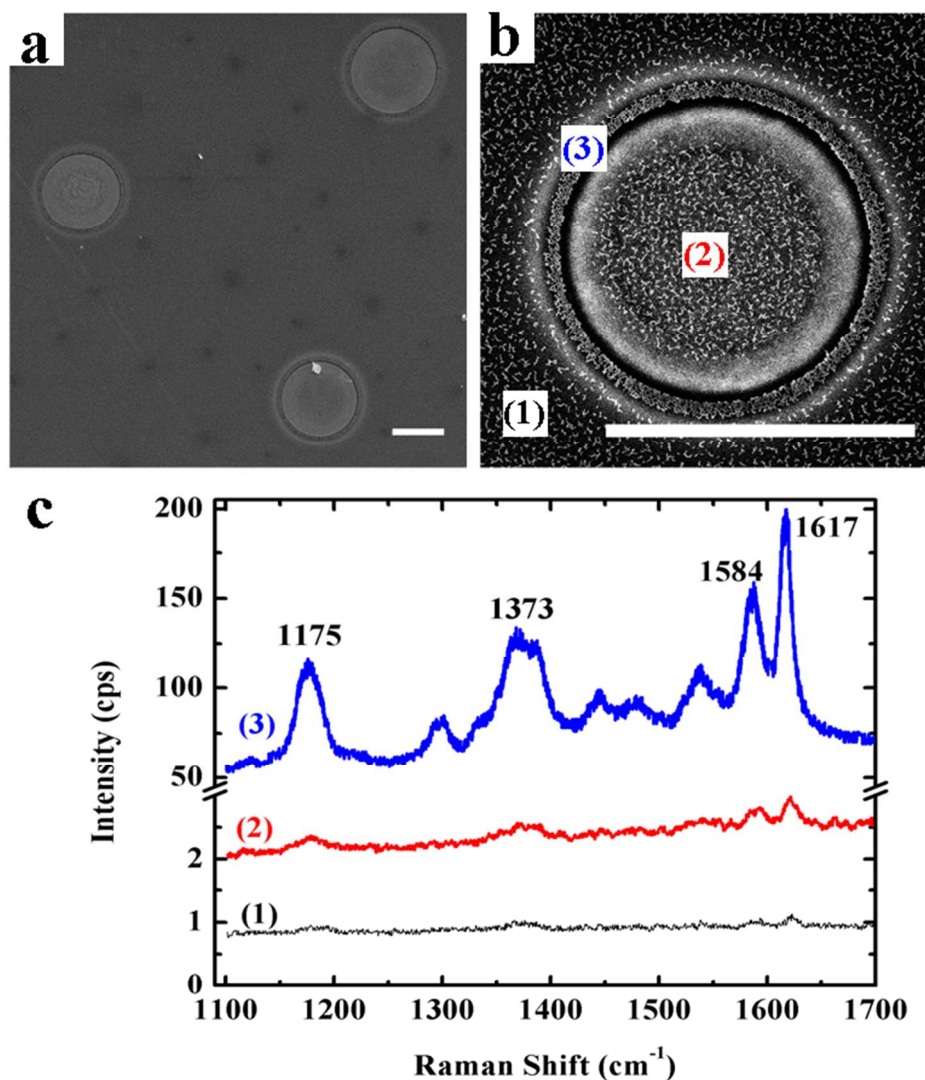
**Figure 1.** Schematic diagram showing the development of instabilities in a region without and with the electric field. (a) Optical micrograph (scale bar (SB) = 50  $\mu\text{m}$ ) of the spinodal instability (as shown by the FFT) developed due to Au coverage in the region without electric field on a film of  $h_{PMMA10} = 29$  nm,  $h_{Au} = 1$  nm after  $t_1 = 21$  min. The SEM inset (SB = 1  $\mu\text{m}$ ) shows the Au distribution. (b) SEM image showing the transition from Region I to Region IIa (SB = 1  $\mu\text{m}$ ). The SEM inset (SB = 1  $\mu\text{m}$ ) describes the Au distribution in and around a peak of the EHD instability that was already in contact with the top electrode. (c) SEM image (SB = 1  $\mu\text{m}$ ) showing the different stages in EHD development at the transition between Region IIa and Region IIb. The black arrows indicate EHD peaks in the early stages (small amplitude) where the Au particles distribution is closed to that for the flat film. The white arrows indicate the beginning of the fractal cluster formation. (d) SEM image (SB = 1  $\mu\text{m}$ ) of a fully formed EHD instability peak in Region II showing the changes in particle density. The inset (SB = 500 nm) shows that as the density become larger on the peak and the particles already form a fractal cluster.



