Hierarchically ordered montmorillonite block copolymer brushes

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Hierarchically ordered montmorillonite block copolymer brushes

by

Ross Eric Behling

A dissertation submitted to the graduate faculty
in partial fulfillment of the requirements for the degree of

DOCTOR OF PHILOSOPHY

Major: Chemical Engineering

Program of Study Committee:
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Iowa State University
Ames, Iowa
2010

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DEDICATION

I would like to dedicate this thesis to my loving family; to my father whose encouragement and support was unwavering and who epitomizes the value of hard work, to my brother who motivated me to keep working when times were tough and continues to remind me that there is more to life than just work, and to my mother whose memory I cherish daily and who never got a chance to pretend to read this work.
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CHAPTER 1 LITERATURE REVIEW

General Introduction

In 1993 a research team at the Toyota automotive company made a remarkable discovery; if as little as 5 wt % filler was added to bulk polymer an incredible increase in mechanical and thermal properties was observed. Kojima et al. was studying the in situ polymerization of nylon 6 with the inclusion 1-5 wt% montmorillonite (MMT) or saponite clay and observed significant improvements in heat distortion temperature (HDT), tensile strength, and tensile modulus relative to bulk nylon. Academic and industrial interests in nanocomposite materials have grown geometrically as a result of the seminal work by the Toyota group. The inclusion of a few weight percent of nanoparticles in bulk polymers has demonstrated significant improvements in strength, stiffness, barrier properties, and fire retardance with little-to-no trade-off in desirable properties, such as optical clarity.

The first industrial application of polymer-clay hybrids by Toyota was the injection molding of a new timing belt cover in 1995. Montmorillonite clay was compatibilized via ion exchange of ω-amino acid of varying side chain lengths in order to increase interlayer spacing and act as a catalyst for ε-caprolactam (nylon 6) synthesis. The best specimen, which contained a 5-alkyl organoclay modifier, displayed an HDT, tensile strength, and tensile modulus improvement of 87°C, 55%, and 91% respectively over bulk nylon 6. Okada and Usuki tested the resulting timing belts and reported good rigidity, excellent thermal stability, no warp, and a weight reduction of up to 25% when compared to glass fiber reinforced nylon belts. Okada and Usuki went on to test alternate clay fillers and found that montmorillonite possessed superior HDT, tensile strength, and tensile modulus compared to synthetic mica, saponite, and hectorite.

Nano-scale fillers are superior to conventional fillers thanks to their naturally high aspect-ratios which allow for more interfacial interactions with the polymer matrix. Composites can be defined as
the combination of at least two materials which produce physical properties superior to those of the contributing components. In order to maximize these synergistic effects it is necessary to generate as much interfacial contact between filler and matrix as possible. The principle challenge to working with inorganic nanoparticle fillers is the difficulty of dispersing filler aggregates since most nanoparticles have strong internal ionic and van de Waals forces causing macroscopic aggregates to form.\textsuperscript{4,6} Computational studies have indicated that only through full dispersion of nanoparticles can the greatest property enhancements be achieved. Simulation and theory have suggested the capacity for as much as an order of magnitude improvement in mechanical modulus and barrier properties may occur if long-range silicate orientation is achieved.\textsuperscript{8} Unfortunately, little experimental research has been conducted which focuses on the fundamental structure-property relationships between filler and matrix materials.

**Silicate Nanocomposites**

Inorganic nanocomposites are mostly composed of high aspect ratio silicate filler materials such as hectrite, saponite, laponite, synthetic mica, or montmorillonite. The simple extrusion melt blending of silicates with polymer can provide significant oxygen barrier enhancements over bulk polymer.\textsuperscript{9} If the disk-like silicate fillers are oriented normal to the disk face, (Figure 1.1) a further barrier enhancement can be anticipated.\textsuperscript{6}

Fredrickson and Bicerano examined an idealized model of penetrant molecules diffusing past
semidilute oriented disks. Their simulations estimated that diffusion perpendicular to the disks ($D_\perp$) would decrease according to Eqn 1.1:

$$\frac{D_\perp}{D_o} \sim \frac{1}{(\alpha \phi)^2}$$  

(1.1)

where $D_o$ is the diffusion through bulk matrix, $\alpha$ is the disk aspect ratio ($\alpha = \frac{\text{radius}}{\text{thickness}}$), and $\phi$ is the average volume fraction of disks. Thus, relative to bulk diffusion, composite diffusion in the normal direction decreases as the disk aspect ratio and volume fraction increases squared. This agrees well with intuition and qualitatively with what has been seen experimentally.

Wang et al. have demonstrated significant orientation of organically modified MMT in polypropylene bricks via dynamic packing injection molding. Three separate regions were denoted: the skin, the oriented-zone (a large interior region of the brick), and the core (the center region of the brick). The zones were then compared for three different MMT concentrations (1wt%, 3wt%, & 5wt%) and measured with wide angle x-ray scattering (WAXS). Wang et al. observed nearly anisotropic MMT platelets in the skin and significant order in the oriented zone; the core however had little to no observable orientation. The lower the MMT wt% in the skin and oriented-zone the greater the order, but for all cases highly anisotropic orientation was observed.

Montmorillonite Clay

As was apparent from the original work by the Toyota group, montmorillonite clay (MMT) has been the most promising silicate nanoparticle filler. MMT is a layered silicate which in its raw state is composed of negatively charged tactoid stacks that are counter-balanced by sodium cations, which form macroscopic aggregates. Fortunately those same sodium cations provide an easy means to modify the interlayer structure by means of ion exchange. There are two distinct MMT composite structures (Figure 1.2): intercalated and exfoliated. Intercalated MMT maintains its tactoid structure with single polymer chains dispersed between individual sheets resulting in a tightly packed repeat structure of a few nanometers. Exfoliated MMT is composed of delaminated layers of clay distributed throughout the polymer matrix.
Figure 1.2  Intercalated MMT nanocomposite displays single polymer chains inside clay galleries with interlayer spacing $\sim$1-3nm. Exfoliated MMT nanocomposite exhibits dispersed clay platelets with multiple layers of polymer between individual sheets (interparticle distances $\sim$ 10-100 nm)

**Silicate Intercalation**

Polymer intercalation has been more of a theoretical pursuit than experimental. Lee et al. developed one of the most rigorous models using Lennard-Jones spheres connected by anharmonic springs to study not just homopolymer intercalation but also block copolymer (BCP) intercalation.\(^{11}\) Lee et al. examined several cases: homopolymers with attractive silicate interactions, homopolymers with repulsive interactions, and block copolymers (BCP) with one attractive block and one repulsive block. As expected, chains with an attractive driving force will spontaneously diffuse into the clay galleries, while repulsive chains avoid interlayer galleries. Remarkably, chains with very high attraction to the silicate fail to fully intercalate because they attach too strongly to the edges of the clay interlayer, preventing further intercalation. BCPs with a moderately attractive block will intercalate both blocks into the clay gallery with the repulsive block aggregating away from the clay surface; while BCPs with a weakly attractive block will only partially intercalate into the interlayer with the repulsive block remaining outside of the galleries.

Although the study of polymer intercalation has been dominated by simulation and modeling, there are a few exceptions.\(^{5,12,13}\) Gianellis et al. reported a comparison of mean-field simulation to experimental results of organically modified silicates intercalated by polymer in the melt state.\(^{14,15}\) An extensive examination of energetic and entropic factors yielded the conclusion that favorable interactions between the alkylammonium modifier and polymer will compensate for confinement penalties.
Chrissopoulou et al. studied how the organic modifier affected polymer intercalation and included BCP experiments. Their results support Lee’s work in that the degree of attraction, altered by compatibilizer content, dictated the degree of intercalation with the best cases resulting in exfoliated silicates.

**Silicate Exfoliation**

Extensive research has gone into exfoliating silicate tactoids by means of melt mixing, solvent blending, or *in situ* polymerization. Regardless of method exfoliation is aided by the addition of an organic modifier to make hydrophilic clays more amenable to hydrophobic polymer interactions and to increase interlayer spacing. *In situ* polymerization has proven to be the most promising method for silicate exfoliation and can be accomplished with large chain *graft-to* reactions or surface growth *graft-from* polymerizations. *Graft-to* polymerizations encounters ever increasing steric hindrance with increasing graft brush density therefore *graft-from* polymerization is preferable to produce high graft-density brushes. The use of sonication to disperse clay aggregates in conjunction with organoclay modifiers has further improved silicate exfoliation. It is beneficial to examine these techniques independently to better understand the contribution of each.

**Organoclay Modifications**

Chrissopoulou et al. stated that “The most important factor controlling the structure and the properties [of a nanocomposite] is the ratio of additive to nanoparticles”, which is supported by the vast number of publications that have been written to discuss the role that the organic modifier plays in silicate NCPs. The degree of intercalation, the control of polymer miscibility, and influence over final microstructure have all been ascribed to the role of the organic modifier. The need for organic modifiers arises from the fact that inorganic silicates are hydrophillic while most polymers are hydrophobic. The two primary means for modifying silicates are (1) cationic ion exchange and (2) BCP intercalation where one block has a moderate to strong affinity for the clay.

Lee et al. examined the exfoliation of MMT with poly(ethylene-33mol%-vinyl alcohol) (EVOH) or poly(styrene-27mol%-acrylonitrile) (SAN) along with an organic modifier (OMMT). Samples with 5 wt % OMMT were prepared by melt mixing with dynamic melt intercalation; EVOH, being a highly
Figure 1.3  Electromicrographs depicting the effect of sonication time on commercially modified MMT exfoliated from the in situ ring opening of dicyclopentadiene. (a) 1 hr sonication time. (b) 3 hr sonication time. Reproduced from Ref. 21

hydrophillic polymer, intercalated readily into the OMMT galleries. Lee et al. found that the EVOH is so strongly hydrophillic that it produces a “glue-effect”, which limits the diffusion of polymer into the gallery in agreement with theory.\textsuperscript{11} SAN however is only weakly hydrophillic so it was able to disasssociated the OMMT platelets to produce an exfoliated composite. EVOH-OMMT, while less exfoliated than SAN-OMMT, exhibited better tensile strength and tensile modulus enhancements over bulk, which was attributed to strong attractive energy between the matrix and filler.

\textbf{Sonication}

The use of sonication during silicate nanocomposite mixing/synthesis has a pronounced enhancement on the degree of exfoliation of the clay tactoids. Figure 1.3 is work by Yoonessi et al. where commercially modified MMT was exfoliated from the in situ ring opening of dicyclopentadiene.\textsuperscript{21} Yoonessi et al. examined different sonication times for 0.5-5 wt\% filler loadings and found that two hours of sonication was sufficient to produce full exfoliation for all samples as measured by x-ray diffraction (XRD). Yoonessi et al. was further able to demonstrate that dispersed clays easily aligned from shear fields.

The benefits of sonication has been demonstrated for melt mixing,\textsuperscript{22} solvent blending,\textsuperscript{17} and in situ polymerization.\textsuperscript{23} Solvent blending is perhaps the easiest method for gaining an immediate and
significant improvement from sonication. Morgan and Harris compared the effects of clay type and the use of sonication on the degree of exfoliation of the final composite. They dispersed fluorinated synthetic mica (FSM) or MMT in chlorobenzene via 20 minutes of stirring with or without an additional 5 minutes of sonication, after which polystyrene was added and further stirred for 4 hours then dried. XRD showed a significant improvement in dispersion from the application of sonication to the solution for both clay types. TEM however showed a much more pronounce benefit from sonication for the MMT composite than the FSM, which the authors suggested might be due to MMT’s lower ionic charge.