**Figure 2.** Schematic diagram showing the development of the instability in a region without and with the electric field applied. SEM images describing (a) the nucleation dewetting in the region without Au and electric field ( $SB = 50 \mu\text{m}$ ); (b) the EHD instability in Region IIa ( $SB = 50 \mu\text{m}$ ). The white arrows indicate some of the contact points with the upper electrode; (c) the early stages of the EHD instability as seen in Region IIb ( $SB = 50 \mu\text{m}$ ) and (d) the uniform distribution that is maintained as deposited in Region II ( $SB = 50 \text{nm}$ ). The polymer film thickness  $h_{PS10} = 25 \text{nm}$  and the nominal thickness of the Au layer  $h_{Au} = 1 \text{nm}$ . The experimental time is  $t_l = 90 \text{min}$ .

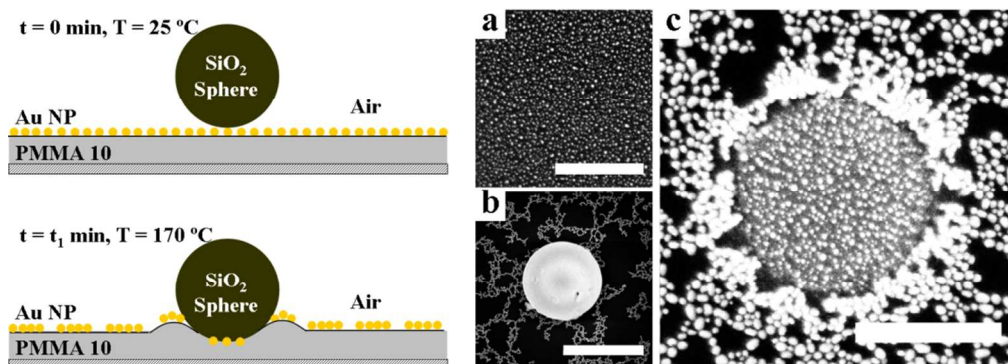


**Figure 3.** Schematic diagram showing the polymer surface and NP distribution in a polymer/AuNP – air – polymer system. (a) The PMMA surface after PS washing from a PS100-air-PMMA100 as described in Ref. [16]; (b) SEM image (SB = 10  $\mu\text{m}$ ) showing the PMMA10 surface and the Au distribution from a PMMA10/AuNP – air – PS10 system. The flat film between the rings indicates the presence of type II behaviour.<sup>16</sup> (c) SEM image (SB = 1  $\mu\text{m}$ ) showing a magnified view of the rings and emphasizing the presence of the fractal cluster on the rim. The linear clusters have the same distribution inside and outside of the rims. The experimental heating time is  $t_l = 28$  min.



**Figure 4.** (a) Secondary electron FIB images showing early contact structures in PMMA (SB = 10  $\mu\text{m}$ ) after the MV solution is applied; (b) SEM magnification of a PMMA feature (SB = 10  $\mu\text{m}$ ) prior to MV adsorption; (c) SERS spectra obtained at the positions identified in (b).





**Figure 5.** Schematic diagram showing the Au distribution under a spacer before and after heating. (a) SEM image (SB = 200 nm) revealing the Au NP distribution prior to heating. SEM images describing the Au distribution after heating before (b) and after (c) spacer removal. (Scale bars represent 1 μm and 200 nm, respectively.) It can be seen that under the spacer (c) the Au NPs retain their initial uniform distribution as in (a). Film polymer thickness is  $h_{PMMA10} = 26$  nm, nominal thickness of the deposit Au layer was  $h_{Au} = 2$  nm, heating time  $t_1 = 17$  min.

## ASSOCIATED CONTENT

**Supporting Information.** The ancillary material contains detailed images showing the nanoparticles clusters and fractal network as seen by the laser spot during the Raman investigation. This information is available free of charge via the Internet at <http://pubs.acs.org/>.

## AUTHOR INFORMATION

### Corresponding Author

\* Dr. George Amarandei, Department of Physics and Energy, University of Limerick, Ireland.

Email: [george.amarandei@ul.ie](mailto:george.amarandei@ul.ie).

### Author Contributions

The manuscript was written through contributions of all authors. All authors have given approval to the final version of the manuscript.