Morgan and Harris included in their publication the importance of sonicating prior to polystyrene addition, since sonication has been known to cause polymer degradation. Torkelson et al. were able to elucidate this mechanism by mixing high molecular weight polystyrene (HMW-PS) with pyrene-labeled polystyrene (PL-PS) and monitoring changes in the gel permeation chromatography (GPC) fluorescence detector signal. After only 2 minutes of sonication a noticeable increase in fluorescence signal at the HMW-PS elution time occurred indicating chain scission and combination of HMW-PS and PL-PS radicals. Further confirmation by Torkelson et al. was demonstrated using a selective solvent for poly(n-butyl methacrylate) (PnBMA). PL-PS was mixed with PnBMA in solution and sonicated for 2-10 minutes, then PnBMA polymer was isolated using solvent selection to measure BCP formation. A control sample with no sonication was isolated using solvent selection and displayed no fluorescence signal proving that no PL-PS signal would occur unless BCP formation took place. Significant amounts of PL-PS scission and PnBMA combination was measured after sonication as confirmed by liquid chromatography fluorescence detection. Although sonication of polymer solutions and melts has the potential for some small percentage of polymer degradation, it is still a necessary element to achieve silicate exfoliation.

**Controlled Radical Polymerization**

Another necessary element for silicate exfoliation is the use of a “living” polymerization method. Due to the confined geometry of planar substrates (such as silicates, silicon wafers, and gold sheets) a “living” polymerization method is required to eliminate immediate chain termination. Anionic poly-
merization is historically the most prevalent controlled polymerization method and is easily applied to polymer brush synthesis from gold surfaces.\textsuperscript{25} Synthesis from silica or clay is more difficult due to their affinity for oxygen and water, where even a few parts per million of either is enough to terminate anionic polymerization. Liao et al. found a method for conducting \textit{in situ} anionic polymerizations of polybutadiene, polyisoprene, and poly(styrene-\textit{b}-butadiene) in the presence of organically modified montmorillonite (OMMT).\textsuperscript{26,27} Their successful exfoliation of OMMT via anionic synthesis was accomplished by the addition of excess \textit{n}-butyllithium to scavenge bound oxygen/water from the OMMT surface. While this method is capable of preparing polymer/MMT composites via anionic polymerization, it lacks the ability to precisely control the final molecular weight (a major benefit of anionic synthesis). Additionally this method is only suitable for preparing free polymer with OMMT inclusions. To date there has been no published examples of tethered anionic initiators from MMT; thus controlled radical polymerization (CRP) methods have been used when tethered polymers are desired.

\textbf{Atom Transfer Radical Polymerization}

Atom transfer radical polymerization (ATRP) is a CRP technique adapted from Kharasch addition by Matyjaszewski and Wang in 1996.\textsuperscript{28} A transition metal catalyst is complexed with a multi-amine ligand that facilitates the removal of a terminal halide to produce free radicals. This is a reversible step in which the halide-capped dormant phase is favored by many orders of magnitude, resulting in very few active radicals at any given time. This persistent radical effect is a result of the significantly more stable dormant state and essentially eliminates radical-radical termination reactions until very high conversion. By suppressing the termination mechanism a “living” polymerization is achieved, which produces nearly monodisperse samples and enables the capacity for complex compositions and topologies.

ATRP is described as a “living” radical polymerization method because it has minimal termination reactions and can be used to produce polymers of narrow polydispersity (PDI). There are two mechanisms for minimizing PDI of ATRP synthesis, changing deactivator concentration or changing temperature. Increasing Cu\textsubscript{II}X\textsubscript{2} concentration will decrease PDI by slowing reaction kinetics, although when approaching the limiting case reactions become prohibitively slow. As for changing reaction
temperature it appears that either increasing or decreasing temperature can improve PDI. Increasing reaction temperature increases the apparent rate constant of propagation, $k_p$, relative to the rate constant of termination, $k_t$, hence decreasing PDI, however increasing temperature often results in a higher rate of side reactions, thermal self-polymerization, and undesirable chain transfers. Research by Samadi et al. indicates that very narrow PDIs can be achieved by significantly reducing the reaction temperature to favor halogen end-capped equilibrium further reducing radical-radical termination.

Krzysztof Matyjaszewski is credited with the well-accepted reaction mechanism for copper catalyzed bidentate assisted ATRP (Figure 1.4). A bromine capped dormant chain (depicted on the left) reversibly undergoes uncapping via a copper catalyst to allow free radical polymerization (on the right). Consistent with the strongly favored dormant equilibrium, the active radical quickly returns to the dormant state. In this figure two bipyridine ligands are complexed with the copper catalyst, but tridentate ligands only require a single ligand to effectively complex transition metals. Although ATRP has been a very successful CRP method, alternate methods exist which do not require a transition metal catalyst, namely nitroxide mediated radical polymerization (NMRP) and reversible addition-fragmentation chain transfer (RAFT).
Nitroxide Mediated Radical Polymerization

Work by Solomon and Rizzardo in 1985 is acknowledged with the development of NMRP, but for nearly a decade afterwards NMRP was limited to only styrene and styrene derivative monomers. During that time the primary nitroxide used was 2,2,6,6-tetramethylpiperidinyloxy (TEMPO) since it was commercially available and produced narrow PDIs. More recently the desire for a more universal nitroxide has prompted numerous groups to test hundreds of alternate compounds. Work by Hawker et al. finally proved fruitful with the discovery of a mediating species capable of “controlled” polymerization of styrenes, acrylonitriles, acrylamides, and acrylates. NMRP has two primary disadvantages relative to its ATRP analog. (1) NMRP is activated via thermal homolysis, requiring higher reaction temperatures, allowing for thermal self-polymerization of monomers and (2) NMRP is as yet unable to polymerize methacrylate monomer species.

Radical Addition-Fragmentation Chain Transfer

RAFT polymerization utilizes thioesters or dithiobenzoates to reversibly release active radicals in a controlled manner. The dithio compound allows for two separate chains to remain dormant on the same RAFT agent. Chong et al. have shown RAFT to be a very versatile method for producing functional polymers of narrow PDI. Chong et al. further demonstrated RAFT’s ability to produce block copolymers with minimal homopolymer contamination (< 5%). Other than the unavoidable side production of homopolymer, the greatest limitation of RAFT block copolymer synthesis is that the first block must disassociate from the thioester more readily than the second block. The flexibility to choose the order of block addition based on chemistry is necessary to create systems with complex architectures (e.g. polymer brushes).

Block Copolymers

Nearly all polymers are macroscopically immiscible due to small chemical incompatibilities between the two monomeric species which add up to large enthalpic energies outweighing the entropic energy of mixing. Block copolymers (BCPs) are a class of polymers which exhibit novel material behavior arising from a covalent linkage between two distinct polymer chemistries. The covalent linkage
of two polymers limits the amount of segregation that is achievable for each chain, thus microphase separation occurs. The difference in volume fraction and chemical incompatibility between the A and B blocks governs the three dimensional structure assumed by the BCP domains, known as the morphology. Figure 1.5a presents the results of self-consistent mean-field theory (SCFT) calculations for diblock copolymer microphase separation. Figure 1.5b provides an experimental phase diagrams of linear poly(isoprene-styrene) BCPs to compare with the theoretical results.

The governing equation for BCP microphase separation is known as the Flory-Huggins equation and it is written as:

$$\frac{\Delta G}{RT} = N_1 \ln(\phi_1) + N_2 \ln(\phi_2) + \chi N_1 \ln(\phi_1)$$  (1.2)

where $\Delta G$ is the Gibb’s free energy, $N_i$ is the degree of polymerization of $i$, $\phi_i$ is the volume fraction of $i$, and $\chi$ is the Flory-Huggins interaction parameter. The interaction parameter, $\chi$, is an empirical relationship introduced by Flory and Huggins to account for the separation during mixing of two polymers. Besides being a function of the chemistry of the two polymer species $\chi$ is also a function of temper-
Figure 1.6   Illustration of multiple block copolymer architectures reproduced from Ref. 43.

nature. The Flory-Huggins equation clearly shows that the best way to enhance microphase separation is to either increase polymer chain length (increase $N_f$) or to change polymer species to create a larger difference in polymer compatibility (increase $\chi$).

Although linear diblock BCPs are the most frequently studied due to the relative simplicity of their preparation, linear multi-blocks, star blocks, and branched-arm BCPs all have been developed (Figure 1.6).43–45 These complex BCP architectures can be applied to many diverse applications such as polymer blend compatibilizers (linear diblocks), tape adhesives (linear triblocks), and engine oil viscosity regulation (star BCPs).

**Gyroid Phase**

Given that the gyroid phase is a three dimensional bicontinuous network, it is something of a novelty to BCP chemistry and as such it was perhaps the last stable phase to be well understood.46,47 Initially the gyroid phase was only investigated at substantial molecular weights (100-200 kg/mol) where its stability could not be confirmed. Lodge et al. recognized the slow dynamics of large chain BCP equilibrium, so to verify that the gyroid phase was not a meta-stable state they synthesized short chain BCP’s (10-13 kg/mol) with very large segregation values.44 The strength of segregation of a
poly(ethylethylene)-b-poly(isoprene) linear diblock was adjusted by the degree of fluorination of the poly(isoprene) block. An approximated interaction parameter of the fluorinated isoprene was calculated from the strong-segregation theory (Equation 1.3) and small angle x-ray scattering (SAXS) data:

\[ D = 1.10\bar{a}N^{2/3}\chi^{1/6} \]  

where the domain spacing D comes from the SAXS relation \( D = 2\pi/q* \), \( \bar{a} \) is the weighted average statistical segment length, and N is the total degree of polymerization. In 1994 Matsen et al. reported a theoretical upper limit of the gyroid phase at \( \chi N \approx 60 \), shown in Figure 1.6a.\(^{41}\) Experimentally in 2003, Lodge et al. found that the gyroid phase was stable for \( 25 < \chi_{eff}N < 120 \).\(^{44}\) To further prove the stability of the gyroid phase, Lodge et al. solvent cast films using selective solvents for one of the BCP domains resulting in the formation of meta-stable morphologies. After annealing the meta-stable BCP’s for several hours at 150°C the gyroid SAXS pattern was recovered. As a result of the work by Lodge et al., in 2006 Cochran et al. reexamined the SCFT simulations of the high \( \chi N \) stability of the gyroid phase.\(^{48}\) They concluded that the gyroid phase persists for a narrow range of BCP compositions up to high segregation due to packing frustration of the cylindrical phase and increased interfacial energy restrictions for the lamellae phase.

**Polydispersity Effects**

Another interesting phenomenon of BCP morphology is that increasing polydispersity can lead to transitions in morphology while maintaining constant volume fraction. Lynd and Hillmyer investigated effects on BCP morphology from the selective increase in PDI of the \( DL \)-lactide block of a poly(ethylene-alt-propylene)-b-poly(\( DL \)-lactide) diblock copolymer.\(^{49}\) Lynd and Hillmyer observed a linear increase in lamellar domain spacing with increasing PDI resulting from large chains filling the void space between domains more easily without the need for unfavorable stretching. When non-lamellar samples were examine the asymmetric increase in PDI of the minority block caused phase transitions towards increased mean interfacial curvature (i.e. lamellar became gyroid etc.), conversely increasing PDI in the majority block decreased mean interfacial curvature. The change in interfacial curvature was attributed to the movement of the interface towards the block that previously had the
larger stretching penalty.

**Block Copolymer Nanocomposites**

Block copolymer synthesis offers an easily tuneable method for producing self-assembled structures on the nanometer length scale. Therefore BCPs have been extensively used to tailor the orientational distribution of nano-scale filler materials. The use of block copolymers in NCPs to date has been largely limited to a supporting matrix for the suspension of homopolymer functionalized inorganic fillers.\textsuperscript{50–52} The most prevalent method has been the grafting of polystyrene to montmorillonite and solvent blending with a compatible block copolymer; BCP morphology in these cases is nearly always lamellar. The simple blending of clays with block copolymers has not yet yielded any greater degree of control of particle distribution than traditional homopolymer nanocomposites.\textsuperscript{53} Even though it has been recognized that the control of the particle morphology is crucial, there has been little done to develop the technologies to realize it. The few exceptions to this practice will be mentioned here in detail as they mark significant strides forward in the field.