## ACKNOWLEDGMENT

We acknowledge support under EU Framework 7 for projects MRTN-CT-2004005728 (PATTERNS), PERG04-GA-2008-239426 (POLYPATT), and from Tyndall National Institute through Science Foundation Ireland (SFI) funded National Access Programme (Project NAP 200). This work was also supported by SFI under contract no 07/SK/B1232a and by the UCC Strategic Research Fund.

## ABBREVIATIONS

PS, polystyrene; PMMA, poly(methyl-methacrylate); NP, nanoparticle; SERS, surface enhanced Raman scattering.

## REFERENCES

- (1) Wong, H. C.; Cabral, J. T. *Phys. Rev. Lett.*, **2010**, *105*, 038301.
- (2) Bae, J. J. *Ind. Eng. Chem.*, **2012**, *18*, 378–383.
- (3) Bae, J. W.; Glogowski, E.; Gupta, S.; Chen, W.; Emrick T.; Russell, T. P. *Macromolecules*, **2008**, *41*, 2722–2726.
- (4) Goldberg-Oppenheimer, P.; Mahajan S.; Steiner, U. *Adv. Mater.* **2012**, *24*, OP175–OP180.
- (5) Kaune, G.; Ruderer, M. A.; Metwalli, E.; Wang, W.; Couet, S.; Schlage, K.; Röhlberger, R.; Roth S. V.; Müller-Buschbaum, P. *ACS. Appl. Mater. Interfaces.* **2009**, *1*, 353–360.
- (6) Roth, S. V.; Herzog, G.; Körstgens, V.; Buffet, A.; Schwartzkopf, M.; Perlich, J.; Abul Kashem, M. M.; Döhrmann, R.; Gehrke, R.; Rothkirch, A.; Stassig, A.; Wurth, W.; Benecke, G.; Li, C.; Fratzl, P.; Rawolle, M.; Müller-Buschbaum, P. *J. Phys.: Condens. Matter* **2011**, *23*, 254208.

- 1  
2  
3 (7) Al-Hussein, M.; Schindler, M.; Ruderer, M. A.; Perlich, J.; Schwartzkopf, M.; Herzog, G.;  
4  
5 Heidmann, B.; Buffet, A.; Roth, S. V.; Müller-Buschbaum, P. *Langmuir* **2013**, *29*, 2490–2497.  
6  
7  
8  
9 (8) Faupel, F.; Zaporojtchenko, V.; Strunskus, T.; Elbahri, M. *Adv. Eng. Mater.* **2010**, *12*, 1177–  
10  
11 1190.  
12  
13  
14 (9) Schwartzkopf, M.; Buffet, A.; Körstgens, V.; Metwalli, E.; Schlage, K.; Benecke, G.; Perlich,  
15  
16 J.; Rawolle, M.; Rothkirch, A.; Heidmann, B.; Herzog, G.; Müller-Buschbaum, P.; Röhlberger,  
17  
18 R.; Gehrke, R.; Stribeck, N.; Roth, S. V. *Nanoscale*, **2013**, *5*, 5053–5062.  
19  
20  
21  
22 (10) Cao, Y. C.; Jin, R.; Mirkin, C. A. *Science*. **2002**, *297*, 1536–1540.  
23  
24  
25  
26 (11) Cialla, D.; A. März, A.; Böhme, R.; Theil, F.; Weber, K.; Schmitt, M.; Popp, J. *Anal.*  
27  
28 *Bioanal. Chem.* **2012**, *403*, 27–54.  
29  
30  
31 (12) Kneipp, K.; Wang, Y.; Kneipp, H.; Perelman, L. T.; Itzkan, I.; Dasari, R.; Feld, M. S. *Phys.*  
32  
33 *Rev. Lett.*, **1997**, *78*, 1667–1670.  
34  
35  
36  
37 (13) Kneipp, K.; Kneipp, H.; Itzkan, I.; Dasari, R.; Feld, M. *J. Phys. Condens. Matter.* **2002**, *14*,  
38  
39 R597–R624.  
40  
41  
42 (14) Kleinman, S. L.; Frontiera, R. R.; Henry, A. I.; Dieringer J. A.; Van Duyne, R. P. *Phys.*  
43  
44 *Chem. Chem. Phys.*, **2013**, *15*, 21–36.  
45  
46  
47  
48 (15) Polavarapu, L.; Liz-Marzan, L. M. *Phys. Chem. Chem. Phys.*, **2013**, *15*, 5288–5300.  
49  
50  
51  
52 (16) Greeneltch, N. G.; Blaber, M. G.; Henry, A. I.; Schatz, G. C.; Van Duyne, R. P. *Anal.*  
53  
54 *Chem.*, **2013**, *85*, 2297–2303.  
55  
56  
57 (17) Ko, H.; Singamaneni, S.; Tsukruk, V. V. *Small*, **2008**, *4*, 1576–1599.  
58  
59  
60



- 1  
2  
3 (18) Kuo, T. C.; Hsu, T. C.; Y. C. Liu, Y. C.; Yang, K. H. *Analyst.*, **2012**, *137*, 3847–3853.  
4  
5  
6 (19) Joseph, V.; Matschulat, A.; Polte, J.; Rolf, S.; Emmerlinga F.; Kneipp, J. *J. Raman*  
7  
8  
9 *Spectrosc.*, **2011**, *42*, 1736–1742.  
10  
11  
12 (20) Adams, S. M.; Campione, S.; Caldwell, J. D.; Bezares, F. J.; Culbertson, J. C.; Capolino, F.;  
13  
14 Ragan, R. *Small*, **2012**, *8*, 2239–2249.  
15  
16  
17 (21) Seemann, R.; Herminghaus, S.; Jacobs, K. *Phys. Rev. Lett.*, **2001**, *86*, 5534–5537.  
18  
19  
20  
21 (22) Thiele, U. in *Thin films of Soft Matter*, (ed. S. Kalliadasis and U. Thiele), Springer, Wien,  
22  
23 **2007**, Ch. Structure formation in thin liquid films, 25–94.  
24  
25  
26 (23) Amarandei, G.; O’Dwyer, C.; Arshak A.; Corcoran, D. *Soft Matter.*, **2013**, *9*, 2695–2702.  
27  
28  
29 (24) Amarandei, G.; O’Dwyer, C.; Arshak, A.; Thiele, U.; Steiner, U.; Corcoran, D. *Langmuir.*,  
30  
31 **2013**, *29*, 6706–6714.  
32  
33  
34 (25) Schaffer, E.; Thurn-Albrecht, T.; Russell, T. P.; Steiner, U. *Nature.*, **2000**, *403*, 874–877.  
35  
36  
37  
38 (26) Wu, N.; Russel, W. B. *Nano Today.*, **2009**, *4*, 180–192.  
39  
40  
41 (27) Barbero, D. R.; Steiner, U. *Phys. Rev. Lett.*, **2009**, *102*, 248303.  
42  
43  
44 (28) Goldberg-Oppeneheimer P.; Steiner U. *Small* **2010**, *6*, 1248–1254  
45  
46  
47 (29) Amarandei, G.; Beltrame, P.; Clancy, I.; O’Dwyer, C.; Arshak, A.; Steiner, U.; Corcoran,  
48  
49 D.; Thiele, U. *Soft Matter.*, **2012**, *8*, 6333–6349.  
50  
51  
52 (30) Merkt, D.; Pototsky, A.; Besthorn, M.; Thiele, U. *Phys. Fluids.*, **2005**, *17*, 064104.  
53  
54  
55 (31) Voicu, N. E.; Harkema S.; Steiner, U. *Adv. Funct. Mater.*, **2006**, *16*, 926–934.  
56  
57  
58  
59  
60