Bockstaller, Mickiewicz, and Thomas wrote an excellent review on the incorporation of homopolymer grafted NCPs into BCPs domains.\textsuperscript{50} The most interesting phenomenon they studied was the selective aggregation of gold nanocrystals and silica spheres in polystyrene-\textit{b}-poly(ethylene propylene) (PS-PEP). The inclusion of gold and silica was first studied independently then concurrently in the same BCP showing that the gold particles segregate to the BCP interface while the silica particles prefer the center of the PEP domain. The characteristic length scale and aspect ratio of the filler material is responsible for where a particle segregates in the BCP domain. In a different experiment Thomas et al. synthesized gold nanoparticle with polystyrene (Au-PS) brushes and doped a PS-\textit{b}-2-vinylpyridine BCP. The subsequent rise in PS volume fraction due to Au-PS domain selectivity induced a morphology transition from lamellar to cylindrical. Thomas et al. conducted a similar study using MMT grafted PS brushes (MMT-PS) and once again found selective segregation into PS domains.\textsuperscript{51} Although domain deformation to accommodate MMT-PS particles was much greater with MMT-PS than with Au-PS, the lamellar morphology persisted thanks to the compatibility of MMT plate-like geometry with the lamellar structure. The three dimensional matching of nanoparticle geometry with BCP domain geometry is
known as symmetry matching and is an important consideration to develop useful nanocomposites.

The functionalization of nanoparticles for incorporation into BCP domains requires the same set of chemistries and techniques developed for surface modification, therefore a discussion of the development of those methodologies follows. The concept of “pinned micelles” was first published in 1996 by Zhulina et al. using self-consistent field theory calculations. Pinned micelles are BCPs that are bound to a substrate and upon exposure to a selective solvent, the favorable block undergoes chain stretching rearrangements to shield the unfavorable block from solvent interactions. The response of BCPs to selective solvents dramatically changes the properties of modified surfaces, with hydrophobicity being the most frequently modified property. The use of BCPs in surface chemistry has been further varied by attaching Y-branched initiators and that produce BCP that are pinned at the BCP interface. The formation of pinned micelles via Y-branched initiation has been demonstrated by Boyes et al. using a combination of ATRP and NMRP synthesis.

The grafting of CRP initiators, targeting specific molecular weight, and the minimization of brush PDI are all techniques developed for surface modification that have now been directly applied to grafted BCP nanocomposites. Zhao and Shipp synthesized the first published BCP nanocomposite brush with a polystyrene-\textit{b}-butyl acrylate grafted from MMT clay. The micrographs they published were of solvent cast films that displayed minimal phase separation and contained butyl acrylate domains of only 2-5 nm. Di and Sogah utilized a Y-branch initiator to simultaneous synthesize polystyrene and poly(caprolactone) from MMT and were able to produce significant block lengths (Mn \( \sim \) 10-70 kDa) with reasonable PDIs (\( \sim \) 1.3-1.5) demonstrating the full gamut of surface synthesis techniques.

**Project Description**

The focus of my research has been the synthesis, characterization, and mechanical testing of block copolymer/silicate nanocomposites. To that end, the controlled synthesis of well-defined homopolymer and diblock copolymers from montmorillonite clay was achieved. The reaction scheme for this process (Scheme 1.7) began with an ion exchange of active initiator onto the MMT surface. Successful ion exchange is confirmed via x-ray diffraction and thermal gravimetric analysis.

Brush synthesis occurs in two steps: (1) styrene polymerization is conducted with sonication to
enhance MMT exfoliation and (2) the polystyrene brushes are cleaned of copper complexes and re-polymerized with butyl acrylate in excess to act as a solvent. Samples are characterized via gel permeation chromatography and nuclear magnetic resonance spectroscopy then ordered by melt annealing and rheological shear alignment before being examined via transmission electron microscopy.

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CHAPTER 2 EXPERIMENTAL DETAILS

This chapter is a comprehensive list of the chemicals and instruments that are essential to the synthesis and study of montmorillonite block copolymer brushes. Chemical purification techniques are included where necessary and instrument testing conditions are also provided. Lastly both a summary and a detailed description of synthesis conditions is provided for the reproduction of the results enclosed in this thesis.

Experimental Materials

2-bromoisoobutyryl bromide (98%), benzyl bromide, copper(II) bromide (CuIIBr2), N,N,N′,N′,N′,N′-pentamethyldiethylenetriamine (PMDETA), ethyl-2-bromopropionate, and Dowex Marathon MSC hydrogen form exchange resin (20-50 mesh) were purchased from the Aldrich Chemical Co. and used without further purification. Tert-butyl acrylate (98%) and n-butyl acrylate (98%) purchase from the Aldrich Chemical Co. and purified with basic alumina flash chromatography and degassed prior to use. Copper powder (99.7%) was purchased from the Aldrich Chemical Co. and heated to 200°C under 5% H2/nitrogen gas for 6 hours prior to use. 11-bromo-1-undecanol (98%) was purchased from the Aldrich Chemical Co. and recrystallized from 75 vol % methanol / 25 vol % deionized water prior to use. Methanol, isopropanol, acetone, toluene, pyridine (99%), and trimethylamine (33 wt % in ethanol) were purchased from the Fisher Scientific Co. and used without further purification. Styrene (99%) was purchase from the Fisher Scientific Co. and purified with basic alumina flash chromatography and degassed prior to use. Montmorillonite clay (MMT) was generously supplied by Southern Clay Products Inc. (92 mequiv / 100g).

Copper(I) Bromide Purification. Copper(I) bromide (99.999%) was purchased from the Aldrich Chemical Co. and purified with glacial acetic acid, rinsed with absolute alcohol and ethyl ether, dried under vacuum, and stored under nitrogen prior to use. While the cleaning of copper(I) bromide with glacial acetic acid works well, it requires highly pure starting material and the increased purity of copper(I) bromide becomes prohibitively expensive. In an alternate preparation method, 10g copper(I) bromide (98%) was stirred for 24 hours with 50mL sulfuric acid (95-98%) then rinsed extensively with absolute alcohol and ethyl ether on a filter frit. It was then transferred to a preheated vacuum oven (80°C), dried for 20-25 minutes and moved into the glovebox for storage.
Figure 2.1 Thermogravimetric analysis of volatile components of MMT (green), functionalized MMT (blue), and 38 kDa PS-MMT nanocomposite (red).

**Equipment**

**Gel Permeation Chromatography (GPC).** The molecular weights and molecular weight distributions of polymer samples were determined via gel permeation chromatography with respect to polystyrene (PS) or poly(methyl methacrylate) (PMMA) standards by a Waters 717 autosampler and Waters 515 HPLC system. It should be noted that PS and PMMA standards report very similar values, with an average difference of 10-15% with respect to the total molecular weight.

**Nuclear Magnetic Resonance (NMR).** $^1$H NMR spectra were determined on a Varian VXR-300 spectrometer in CDCl$_3$ or deuterated dimethyl siloxane at room temperature.

**X-Ray Diffraction (XRD).** X-ray diffraction experiments were conducted using a Scintag XDS-2000 powder diffractometer utilizing a copper $K_{\alpha}$ source in order to determine d-spacing of clay layers prior to and after ion exchange. Clay d-spacing was determined using Bragg’s law: $d = 2 / (\lambda \sin(\theta/2))$. A complete XRD analysis of the ion exchange process is summarized in Chapter ?? and graphically displayed in 3.3.

**Differential scanning calorimetry (DSC).** DSC was used to investigate the Tg confinement effects of MMT tethered polymers. Samples were tested on a TA Instruments DSC-Q2000 using a heat-cool-heat cycle to erase thermal history at a heating/cooling rate of 10°C min from -100°C to 150°C. Prior to DSC measurements all samples were dried above 150°C under vacuum for 12-48 hours (until chamber pressure measured < 150 millitorr) to eliminate the effects of small molecule plasticizing.
Figure 2.2 Time-temperature superposition of MSnB-68-78 demonstrating overlay of frequency sweeps at 20°C intervals from 120-220°C.

**Thermogravimetric analysis (TGA).** A TA Instruments TGA-Q50 was used to determine success of initiator ion exchange and final clay content. Samples were heated to 800 °C using a heating rate of 10°C min⁻¹. TGA initiator experiments were investigated under atmospheric conditions to determine final clay content as TGA tested under nitrogen airflow displayed incomplete degradation. A comparison of virgin MMT, fMMT, and a 38 kDa PS-MMT nanocomposite is presented in Figure 2.1.

**Transmission Electron Microscopy (TEM).** TEM experiments were conducted on a Tecnai G² F20 scanning / transmission electron microscope at a high tension voltage of 200 kV. Alternatively, TEM experiments were conducted on a JEOL 1200EX at an accelerating voltage of 80 kV. Ultrathin (~ 100 nm) sections of material were obtained at cryogenic temperature using a Leica Ultramicrotome Ultracut 125UCT with a Leica EM FCS cryo-stage.

**Rheology** A TA Instruments ARES-LS1 strain controlled rheometer with convection oven was used under nitrogen gas flow to prevent polymer degradation. Prior to testing, samples were pressed with a Carver press with 1,000 lb of force at 150 °C into 25 mm diameter and 1 mm thickness discs. Samples were tested in a parallel plate geometry with frequency scans at multiple operating temperatures. At each operating temperature the linear viscoelastic (LVE) region was identified and samples were only tested within this regime. The LVE was identified by measuring the modulus as a function of strain and when the modulus just began to change with strain it was notated as the maximum allowable strain. Master curves of the viscoelastic spectra (see Figure 2.2) were generated using time-temperature superposition using oscillatory experiments in frequencies between 0.1
and 100 rad/sec. Frequency scans were taken every 20°C from 120-220°C.

**Small Angle X-ray Scattering (SAXS)** Small angle x-ray scattering experiments were conducted on a Rigaku copper Kα source instrument. The microfocus cathode source was operated at 40kv and 30mA and the CCD detector measured final x-ray scattering in pico-amperes. Samples were enclosed in Kapton tape and suspended in an evacuated chamber with a sample to detector distance of 2 m. Silver behenate was used as a calibration standard.

**Tensile Testing** Tensile testing experiments were conducted on a Instron Load Frame at 0.33mm/min. Dog bones were prepared by first melt pressing a 2.5in×1.25in rectangle in the Carver press at 160°C then having samples milled down to the desired shape by the machine shop; dog bone necks were 12mm in length and ~4mm in width.

**Initiator Synthesis**

**Summary of Initiator Synthesis** 11’-(N,N,N-Trimethylammonium bromide)-undecyl-2-bromo-2-methyl Propionate. The production of active (inactive) initiator is a simple two step synthesis (Figure 2.3) combining 11-bromo-1-undecanol with 2-bromoisobutyryl bromide (2,2-dimethyl acetyl chloride) and subsequently adding trimethyl amine to produce a charged alkyl chain with(out) a terminal halide. The two species can then be mixed into desired molar ratios and ion exchanged onto the anionic clay surface to produce prescribed graft-densities. Initiator and MMT clay were continuously stirred in 50 mL of acetone in an erlenmeyer flask at a mass ratio of 2:2.5 respectively for 96 hours. Functionalized MMT was then filtered, washed alternately with 25 mL water and acetone, then dried in vacuum at 80°C for 24 hours, and ground into a fine powder. Successful ion exchange was confirmed via XRD and TGA, and then stored under nitrogen.
Detailed Initiator Synthesis  The detailed synthesis of active initiator began with the premixing of 20g 11-bromo-1-undecanol in 300mL diethyl ether in an erlenmeyer flask followed by the addition of 14.3mL pyridine. Mixing vessel geometry was found to significantly affect final yield, so an erlenmeyer flask is strongly recommended. A second solution of 14.3mL 2-bromo-2-methyl propionyl bromide in 50mL diethyl ether was prepared and poured into a graduated burette for dropwise addition to the bulk solution. Ether is highly volatile so it is necessary to cover all open containers with foil, especially the top of the graduated burette. Dropwise addition should take at least 1h followed by 5h of continuous stirring.

The salt that formed was filtered out of the bulk solution and was washed with a total of 125mL diethyl ether. The eluent was then transferred to a 1000mL round bottom flask and roto-evacuated until a light yellow oil remained. Note that ether is extremely volatile so great caution must be exercised to prevent bumping. After rotovapor extraction 75mL, diethyl ether was added back into the solution and three liquid-liquid extraction cycles with 75mL of deionized water were conducted. A typical extraction cycle began with 1h of slow stirring with the ether and water in the round bottom followed by at least 1h of separation in a sep-funnel. The oil and water phases will change positions in the funnel as the salt content of the oil/ether phase decreases, so it is important to carefully select the oil containing phase during liquid-liquid extraction. The slight yellow tint of the desired product and the strong odor of the ether is useful for choosing the correct phase. After extraction 5g magnesium sulfate was added to the ether solution and it was allowed to stir for at least 1h to remove all remaining water. 50mL of hexanes were then added to the ether solution and allowed to stir. A basic alumina flash chromatography column was prepared containing 100mL alumina and filled with hexanes until the liquid level rose above the alumina. The hexane/ether solution was then carefully added to the flash chromatography column and passed through the column twice, being sure to never let the liquid level pass below the top of the alumina. The remaining ether solution was then pushed through the column with 50mL of pure hexanes. A second rotovapor extraction was conducted to reduce the solution to a viscous yellow oil, henceforth referred to as product 1.

Product 1 success was confirmed via NMR with an average yield of ~25g. All 25g of product 1 was combined with 65mL 30vol% trimethyl amine in ethanol and 200mL of ethanol into an erlenmeyer flask. The flask was then immersed in an oil bath and stirred at 40°C for 96 hours. The resulting solution was then roto-evaporated at 60°C until a yellow oil was recovered. The oil was transferred to labeled 20mL scintillation vials with a pasteur pipette; no vial was filled beyond 10mL as the final product greatly expands upon solidification. The vials were then very slowly dried (2-3 days) under first static then dynamic vacuum to yield a white waxy solid. Further extensive purification steps were originally undertaken by Zhao et al., but were omitted in this work and confirmed by NMR to be unnecessary.
Polymer Synthesis

To a 50mL round bottom flask equipped with a 24/40 rubber septum and magnetic stir-bar were added benzyl bromide (or ethyl-2-bromopropionate), styrene, CuI Br, CuII Br2, and PMDETA. Chemicals were added in molar ratios relative to the amount of initiator used; 1 : 1000 : 1 : 0.06 : 1.06 respectively. All samples were prepared in a nitrogen (argon) glove box and pressurized with 5 psi N2 (Ar) during reaction. The sealed round bottom was removed from the glove box and placed in an oil bath at constant temperature with continuous stirring. Reactions at 100°C were found to produce narrow polydispersities (Mw/Mn < 1.15) and have relatively fast kinetics, reaching ~50% conversion after only 12 hours. The preceding synthesis procedure was derived from numerous articles written by Matyjaszewski et al.,63,64 principal amongst them is an extensive review on ATRP.65 Samples were precipitated in a 3:1 methanol to isopropanol volume ratio, decanted, and redissolved in toluene. This cycle was repeated until precipitated polymer appeared white. CuI Br and CuII Br2 were premixed prior to addition to the reaction vessel to greatly improve reproducibility of results. CuI BrCuII Br2 mixtures were used for a maximum of two weeks before a fresh batch was prepared to remove the possibility of airborne contaminants decelerating reaction kinetics.

Synthesis of Bulk Block Copolymers To a 50mL round bottom flask equipped with a 24/40 rubber septum and magnetic stir-bar were added dry PS macroinitiators, tBA, CuI Cl, Cu0, and PMDETA. Chemicals were added in molar ratios relative to amount of macroinitiator used; 1: 3000: 1 : 0.05: 1.05 respectively. All samples were prepared in a nitrogen glove box and pressurized with 5 psi N2 during reaction. The sealed round bottom was removed from the glove box and placed in an oil bath at 65°C with continuous stirring. Reaction time was varied to produce blocks of differing lengths. ATRP of both tBA homopolymers and BCPs has been described in detail elsewhere.66

Synthesis of Confined Homopolymers The synthesis of homopolymer from functionalized MMT (fMMT) followed the same procedure outlined for bulk PS synthesis except in place of benzyl bromide, bromine terminated initiators tethered to the clay surface were used. A sonication probe was placed within one inch of the exterior of the reaction vessel for the duration of the experiment to maximize exfoliation without compromising the closed atmosphere. The synthesis of PS from MMT via ATRP has been previously reported by others, excluding the application of ultrasonication which is unique to this work.62,67 tBA and nBA were also synthesized as homopolymers from the fMMT surface at 75°C and 65°C respectively. nBA samples initially showed poor PDI, so they were allowed to sonicated for 3-6 hours at 30°C before increasing the bath temperature to 65°C even after this added step the acrylate samples demonstrated slightly elevated PDI relative to the PS samples (1.15 < Mw/Mn < 1.25).
Synthesis of MMT Block Copolymer Brushes (MMBs) To a 50mL round bottom flask equipped with a 24/40 rubber septum and magnetic stir-bar were added dry PS macroinitiators, nBA, CuI Br, Cu0, and PMDETA. Chemicals were added in molar ratios relative to amount of macroinitiator used; 1 : 4000 : 2 : 0.1 : 2.1 respectively. All samples were prepared in an argon glove box and pressurized with 5 psi Ar during reaction. The sealed round bottom was removed from the glove box and placed in an oil bath at 65°C with continuous stirring. Reaction time was varied to produce blocks of differing lengths. In the reverse reaction where PS was synthesized from nBA macroinitiators a second equivalent amount of CuI Br, Cu0, and PMDETA were cannula transferred into the reaction vessel after 12h using additional styrene monomer as a solvent.

TEM Preparation

Shear Alignment of MBBs All 25mm nBA containing MBB polymer samples were shear aligned on a TA Instruments ARES-LS1 strain controlled rheometer with convection oven heated to 160°C in the parallel plate geometry. Samples were initially rotated at 0.5 rad/sec for 5 minutes to assure polymer wetting of the rheology plates. Extra polymer was then trimmed from the edges with a wooden spatula and the sample was compressed to < 1.75mm. The rheometer was then switched to dynamic mode and an isochronal time sweep was conducted for 20 minutes with a 1% strain and a frequency of 1 rad/s. After shear alignment samples were cryo-fractured to obtain an ideal sample for cryo-ultramicrotomy. Samples that failed to demonstrate microphase separation were re-sheared in steady mode for 1hr at 0.5 rad/sec, this primarily occurred in very asymmetric diblocks.

Cryo-Ultramicrotomy & Staining MBB’s of narrow PDI and appreciable block sizes were prepared for TEM by means of cryo-ultramicrotomy. Fractured samples were mounted on metal posts using a sucrose solution as an adhesive, then cooled to cryogenic temperatures (-100°C). Mounted samples were trimmed using a DiATOME ultratrim with a 35° blade angle, and subsequently sectioned on a DiATOME ultra-thin cryo-knife also with a 35° blade angle. Cut sections were transferred to 400 mesh copper grids with the same sucrose adhesive suspended on a EMS Perfect Loop tool. TEM contrast was obtained by exposing sections of MBBs to RuO4 vapors formed in situ by the combination of 5 mL 13% aqueous sodium hypochlorite and 40 mg ruthenium trichloride for 5-7 minutes. A good indicator of sufficient staining is that the double sided tape used to mount the copper grids began browning but had not yet turned black.

Preparation of PVP-sucrose solution To a 100mL beaker 78.73g sucrose, 20mL of 5× working strength PIPES saline solution, and 60mL of deionized water was added. The solution was heated to 50°C and stirred continuously until the sucrose dissolved. To a 250mL beaker 20g of 10kDa polyvinylpyrrolidone (PVP-10) was added and then 10mL increments of sucrose solution was added until the 100mL beaker was empty. An additional 10mL of warm deionized water was added, then the 250mL beaker was covered and allowed to stir.
overnight. Finally the PVP-sucrose solution was uncovered, measured into aliquots, and stored in the freezer.

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CHAPTER 3 INFLUENCE OF GRAFT DENSITY ON KINETICS OF SURFACE-INITIATED ATRP OF POLYSTYRENE FROM MONTMORILLONITE

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Abstract

Here we report the kinetics of the surface initiated atom transfer radical polymerization (ATRP) of styrene from the surface of functionalized montmorillonite clay as a function of graft density. Compared to analogous ATRP reactions with free initiator, we observe a seven-fold increase in the polymerization rate at the highest graft density, ≈ $1 \text{chain} / \text{nm}^2$, while bulk kinetics are recovered as the graft density is reduced. We hypothesize that this phenomenon is a consequence of local concentration heterogeneities that shift the ATRP equilibrium in favor of the active state, and present a phenomenological based kinetic model that accounts for our data. These findings present an important consideration relevant to the design of precisely defined molecular architectures from surfaces via surface-initiated ATRP.

Introduction.

Atom transfer radical polymerization (ATRP) has proven a versatile method for synthesizing polymers of narrow polydispersity at targeted molecular weights.⁶⁸ ATRP provides an industrially scalable process through means of an inexpensive copper-mediated halide exchange using commercially available ligands to facilitate catalyst solubility.⁶⁹,⁷⁰ In more recent years researchers have realized the potential for utilizing ATRP to modify the surface chemistry of many materials. Colloidal silica,⁷¹ montmorillonite clay (MMT),⁷² gold films,⁷³–⁷⁵ and

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silicon wafers\textsuperscript{76–78} have been the most frequently studied substrates for surface initiated (SI)-ATRP, where the first is a spherical particle and the last three are high aspect ratio materials.

The production of polymer brushes advances many fields of science and technology through the production of stimuli responsive interfaces,\textsuperscript{79,80} improving colloidal stability,\textsuperscript{81} and the synthesis of nanocomposites.\textsuperscript{82–84} Polymer brushes are used to alter physical properties such as surface wetting and roughness, as well as to add chemical functionality. Stimuli responsive materials have been shown to respond to temperature, pH, and solvent quality among other environmental factors. In principle the physical and chemical properties of the brush layer may be precisely tuned through the monomer chemistry, polymer architecture, and molecular weight distribution, which in turn may only be controlled with a thorough understanding of the polymerization chemistry.

In the context of ATRP, the apparent polymerization rate, $R_p$, is governed by the equilibrium between dormant and active chains through reactions with the catalyst system. In SI-ATRP, the spatial proximity of the chain ends is significantly closer than in an analogous bulk polymerization, where the catalyst system is homogeneously distributed throughout the reaction media. These confinement effects can have a marked influence on $R_p$; to date, studies relating the kinetics of SI-ATRP to those in bulk media have thus far yielded conflicting and sometimes orthogonal conclusions.\textsuperscript{71,74,75,77,85}

Substrate geometry and graft density should play key roles in the reaction kinetics of SI-ATRP. Densely grafted spherical particles, for instance, initially feature an average chain-end-to-chain-end distance drastically smaller than in an “analogous” bulk system containing the same number of chain ends per volume. One may reasonably speculate that surface-confinement effects should be significant in this scenario. However, as the brush thickness increases, propagating radicals should spatially diverge from one another and recover bulk kinetic rates, since the distance between chain ends scales with the particle radius. Consistent with this notion, Fukada et al. studied the kinetics of an SI-ATRP from functionalized monodisperse silica particles ranging from 100 and 1500 nm in diameter with the inclusion of an unbound “sacrificial” initiator and found both species’ kinetic rates to be equivalent.\textsuperscript{71}

While SI-ATRP from spherical particles evidently assumes bulk kinetics, SI polymerization from planar substrates should be expected to exhibit confinement effects irrespective of the brush thickness. Literature studies of polymer brushes have focused largely on gold and silicon substrates that may be viewed as infinite planes; additionally, MMT clays approximate this geometry owing to their high aspect ratio.

Unfortunately, the number of investigations that explicitly account for surface confinement effects is limited. Early reports in the field indicate that $R_p$ decreases with time attributed to termination reactions that were not suppressed due to an inadequate supply of deactivating copper (II) (Cu\textsubscript{II}X\textsubscript{2}) species.\textsuperscript{73,86} This may be understood by considering that only microscopic quantities of Cu\textsubscript{II}X\textsubscript{2} are generated through the activation of dormant sites on
the surface, which are then diluted into the macroscopic quantity of monomer/solvent into which the substrate is immersed. These termination effects may be circumvented through the introduction of excess CuII X2 or unbound initiator at the onset of polymerization.  