- 1  
2  
3 (32) Morariu, M.; Voicu, N. E.; Schaffer, E.; Lin, Z.; Russell, T. P.; Steiner, U. *Nat. Mater.*,  
4  
5 **2003**, *2*, 48–52.  
6  
7  
8  
9 (33) Leach, K.; Gupta, S.; Dickey, M.; Willson, C.; Russell, T. P. *Chaos.*, **2005**, *15*, 047506.  
10  
11  
12 (34) Pototsky, A.; Bestehorn, M.; Merkt, D.; Thiele, U. *Phys. Rev. E.*, **2004**, *70*, 025201(R).  
13  
14  
15 (35) Pototsky, A.; Bestehorn, M.; Merkt, D.; Thiele, U. *J. Chem. Phys.*, **2005**, *122*, 224711.  
16  
17  
18 (36) Pototsky, A.; Bestehorn, M.; Merkt, D.; Thiele, U. *Europhys. Lett.*, **2006**, *74*, 665–671.  
19  
20  
21 (37) Lopes, W. A.; Jaeger, H. M. *Nature.*, **2001**, *414*, 735–738.  
22  
23  
24  
25 (38) Lopes, W. A. *Phys. Rev. E.*, **2002**, *65*, 031606.  
26  
27  
28 (39) Jia, X.; Listak, J.; Witherspoon, V.; Kalu, E. E.; Yang X.; Bockstaller, M. R. *Langmuir.*,  
29  
30 **2010**, *26*, 12190–12197.  
31  
32  
33 (40) Fu, T. Z.; Stimming, U.; Durning, C. J. *Macromolecules.*, **1993**, *26*, 3271–3281.  
34  
35  
36  
37 (41) Fraire, J. C.; Perez, L. A.; Coronado, E. A. *ACS Nano.*, **2012**, *6*, 3441–3452.  
38  
39  
40 (42) Ganbold, E. O.; Park, J. H.; Dembereldorj, U.; Ock, K. S.; Joo, S. W. *J. Raman Spectrosc.*,  
41  
42 **2011**, *42*, 1614–1619.  
43  
44  
45 (43) Que, R.; Shao, M.; Zhuo, S.; Wen, C.; Wang, S.; Lee, S. T. *Adv. Funct. Mater.*, **2011**, *21*,  
46  
47 3337–3343.  
48  
49  
50 (44) Le Ru, E. C.; Blackie, E.; Meyer, M.; Etchegoin, P. G. *J. Phys. Chem. C.*, **2007**, *111*,  
51  
52 13794–13803.  
53  
54  
55 (45) Cyriac, J.; Wleklinski, M.; Li, G.; Gao, L.; Cooks, R. G. *Analyst.*, **2012**, *137*, 1363–1369.  
56  
57  
58  
59  
60

- 1  
2  
3 (46) Hoflich, K.; Becker, M.; Leuchs, G.; Christiansen, S. *Nanotechnology.*, **2012**, *23*, 185303.  
4  
5  
6 (47) Cole, D. H.; Shull, K. R.; Baldo, P.; Rehn, L. *Macromolecules.*, **1999**, *32*, 771–779.  
7  
8  
9 (48) Peng, H. G.; Kong, Y. P.; Yee, A. F. *Macromolecules.*, **2010**, *43*, 409–417.  
10  
11  
12 (49) Yang, Z.; Fujii, Y.; Lee, F. K.; Lam, C. H.; Tsui, O. K. C. *Science.*, **2010**, *328*, 1676–1679.  
13  
14  
15 (50) McLeod, E.; Liu, Y.; Troian, S. M. *Phys. Rev. Lett.*, **2011**, *106*, 175501.  
16  
17  
18 (51) Cohen, Y.; Reich, S. *J. Poly. Sci. B.*, **1981**, *19*, 599–608.  
19  
20  
21 (52) Menard, E.; Meitl, M. A.; Sun, Y. G.; Park, J. U.; Shir, D. J. L.; Nam, Y. S.; Jeon, S.;  
22  
23 Rogers, J. A. *Chem. Rev.*, **2007**, *107*, 1117–1160.  
24  
25  
26 (53) Reboud, V.; Kehagias, N.; Zelsmann, M.; Striccoli, M.; Tamborra, M.; Curri, M. L.;  
27  
28 Agostiano, A.; Fink, M.; Reuther, F.; Gruetzner, G.; Sotomayor-Torres, C. M. *Appl. Phys. Lett.*,  
29  
30  
31 **2007**, *90*, 011115.  
32  
33  
34 (54) Tamborra, M.; Striccoli, M.; Curri, M. L.; Alducin, J. A.; Mecerreyes, D.; Pomposo, J. A.;  
35  
36 Kehagias, N.; Reboud, V.; Sotomayor-Torres, C. M.; Agostiano, A. *Small.*, **2007**, *3*, 822–828.  
37  
38  
39 (55) Merlen, A.; Chevallier, V.; Valmalette, J. C.; Patrone, L.; Torchio, P.; Vedraine, S.; Flory,  
40  
41  
42 F.; Moula, G. *Surface Science.*, **2011**, *605*, 1214–1218.  
43  
44  
45  
46  
47  
48  
49  
50  
51  
52  
53  
54  
55  
56  
57  
58  
59  
60

## TOC

Electro-hydrodynamic (EHD) instabilities induced in thin polymer films covered with gold nanoparticles leads to a panoply of patterns. The resulting polymer features are decorated with Au nanoparticles that remain uniformly distributed or form fractal arrangements as dictated by nanoparticle mobility. Fractal patterned surfaces have potential as functional substrates for surface enhanced Raman spectroscopy.