There is some evidence that the living character of SI-ATRP reactions depends on the monomer; for example, a number of reports indicate linear monomer conversion vs. time in poly(methyl methacrylate) (PMMA), whereas \( R_p \) plateaus in polymerizations of poly(N-isopropylacrylamide), polyacrylamide, or poly(glycidyl methacrylate) (PGMA).

The role of graft density, \( \rho_g \), on \( R_p \) is unclear. Huck et al. investigated a series of poly(methyl methacrylate) (PMMA) brushes on gold substrates and found no dependence, although direct comparisons with bulk polymerization rates were not available. Genzer et al. studied graft-density effects using a gradient approach in poly(acrylamide) (PAA), and found that in the brush regime the thickness scaled as \( \rho_g^{1/3} \), suggesting all chains along the gradient were the same size.

Fukuda et al., however, reported that the brush height in a PMMA system from silica wafers at moderate graft density obeyed the \( \rho_g^{1/3} \) law, but at high graft density (\( \rho_g > 0.7 \text{ chains/nm}^2 \)) approached \( \rho_g^{1/2} \); again direct molecular weight measurements of the brush layer were not conducted and so here the possibility of a \( R_p \) dependence on \( \rho_g \) cannot be discounted.

In this publication we report a strong dependence of \( R_p \) on \( \rho_g \) in SI-ATRP of polystyrene (PS) from MMT. At high graft density, \( \approx 1 \text{ chain/nm}^2 \), we have observed an apparent polymerization rate nearly an order of magnitude larger than that of the analogous bulk system, which rapidly decays to the bulk kinetics as \( \rho_g \) is reduced. We hypothesize that this phenomenon is a consequence of local concentration heterogeneities that shift the equilibrium in favor of the active state, and present a simple phenomenological model that accounts for our data.

**Experimental Details.**

**Functionalization of Montmorillonite.** Montmorillonite clay (MMT) was generously supplied by Southern Clay Products Inc. Based on the ion exchange capacity and measurements of the specific surface area, MMT contains \( \approx 1 \text{ site/nm}^2 \). MMT was functionalized via ion exchange with a combination of 11’(N,N,N-trimethylammonium bromide)-undecyl-2-bromo-2-methyl propionate (BMP) and 11’(N,N,N-trimethylammonium bromide)-undecyl-2,2-dimethyl propionate (DMP).

The synthesis and ion exchange of BMP, an ATRP initiator, with MMT is a simple two step synthesis previously described elsewhere. It may be summarized as the addition of 2-bromoisobutyryl bromide to 11-bromo-1-undecanol and subsequent addition of trimethyl amine to produce a cationic alkyl chain with a terminal halide. The synthesis of an analogous inert compound, DMP, uses 2,2-dimethyl acetyl chloride in place
of 2-bromo-isobutyryl bromide. 2-bromo-isobutyryl bromide, ethyl-2-bromo-propionate (EBP), and 2,2-dimethyl acetyl chloride were purchased from the Aldrich Chemical Co. and used as received. BMP and DMP can then be mixed into desired molar ratios and ion exchanged to produce specified active graft-densities.

**Atom Transfer Radical Polymerization of styrene.** Benzyl bromide, copper(II) bromide (Cu_{II}Br_{2}), N,N,N',N',N',N'-pentamethyldiethylenetriamine (PMDETA), and tert-butyl acetyl chloride were purchased from the Aldrich Chemical Co. and used without further purification. Copper(I) bromide (Cu_{I}Br) was purchased from the Aldrich Chemical Co. and purified with acetic acid. Styrene was purchased from the Fisher Scientific Co., purified over basic alumina, and degassed prior to use. The unconfined ATRP of PS from either benzyl bromide (BBr) or ethyl-2-methyl-2-bromopropionate (EBP) followed the procedure described in numerous articles by Matyjaszewski et al. SI-ATRP of PS from MMT was similar to the work of others, with the addition of ultrasonication to enhance MMT tactoid dispersion. Monomer, initiator, Cu_{I}Br, Cu_{II}Br_{2}, and PMDETA were mixed under N_{2} in a round bottom flask with molar ratios of 1000:1:1:0.06:1.06 respectively. All polymerizations were conducted at 100 °C. Aliquots of PS brushes were cleaved from MMT over basic alumina for molecular weight analysis. Molecular weight distributions were determined by size exclusion chromatography (SEC).

**Results.**

To systematically control the polymer graft density, a series of MMT compounds were prepared via complete ion exchange with mixtures consisting of a mole fraction σ of BMP (ATRP-active) and (1 − σ) DMP (dormant) surfactants, depicted schematically in Figure 3.1. Figure 3.2 demonstrates that this process occurs in a two stage addition; the first after 48 hours and the second by 96 hours and that longer mixing times do not enhance XRD d-spacing further. Figure 3.3 displays all XRD d-spacings and corresponding TGA data which shows the fraction of volatiles reaches a maximum after 96 hours of ion exchange. Given the chemical similarity of BMP and DMP, we assert that the composition of the surface-bound ligands is essentially identical to that in the bulk medium, and thus the graft density of MMT-PS materials may be estimated as $\rho_g = \sigma \rho_{g,max}$. The maximal graft density $\rho_{g,max}$ is ≈ 1 nm$^{-2}$ as calculated from the estimated specific surface area of MMT, 600 m$^2$ g$^{-1}$, and ion exchange capacity of 0.92 meq g$^{-1}$.

ATRP reactions were conducted using unbound EBP or BBr as the initiator using molar ratios of monomer, initiator, Cu_{I}Br, Cu_{II}Br_{2}, and PMDETA equal to 1000:1:1:0.06:1.06. In reactions where molecular weights in excess of 100 kDa were expected, monomer to initiator molar ratios were raised to 1500:1 to maintain low polydispersity. Significant improvement in reproducibility was realized by premixing Cu_{I}Br with Cu_{II}Br_{2} in a 16:1 molar ratio prior to use. The addition of Cu_{II}Br_{2} also served to slow propagation so that growth of chains
Figure 3.1 Route to MMT with precisely defined graft densities were prepared via complete ion exchange with mixtures comprised of a fraction $\sigma$ of ATRP active alkyl-ammonium surfactants.
Figure 3.2 X-ray diffraction spectrum demonstrating the two stage addition of initiator to MMT during ion exchange. As is evidenced by comparison of 96 hr and 168 hr traces, no further ion exchange occurs after 96 hrs.

Figure 3.3 X-ray diffraction (○) and thermogravimetric analysis (◇) data demonstrating that by 96 hours of stirring maximum ion exchange has occurred.
Table 3.1  Results of ATRP polymerization of bulk and confined polystyrene systems.

<table>
<thead>
<tr>
<th>Initiator</th>
<th>Time, hr</th>
<th>M&lt;sub&gt;n&lt;/sub&gt;, kDa</th>
<th>PDI</th>
<th>Conv&lt;sub&gt;theo&lt;/sub&gt;</th>
<th>Conv&lt;sub&gt;exp&lt;/sub&gt;</th>
<th>dM&lt;sub&gt;n&lt;/sub&gt;/d&lt;sub&gt;t&lt;/sub&gt;, kDa/hr</th>
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<tbody>
<tr>
<td>BBr</td>
<td>6</td>
<td>23.3</td>
<td>1.11</td>
<td>0.224</td>
<td>NA</td>
<td>NA</td>
</tr>
<tr>
<td>BBr</td>
<td>12</td>
<td>35.8</td>
<td>1.12</td>
<td>0.345</td>
<td>NA</td>
<td>1.89</td>
</tr>
<tr>
<td>BBr</td>
<td>24</td>
<td>59.1</td>
<td>1.24</td>
<td>0.569</td>
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<td>NA</td>
</tr>
<tr>
<td>EBP</td>
<td>6</td>
<td>22.7</td>
<td>1.05</td>
<td>0.219</td>
<td>0.264</td>
<td>NA</td>
</tr>
<tr>
<td>EBP</td>
<td>12</td>
<td>43.4</td>
<td>1.07</td>
<td>0.416</td>
<td>0.131</td>
<td>1.89</td>
</tr>
<tr>
<td>EBP</td>
<td>24</td>
<td>56.4</td>
<td>1.07</td>
<td>0.542</td>
<td>0.542</td>
<td>NA</td>
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<tr>
<td>33%-MMT</td>
<td>6</td>
<td>21.5</td>
<td>1.20</td>
<td>0.069</td>
<td>0.071</td>
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<td>33%-MMT</td>
<td>12</td>
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<td>1.15</td>
<td>0.114</td>
<td>0.101</td>
<td>2.54</td>
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<td>33%-MMT</td>
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<td>75.9</td>
<td>1.28</td>
<td>0.243</td>
<td>0.178</td>
<td>NA</td>
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<td>67%-MMT</td>
<td>6</td>
<td>30.4</td>
<td>1.12</td>
<td>0.195</td>
<td>0.146</td>
<td>NA</td>
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<td>67%-MMT</td>
<td>12</td>
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<td>1.09</td>
<td>0.330</td>
<td>0.192</td>
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<td>24</td>
<td>101.5</td>
<td>1.16</td>
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<td>100%-MMT</td>
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<td>12.9</td>
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<td>100%-MMT</td>
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<td>1.12</td>
<td>0.927</td>
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<td>NA</td>
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<td>6</td>
<td>39.9/80.8</td>
<td>1.08/1.01</td>
<td>NA</td>
<td>NA</td>
<td>—</td>
</tr>
</tbody>
</table>

within the interior of the clay galleries was not diffusion limited.

A series of control experiments from unbound initiators employed either BBr or EBP as the initiator system. BBr was chosen for its chemical similarity with the monomer, whereas EBP was chosen to provide a closer comparison to the surface-tethered initiator species (BMP). Included in the control experiments were a series of unbound initiator syntheses in the presence of pristine MMT; no discernable affect was measured via SEC as a result of the unfunctionalized MMT additive. The SEC results obtained for BBr and EBP were comparable in all experiments, although EBP exhibited a modest improvement in overall PDI. The decrease in PDI can be attributed to the faster initiation of the EBP species, resulting in a more uniform initiation of polymer chains. A representative sample of bulk and confined SEC results are shown in Table 3.1.

To assess the initiator efficiency of the MMT-initiated polymerizations, we compared the monomer conversion as measured by total yield versus the theoretical conversion calculated from M<sub>n</sub> assuming 100% initiator activity, 3.1. From these data we conclude that we reproducibly achieve at least 95% initiator efficiency in all systems.
Figure 3.4 Bulk and confined polystyrene kinetic rates with active initiator percentages shown as 100% MMT (△), 67% MMT (+), 33% MMT (⋄), and bulk (◦). Samples are shown with a linear fit to guide the eye.

Figure 3.4 compares the rates of polymerization for bulk PS and MMT-PS with three different graft-densities, with the MMT samples listed in order of the mole percent active initiator (BMP). Immediately evident from inspection of Figure 3.4 is that \( R_p \) for PS grafted from 100%-MMT is nearly an order of magnitude greater than unbound PS produced under analogous conditions. The SI-ATRP of styrene from 33%-MMT and 67%-MMT are also accelerated relative to bulk kinetics, but lie between bulk and 100% MMT values respectively. Clearly, MMT graft-density has a profound influence on the apparent propagation rate. The non-zero intercepts of the kinetic data appearing in Figure 3.4 is also noteworthy and likely indicates a short induction period during which uncontrolled polymerization proceeded prior to the development of the persistent radical effect.

All polymerizations exhibited living character with linear monomer consumption versus time for all graft-densities (Figure 3.5). Successful reinitiation of MMT-PS brushes gave further evidence that termination reactions were effectively suppressed. In order to avoid clay aggregation, sufficient solvent needed to be present in the system to disperse MMT platelets. Thus in our system where monomer was used in excess as a solvent, a constant reaction volume was maintained by preparing molar ratios of monomer to combined active and inactive initiator concentrations. The resulting monomer to active initiator molar ratios were bulk 1000:1, 100% MMT 1000:1, 67% MMT 1500:1, and 33% MMT 3000:1. Figure 3.6 clearly illustrates the expected differences in
Figure 3.5  Bulk and confined polystyrene kinetic rates with active initiator percentages shown as 100% MMT (△), 67% MMT (+), 33% MMT (⋄), and bulk (○). Samples are shown with a linear fit to guide the eye.
Figure 3.6  Bulk and confined polystyrene kinetic rates with active initiator percentages shown as 100%MMT (△), 67%MMT (□), 33%MMT (⋄), and bulk (○). Open and closed symbols correspond to PDI and \( M_n \) respectively. Samples are shown with a linear fit to guide the eye.

\( M_n \) versus conversion and results agree with theoretical conversion limits for all graft-densities. Figure 3.6 also presents the PDI’s for all graft-densities demonstrating good molecular weight control up to 80% conversion; a loss of PDI control of ATRP reactions at high conversions is well documented in the literature.

To assess the potential influence of the MMT particle itself on the polymerization kinetics, we conducted experiments using a mixture of tethered 100%-MMT initiator and free EBP. A representative SEC trace appears in Figure 3.7, which shows the results from a polymerization using a 10:1 molar ratio of 100%-MMT:EBP at 100 °C for 6 hours. The distribution is bimodal and was deconvolved via least squares regression to a bimodal Schulz-Zimm distribution yielding \( M_n,\text{MMT} = 80.8 \) kDa and \( M_n,\text{EBP} = 39.9 \) kDa. The mass fraction of the MMT distribution in the SEC sample is evidently \( \approx 23\% \), which reflects the repartitioning of the free vs. tethered chains during the removal of grafted chains in the sample preparation.

To assess the degree of exfoliation the MMT-PS materials, representative samples were sectioned using a diamond knife at cryogenic temperatures (−100°C) with a Leica UCT ultramicrotome. Light diffraction indicated an average section thickness of \( \approx 100 \) nm. Polymer sections were transferred to copper grids and exposed to RuO\(_4\) vapor. MMT makes an excellent barrier to RuO\(_4\) deposition, resulting in strong contrast of MMT sheets. As can be seen in Figure 3.8 discrete MMT platelets are easily visible and display complete exfoliation.
Figure 3.7 A representative SEC trace showing the molecular weight distribution from an ATRP at 100 °C initiated from a 10:1 molar ratio of 100%-MMT and EBP.

Figure 3.8 Electron micrograph of single montmorillonite silicate particles with grafted polystyrene chains shown to be very well exfoliated. Contrast results from RuO$_4$ deposition, where clay particles pose an excellent permeation barrier.
Discussion.

In conventional uncontrolled free radical polymerization, auto-acceleration is frequently observed at high molecular weight due to viscosity effects, i.e., the Trommsdorff effect. The Trommsdorff effect attributes faster kinetics at high conversion due to the elevated viscosity, which disproportionately hinders termination reactions by limiting the mobility of large molecules, favoring propagation. Here, we believe that kinetic enhancement occurs not due to a reduction in termination reactions, which are already strongly suppressed in ATRP, but rather due to local concentration heterogeneities that shift the equilibrium governing how frequently a chain is in its active state. The essence of our argument lies in the fact that all of the active sites (chain ends) in this surface-initiated system are constrained to reside in a volume significantly smaller than the macroscopic reaction volume. The consequences of this, directly related to the graft density, are twofold:

- The local viscosity in the vicinity of reactive front will be significantly elevated compared to the macroscopic average.

- The RMS separation between active sites will be dramatically lower than in a corresponding homogeneous ATRP.

We assert that the result of these considerations is that within the growing viscous front, there is a repartitioning of the activating and deactivating catalyst species that leads directly to an elevation of the probability that any given chain end will be in its active state.

ATRP suppresses terminations reactions via the so-called “persistent radical effect”, an equilibrium between dormant polymer chains and propagating free radicals, where the dormant state is typically favored by many orders of magnitude depending on ligand quality:

\[
P X + Cu_{II}X \rightleftharpoons K \cdot + Cu_{II}X_2
\]

\[
P_{i,j} + [M] \xrightarrow{kp} P_{i+1}
\]

It is useful at this point to consider a kinetic model in terms of how fast, on average, a single chain is growing. This is simply the total polymerization rate times the monomer molecular mass \( M_0 \), divided by the total number of chains in the system per volume, which can be approximated as \([PX]\). Eqn 3.3 defines the rate of polymerization of each chain in terms of the rate constant of propagation \( kp \), monomer concentration \([M]\), and
the probability that any chain is in its active state \( \frac{[P\cdot]}{[PX]} \).

\[
\frac{dM_n}{dt} = \frac{M_0 R_p}{[PX]} = k_p[M]\frac{[P\cdot]}{[PX]}
\]

(3.3)

In a truly pseudo-first order system, such as anionic polymerization or even uncontrolled free radical polymerization, the local concentration of active centers would assume a constant value throughout the system, irrespective of any local concentration inhomogeneities. The activation probability \( \frac{[P\cdot]}{[PX]} \) in ATRP is governed by the activation-deactivation equilibrium and may be expressed as:

\[
\frac{[P\cdot]}{[PX]} = K \frac{[CuI]X}{[CuII]X_2}
\]

(3.4)

where \( K \) is the equilibrium constant between active and dormant chains and has previously been reported for styrene, using 4,4′-di-(5-nonyl)-2,2′-bipyridine as a ligand, to favor the dormant state by seven orders of magnitude.\(^{92}\) This model for SI-ATRP equilibrium has been similarly derived by Xiao and Wirth while studying SI-ATRP of acrylamide from silica.\(^{76}\) In a bulk ATRP, the equilibrium value of \( \frac{[CuI]X}{[CuII]X_2} \) is homogeneous throughout the system, and thus the pseudo-first order character of the polymerization is preserved.

In the present system, however, all of the chain ends are confined within a “viscous front” of small but finite thickness, separated from the bulk media by a region enriched in polymer chains. The local density of initiation sites is zero except within this “growing viscous front”, depicted schematically in Figure 3.9.

Within the “growing viscous front”, the local concentration must be related to the mean interparticle spacing, which is a function of initiator density and scales as \( (\sigma \rho_{g,max})^{-1/2} \). The average volume that may be assigned to each initiation site must then scale as \( (\sigma \rho_{g,max})^{-3/2} \), and accordingly the local chain end concentration scales as \( (\sigma \rho_{g,max})^{3/2} \). The diffusion of activating and deactivating catalyst through the growing front will be suppressed due to its enrichment with viscous polymer. Now consider an isolated propagating chain within this region: any encounter with a CuIIX2 complex that results in a deactivation reaction will deplete the local environment of a single CuIIX2 while enriching it with CuI X (Figure 3.9a). The action of diffusion will work rapidly to restore this local disruption to a homogeneous state. However, over sufficiently small time and length scales there will be a finite elevation of CuI X balanced by the depletion of CuII X2. Concretely, at some position \( r \) and time \( t \) from its formation, the probability of finding an “extra” CuI X scales as \( t^{-3/2} e^{-r^2/2} \). In a similar vein, the dynamics of replenishment of CuII X2 will be governed by its diffusion from the bulk media. For the activation/deactivation kinetics to be influenced, there must be other chain ends present at small enough values of \( |r| \) for this probability to be non-vanishing. In Figure 3.9, the position dependence of the probability with \( |r| \) is illustrated with a color gradient centered around the activated chain end. Thus in a homogeneous ATRP, where the distance between
Figure 3.9  Schematic of the surface initiated ATRP of polymer from MMT with the dark gray box portraying the “growing viscous front” of small but finite thickness. **Legend:** Halide terminated chains (●), non-functional initiators (×), and an active chain (+). In this region there is a locally elevated concentration of chain ends, which observe local concentration heterogeneities resulting from the conversion of (a) Cu_{II} (.annotated) to Cu_{I} (○) or (b) Cu_{I} to Cu_{II}. The bulk catalyst concentrations then work to re-establish equilibrium, where the more abundant Cu_{I} has a much stronger driving force and lower steric resistance to mass transfer.

chain ends is large, such mass transport effects are negligible. In the confined ATRP experiments considered in this study, however, this distance is on the order of nanometers and the effect is evidently substantial.

Conversely, the activation of a dormant chain would locally enrich the environment with Cu_{II}X\textsubscript{2} while depleting Cu_{I}X (Figure 3.9b). Since at equilibrium the rate of activation and deactivation must balance, it is important to consider why these enrichment/depletion events do not cancel each other. The probability of finding the Cu_{II}X\textsubscript{2} formed in an activation event again scales as \( t^{-3/2} e^{-r^2/\tau} \), and thus it should encounter other chain ends in the local vicinity. However, since these other chain ends are likely to be dormant, this local excess of Cu_{II}X\textsubscript{2} should have no effect. Additionally, the replenishment of the locally depleted Cu_{I}X will be a significantly faster process due to its higher bulk concentration relative to Cu_{II}X\textsubscript{2}. Moreover, the diffusivity of Cu_{I}X should be greater than that of Cu_{II}X\textsubscript{2} owing to its comparatively smaller size. Thus collectively, the surface initiated system experiences a net local elevation in \([Cu_{I}X]\) at the expense of \([Cu_{II}X_{2}]\) due to mass transport effects, induced by the elevated local viscosity and small chain-end-to-chain-end distances.

Similar arguments suggest that the local monomer concentration in the “reaction plane” may be depleted in the neighborhood of an active chain, owing to the reduced monomer diffusivity and elevated local viscosity. This depletion should again scale with the local chain concentration and, in principle, should oppose the acceleration
due to the shift in the local catalyst concentration. In our experiments, however, we do not observe monomer exclusion effects since the monomer plays a dual role as solvent. Thus, any depletion event induces a strong concentration gradient such that the time scale for monomer diffusion to the active site is still much shorter than that of the propagation reaction. Accordingly, we suspect that local monomer depletion effects may become observable in dilute monomer/solvent systems.

Accordingly, the acceleration of $\frac{dM_n}{dt}$ relative to an ATRP from free initiators may be expressed as

$$\frac{dM_n}{dt}_{\text{MMT}} = \frac{\langle [Cu_{II}X_2] \rangle}{\langle [Cu_{II}X] \rangle} \frac{\langle [Cu_{I}X] \rangle + \lambda (\sigma \rho_{g,max})^{3/2}}{\langle [Cu_{II}X_2] \rangle - \lambda (\sigma \rho_{g,max})^{3/2}}$$

Free initiator kinetics are recovered as $\sigma \rightarrow 0$. As shown in Figure 3.10, this single-parameter model reproduces our kinetic data with $\lambda = 0.864$. 

Figure 3.10  Propagation rates, relative to bulk ATRP of PS, of PS-graft-MMT produced by SI-ATRP. The data are fit with a single parameter $\lambda$ using the relationship of Eqn 3.5. For a 16.7:1 molar ratio of CuBr : CuBr$_2$, $\lambda = 0.864$. 

In consideration of these enrichment/depletion arguments, we may construct a simple phenomenological model that encapsulates the effects of this coupled network of mass transport and reaction events. In the “reaction plane”, chain ends locally experience net catalyst concentrations of $[Cu_I X] = \langle [Cu_I X] \rangle + \delta$ and $[Cu_{II}X_2] = \langle [Cu_{II}X_2] \rangle - \delta$, where $\langle \ldots \rangle$ denotes the homogeneous equilibrium concentration. The local deviation $\delta$ should be proportional to the local chain concentration, $\propto (\sigma \rho_{g,max})^{3/2}$, and will be influenced by a host of factors including the activation/deactivation rate constants and the local diffusivity. We introduce a dimensionless lumped “effectiveness” parameter $\lambda$ that incorporates these considerations, such that $\delta = \lambda (\sigma \rho_{g,max})^{3/2}$. Accordingly, the acceleration of $\frac{dM_n}{dt}$ relative to an ATRP from free initiators may be expressed as
Our results support the plausibility of this simple enrichment/delpletion model and demonstrate that the kinetics of surface-initiated ATRP under certain conditions, i.e., in the strong brush regime, depart significantly from an analogous ATRP from free initiators. These considerations become important in the context of molecular design. Moreover, caution should be exercised in experiments in which both tethered and free initiators are employed, since inferences of the molecular weight distribution of the brush layer derived from that of the free polymer may be flawed.

**Summary.**

The kinetics of polystyrene brush polymerization from MMT clay by SI-ATRP was investigated as a function of graft density and compared to bulk ATRP of styrene under analogous conditions. A seven-fold increase in PS growth rate was observed, relative to bulk kinetics, at the highest graft density, $\approx 1\text{chain nm}^{-2}$. Bulk kinetics were rapidly recovered as the graft density was reduced. We hypothesized that local concentration heterogeneities shift the ATRP equilibrium in favor of the active state, and developed a single-parameter kinetic model based on this hypothesis that was able to account for our kinetic data. These findings present an important consideration relevant to the design of precisely defined molecular architectures from surfaces using surface-initiated ATRP.

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**Bibliography**


CHAPTER 4 HIERARCHICALLY ORDERED MONTMORILLONITE BLOCK COPOLYMER BRUSHES


Ross E. Behling$^1,2$, Lynn M. Wolf$^3$, and Eric W. Cochran$^1,3$

**Introduction**

Block copolymers (BCPs) have been known for decades to offer an easily tunable method for producing self-assembled structures with length scales on the order of nanometers to tens of nanometers. More recently, the community has shown significant interest in exploiting the properties of BCP self-assembly to tailor the spatial and orientational distribution of nanoscale filler materials, with targeted applications ranging from high-density storage devices, to organic electronics, to optical devices, and to separation devices/catalytic membranes.

From a commercial perspective, currently the most important class of polymer nanocomposites features layered silicates—typically montmorillonite (MMT)—suspended in a matrix of homopolymer. This combination of materials yields dramatically enhanced mechanical, barrier, and thermal properties compared to the neat homopolymer. The degree to which these enhancements may be realized depends strongly on the morphology of clay inclusions and the clay-polymer interactions at the interphase. Many studies have treated the problem of optimizing the dispersion of the clays within the polymer matrix. The degree of dispersion, and the spatial/orientational distribution of the particles in these systems is largely governed by the strong shear fields imposed on the system during processing. Typically this yields a distribution of particle aggregates and exfoliated particles, isotropically distributed throughout the system. The most successful examples to-date of nanoparticle

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inclusion have nanoparticles functionalized with homopolymer and then dispersed in a BCP host matrix. Bockstaller et al. reviewed the successful inclusion of gold, silica, and montmorillonite (MMT) particles using this technique, providing an excellent overview of how BCPs can be employed to control nanoparticle dispersion and orientation. \textsuperscript{93,112} Significantly, physical blending with nanoclays has not yet yielded any greater degree of control of the particle distribution than traditional homopolymer nanocomposites (NCPs). \textsuperscript{113} This is evidently due to the difficulty of overcoming the strong particle-particle interactions and the disparity between the particle size, $≈ 100–300 \times 1$ nm, and typical BCP domain sizes, $≈ 10–50$ nm.

A relatively unexplored alternative approach to nanocomposite formulation is to dispense with the matrix altogether by integrating it directly with the filler particle. For example, Gianellis and others have used this idea leading to the discovery of nanoparticle fluids by the attachment of oligomers to colloidal particles. \textsuperscript{114} We have discovered that aggregation and particle distribution limitations can be circumvented by directly grafting BCPs to the silicate surface, effectively encapsulating the filler particle in a dense brush layer. Krishnamoorti and Vaia have speculated that “...block copolymers and other structured polymers onto the surfaces of nanoparticles can alter significantly the natural topologies adopted by those materials and therefore lead to interesting characteristics”. \textsuperscript{115} We find that this is indeed the case, with new emergent physics from the synergism generated by the combination of polymer brushes, nanoparticles, and block copolymer self-assembly. In this system the
relationship between matrix and filler is uniquely dictated by molecular design, and the resultant system represents a distinct class of self-organizing materials. The thermodynamics of these MMT block copolymer brush (MMB) systems reflect the consequences of microphase separation constrained by the brush extension and the connectivity of the chain ends to the semiflexible MMT substrate. The morphology of these materials is intrinsically hierarchical, with characteristic length scales prescribed by the supporting MMT particle, Figure 4.1a; the polymer chain dimensions, Figure 4.1b; and the chain stretching/interfacial curvature imposed by microphase separation, Figure 4.1c. In this Communication we present and interpret examples of the hierarchically ordered structures that form as a consequence of these multiple length scales.

Experimental Section

Montmorillonite clay was generously supplied by Southern Clay Products Inc. Based on the ion exchange capacity, 92 mequiv\( \cdot 100g^{-1}\), and measurements of the specific surface area,\(^{116}\) MMT contains \(\approx 1\ \text{site}\ \text{nm}^{-2}\). MMT was functionalized with a bromine terminated alkylammonium surfactant (Fig. 4.1a), and was subsequently polymerized via surface-initiated ATRP (SI-ATRP) to yield polystyrene (Fig. 4.1b) and poly(styrene-block-t-butyl acrylate) brushes (Fig. 4.1c) as described in detail elsewhere.\(^{117}\) We refer to MMT-graft-poly(styrene) homopolymer brushes as MMT-X, and MMT-graft-poly(styrene-block-t-butyl acrylate) block copolymer brushes as MBB-X-Y; X refers to polystyrene (PS) and Y to poly(tert-butyl acrylate) (PtBA) number-average molecular weight (\(\mathcal{M}_n\)) in kDa. MBBs were annealed at 150\(^\circ\)C in vacuo for over 96 hours prior to being steady shear processed at 1\(^{-s}\) and 160-200\(^\circ\)C for 25 minutes on a TA Instruments ARES-LS1 strain controlled rheometer in the parallel plate geometry under N\(_2\). Ultrathin (\(\approx\)100 nm) sections were obtained at -100\(^\circ\)C using a Leica Ultramicrotome Ultracut 125UCT with a Leica EM FCS cryo-stage. RuO\(_4\)-stained sections were examined on a Tecnai G\(^2\) F20 scanning / transmission electron microscope at a high tension voltage of 200 kV. Length scale information was extracted from TEM images by a combination of line measurements taken using the Gatan DigitalMicrograph\textsuperscript{TM} software, and analysis of the azimuthal average of the discrete Fourier transform.

Results and Discussion

Figure 4.2: Part I shows a series of TEM micrographs of ascending corona block length and fixed core block length, where the PtBA block was stained to enhance contrast. Figure 4.2a demonstrates partial phase separation as evidenced by 25–100 nm ellipsoidal “pockets” locally enriched in PS, distributed in a homogeneous PS/PtBA background. The mean distance between “pockets”, as determined by discrete Fourier transform (DFT) analysis, is 31nm. MBB-44-64 (Fig. 4.2b) features a corona block nearly twice the size of MBB-44-36 (Fig. 4.2a) and
Figure 4.2 (a–e) Representative TEM micrographs of multiple NCP morphologies (a) MBB-44-36, (b) MBB-44-64, (c) MBB-44-90, (d) MBB-70-30, and (e) MBB-74-73 with the associated schematics of the unit structures. Scale bars for micrographs and schematics are respectively identical; tBA domain stained with RuO$_4$ for contrast. (f–j) Schematic illustrations of the polymer/clay configuration suggested by the micrographs in the top row. (Part I) Scale bar is 100 nm. (Part II) Scale bar is 200 nm.
exhibits a fully ordered morphology. Here the PS domains form an interpenetrating network characterized by domains $\approx 28\text{nm}$ in diameter, separated on average by $51\text{nm}$ with irregular connectivity. Further increasing the corona block size to nearly triple that of MBB-44-36, MBB-44-90 (Fig. 4.2c) displays modulated cylindrical domains, $90\text{ nm}$ in diameter, of PS and PtBa interpenetrating in a manner reminiscent of worm-like micelles that are observed in dilute solutions of amphiphilic BCPs. Figure 4.2:Part II depicts structures nearly twice as large as those seen in Figure 4.2:Part I. The composition of MBB-70-30 is roughly that of MBB-44-90 with reverse majority and minority components and comparable overall molecular weights; however, the morphology of MBB-70-30 bears no resemblance to that of MBB-44-90. MBB-70-30 exhibits single and concentric tori of PtBA $25–35\text{nm}$ thick that average $200\text{nm}$ ($450\text{nm}$) in the axial direction and $100\text{nm}$ ($200\text{ nm}$) in the cross-axial direction. The nearly symmetric composition of MBB-74-73 is comprised of oblate PtBA ellipsoids averaging $250\text{nm}$ in the major axis and $150\text{nm}$ in the minor axis with $\approx 20\%$ deviation in both axes.

To facilitate our interpretation of the unique progression of the morphology in these specimens we first make a number of observations. One, the segregation strength $\chi N$, where $\chi$ is the Flory interaction parameter and $N$ is the polymerization index, required to induce complete microphase separation in MBBs is significantly larger than that in analogous untethered AB diblocks. Two, the characteristic domain sizes in the structures spans an incredibly wide range — from $15\text{nm}$ to $500\text{nm}$ — whereas the brush $\overline{M}_n$ varies over a much smaller range, $80–150\text{ kDa}$. This is in stark contrast to AB diblocks of comparable molar mass, where the feature sizes would be $<50\text{nm}$. Three, the morphology is strongly dependent on the absolute brush polymerization index $N$, again in contrast to AB diblocks where the entire phase space is mapped by the chain composition $f$ and $\chi N$. Four, the block sequence plays a key role in the evolution of the morphology.

Self assembly in MBBs is intrinsically hierarchical due to the mixture of length scales fundamentally present in the system: the supporting MMT particle is discoidal with diameter on the order of $10^2\text{ nm}$; polymers grafted to the MMT surface are constrained to have an interchain spacing on the order of $10^0\text{ nm}$ at the graft site; and the length scale of polymer microphase separation is dictated by the RMS end-to-end distance $h$. $h$ is maximally bounded by the contour length $\ell = 0.1258(2n − 1)\text{nm}$, where $n$ is the number of ethylene repeat units ($10^1 − 10^2$ in our system). The degree of chain stretching in these materials is significant, and can be estimated through a qualitative calculation as depicted in Figure 4.3a: Consider an ideal MMT particle to be a disk with radius $R = 200\text{nm}$, grafted with $50\text{ kDa}$ PS chains at a graft density $\rho_g = 1\text{ chains nm}^{-2}$. The volume of PS attached to this particle is then $V_{PS} = \pi R^2 \rho_g V_{PS}$, where $V_{PS} \approx 90\text{ nm}^3\text{ chain}^{-1}$ is the volume of a $50\text{ kDa}$ PS chain. Now suppose that $r$ represents the maximum distance of PS from the MMT particle. Equating $V_{PS}$ to the volume defined by the
Figure 4.3  Schematic illustrations that describe the physics of MBB self-assembly (a) Chains are strongly stretched since they are constrained to reside within the shaded volume, a distance $r$ from the supporting MMT particle. At sufficient segregation strength this leads to either (b) intraparticle self-assembly or (c) interparticle self-assembly. This mode of phase separation is characterized by large domain spacing. (d) Interfacial curvature for interparticle self-assembly must be accompanied by bending of the supporting particle, requiring the symmetry of the mirror plane to be broken.
geometrical object (Fig. 4.3a) yields:

\[ \rho_g V_{PS} = \frac{\pi r^2}{R} + 2r \]  

(4.1)

with \( r = 30.4\text{nm} \), \( \approx 25\% \) of the contour length, in strong qualitative agreement with the mean interparticle spacing of \( \approx 50\text{nm} \) in MMT-50 (Fig. 4.1b).

Self assembly in MBBs is intrinsically hierarchical due to the mixture of length scales fundamentally present in the system. For microphase separation to occur in MBBs, the system must accommodate additional chain stretching to form discrete domains. Consequently MBB materials should require a significantly higher degree of segregation strength than the analogous untethered BCPs. MBB-44-36 (Fig. 4.2a) is only partially ordered in spite of the 80kDa size of the polymer brush and is consistent with this idea. Clearly, the weak ordering exhibited by MBB-44-36 demonstrates that this material is only weakly segregated even though \( \chi_N \gg 10.5 \). As depicted schematically in Figure 4.2f, PS-rich domains are stable only near the PS-MMT interface; beyond this region, the additional entropic cost of phase separation is not sufficient to compensate for the enthalpic cost of remaining homogeneous. Consequently, the PS domain size in MBB-44-36 is thus governed by the MMT particle size whereas the interdomain spacing is dictated by the polymer molecular weight.

Complete phase separation begins to occur in MBB-44-64 where the PtBA block is sufficiently large to form distinct domains with diffuse PS/PtBA interfaces curved towards the minority PS regions as depicted in Figure 4.2g. This system evidently exhibits intraparticle phase separation, where the PS domains form within the plane defined by the MMT particle. Here phase separation is essentially constrained to occur within this plane, within a single MBB pseudo-particle; interactions with adjacent particles only occur to fill the overall void space. The domain spacing, 46nm, is consistent with traditional diblock copolymers of comparable size.

In MBB-44-90 (Fig. 4.2c), however, the larger PtBA block completely fills the space between opposing MBB surfaces, and we observe a fundamental difference in the way the system self-organizes. As a consequence of the formation of contiguous polymer microdomains, the supporting MMT particles are now also ordered, in contrast to MBB-44-64, and the dominant structural unit arises from interparticle assembly as illustrated in Figure 4.3c. In this system the characteristic length scale of MMT emerges directly in the microdomain structure as evidenced by the dramatic increase of the interdomain spacing of 46nm in MBB-44-64 to 90nm in MBB-44-90; that is, a 24% increase in the molecular weight induces a 96% increase in the domain spacing.

Similarly, increasing the size of the interior PS block also has a remarkable influence on the morphology. MBB-74-73 (Fig. 4.2e), is similar in composition to MBB-44-36 and nearly double the molecular weight. The structure of MBB-74-73 again reflects interparticle self-assembly with an average domain size of 250×150nm.
Here intraparticle phase separation is not possible, the only mechanism by which the system can reduce unfavorable PS/PtBA interactions is through coherent ordering of the entire MBB particle, which is evidently responsible for the uncharacteristically large domain size. We observe the interparticle assembly mechanism in MBB-70-30 as well, where the corona PtBA block is the minority component. Here, an untethered diblock of identical composition would be expected to form PtBA cylinders, with the interfaces curved towards the minority phase. However, in MBB-70-30, the large core PS block is unable to provide this interfacial curvature locally. Rather, the concentrically arranged tori that appear in Figure 4.2d result from the long-range ordering of MBB particles and provide an alternative route to forming the interfacial curvature. The major axis diameter of these tori are as large as 500nm, which is larger than that expressed by MBB-74-73 although the molecular weight is appreciably smaller. The tori thickness is ≈ 30nm consistently throughout the specimen, expressing the length scale associated with the PtBA block size.

It is interesting to compare MBB-70-30 with MBB-44-90, which are nearly complementary in chemical composition and of similar molar mass, yet the differences in their morphology serves to illustrate the influence of the location of the minority block. In MBB-44-90 the minority block is the core, and its encapsulation with the majority corona block naturally promotes the formation of interfaces curved towards the PS domains and allows the formation of a nearly periodic structure. In contrast, there is no manner in which MBB-70-30 can tile space while forming interfaces curved towards the minority corona block.

The formation of curved interfaces through interparticle self-assembly requires the bending of the supporting MMT particles we depict schematically in Figure 4.3d, breaking the symmetry of the internal mirror plane defined by the MMT support. In neat BCPs, the interfacial curvature is the result of a precise balance between the enthalpic interactions, $F^{\text{int}}$, with the elastic energy contributed by chain stretching, $F^{\text{chain}}$. In MBBs, the introduction of the high-aspect ratio interface imposed by the clay particles should dampen the degree of curvature possible with the introduction of the clay bending energy $F^{\text{clay}}$. One may reasonably speculate, as in other quenched bilateral brush systems, that this bending energy should dominate the system and, therefore, only lamellar configurations such as that appearing Figure 4.1c would be stable. MBB-44-90, MBB-70-30, and MBB-74-73 suggest otherwise by exhibiting interfaces with radii of curvature on the order of the estimated persistence length of MMT, $\approx 140 \text{ nm}$. In homopolymer/MMT mixtures, Drummy et al. have observed clay platelets bent even more severely with the radius of curvature as small as 15nm. Clearly, while the bending stiffness of MMT is significant, $\approx 1.25 \text{ N/m}$, the energy increase with clay deformation is matched to that of the energy decrease to the relaxation of the polymer conformations accessible through bending. The precise mechanism for this spontaneous bending is thus likely a subtle consequence of a delicate energetic balance; we are currently developing a self-consistent field theoretic model of this system to quantitatively address this issue.
Conclusions

The interplay of $F_{\text{chain}}$, $F_{\text{int}}$, and $F_{\text{clay}}$ lead to the rich polymer physics we observe in MBBs. MBBs represent a new class of materials capable of self-assembly into structures with length scales previously unaccessible to block copolymers of comparable molecular weights, and unparalleled control of the spatial and orientational ordering of the filler particles. The physics of the MBB system are governed by the canonical parameters $f$ and $\chi N$ that pertain to AB diblocks; moreover, $N$ and the block sequence play critical roles in the delicate balance of energies that lead to the mesophases we have observed. Important technological implications of these materials arise from two important features of MBBs: the length scales associated with ordered phases easily approach the microscopic range; and we expect that the extremely strong degree of chain stretching may have a profound influence on physical properties.

These materials may lead to a new route to non-linear optical materials owing to their ability to reach domain sizes on the order of visible light. For example, BCP domain spacings of 100–200 nm were previously only accessible by synthesizing lamellar diblocks ($f = 0.50$) of $M_{n,\text{total}} > 1000$ kDa or using relatively smaller blocks of 200 kDa each and swelling the domains with homopolymer. By comparison, MBBs can reach domain sizes as large as 150 nm at only 100 kDa. Finally, the high level of chain stretching in MBBs may offer a route to materials with elevated glass transition temperature or unprecedented degrees of crystallinity.

Supporting Information Available: Further experimental details and additional microscopy data. This material is available free of charge via the Internet at http://pubs.acs.org.

Acknowledgements

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[94] Antonino Salvatore Arico, Peter Bruce, Bruno Scrosati, Jean-Marie Tarascon, and Walter van Schalkwijk.


CHAPTER 5  MORPHOLOGICAL AND MECHANICAL BEHAVIOR OF MONTMORILLONITE GRAFTED BLOCK COPOLYMER BRUSHES

Modified from a paper to be submitted for publication in Macromolecules. Unpublished work copyright 2010 American Chemical Society.

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Abstract

We report the phase behavior of a new class of polymer nanocomposites, montmorillonite (MMT) block copolymer brushes (MBBs). MBBs are comprised of discrete MMT particles encapsulated with block copolymer brushes synthesized via surface initiated atom transfer radical polymerization. Polystyrene-b-poly(n-butyl acrylate) MBBs of varying composition and total $M_n = 80$–250 kDa were compared with bulk block copolymers via electron microscopy and a host of mechanical tests. MBBs were found to self assemble into single grain morphologies across incredibly large areas ($>3 \mu m$). Perhaps most promising is that MBBs maintain similar melt state properties to bulk block copolymers and are thus amenable to current processing techniques.

Introduction

Hierarchical control of complex block copolymer (BCP) morphology has been progressively advancing for the past decade. Recently work has focused on confinement effects of BCP assemblies in lithographic patterned channels\textsuperscript{123–127}, oil-in-water emulsions,\textsuperscript{128,129} and surface grafted nanoparticles.\textsuperscript{130–132} Each of these phenomena can be viewed independently as specialized applications for self-assembled BCPs or collectively as the tools for hierarchically engineered materials.\textsuperscript{133} Of primary concern to this work is the application of BCP self-assembly to nanoparticle alignment; due to the sparse amount of literature about BCP grafted nanoparticles it is

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necessary to examine BCP behavior in lithographic channels (known as graphoepitaxy) to better understand the directed self-assembly (DSA) of BCPs.

The DSA of BCPs utilizes interfacial interactions and confinement effects to orient BCP morphology in a desired direction/pattern. Specifically, BCP channel orientation is governed by surface interactions, where neutral interactions with the substrate and sidewalls lead to lamellae oriented perpendicular to the sidewalls while a neutral substrate with preferential sidewall interactions results in parallel orientation.\textsuperscript{125,134} Similarly, BCP cylinders align normal to a neutral substrate and parallel to a preferential substrate.\textsuperscript{124,134–136} In graphoepitaxy, to achieve parallel lamellae alignment the channel width, $w$, must be nearly commensurate with an integer value, $n$, of the BCP period, $L_o$, (i.e. $w \sim nL_o$).\textsuperscript{123,124} Ruiz et al. demonstrated the ability to produce defect-free lamellae by using an open-ended channel design which created a high free energy penalty for lamellae defects inside the channel, resulting in near-perfect pattern orientation.\textsuperscript{123} They further tested the tolerance of BCP alignment by increasing channel spacing from $4L_o$ to $20L_o$ and found a constant deviation of $0.2L_o$ to be the maximum error tolerance. The greatest short-coming of lithographic patterning for BCPs is that the channel edge roughness is governed by the lithography process and cannot be improved upon by the BCP self-assembly.\textsuperscript{133} Cheng et al. avoided lithography edge roughness limitations by directly patterning the surface chemistry of the substrate.\textsuperscript{134} They created substrates with pinning stripes that selectively interacted with only one of the lamellar diblocks. A neutral area is then adjacent to the pinned stripes which allows the BCP to double or quadruple the pattern frequency and simultaneous improve edge roughness through a self-healing mechanism. While this method allows for frequency multiplying and is much improved on traditional channel graphoepitaxy, it still requires lithographic patterning from a negative-tone resist.

In the past two years researchers have begun working on means of lithography-free nanopatterning of BCPs. The Thomas Russell research group at the University of Massachusetts has been a leader in demonstrating how graphoepitaxy principles can be applied to broader applications. In 2008 Kim et al. researched two different methods for producing normal BCP cylinder orientation on bare silicon substrates.\textsuperscript{136} The first method involved the complicated anchoring of random copolymers directly to the substrate, while the second method simply required the mixing of hydroxyl-terminated homopolymers with the block copolymer prior to film casting. Both methods produced BCP cylinders aligned normal to the substrate by effectively neutralizing any surface interactions. Then in 2009 Park et al. published the ordering of perpendicular BCP cylinders, using sapphire crystals as a substrate.\textsuperscript{127} Park et al. further improved this lithography-free method by patterning polymeric replicas in order to increase the dimensions of BCP ordering.\textsuperscript{137}

To date the best example of combining graphoepitaxy with nanoparticle inclusion to produce highly aligned lamellar BCPs was accomplished by Park et al. of the IBM Almaden Research Center.\textsuperscript{125} Assymetric diblocks
of polystyrene-b-poly(ethylene oxide) (PS-b-PEO) were mixed with organosilicate spheres (OS) and spin-coated onto polydimethylglutarimide treated substrates. Even though the OS particles were not compatibilized with any short-chain polymers, they selectively segregated due to a strong enthalpic affinity for the poly(ethylene oxide) block. The resulting composite material successfully spanned channel widths of 100-400nm and channel lengths of 300-4000nm. Park et al. further demonstrated the ability of these BCP/OS composites to order along complex U- and L-shaped geometries.

Graphoepitaxy is very useful for the DSA of BCPs, conversely BCPs are capable of orienting nanoparticle filler materials in-plane with the developing BCP morphology. In order to achieve the maximum benefit of nanoparticle inclusions inside BCP structures a “symmetry matching” of the three dimensional nanoparticle shape with the three dimensional BCP assembly is necessary (e.g. carbon nanotubes with BCP cylindrical morphology). In an effort to produce matched anisotropic nanocomposites Ha et al. grafted polystyrene (PS) chains to montmorillonite (MMT) platelets and solution-blended the resulting PS-MMT brushes with lamellar styrene-butadiene-styrene (SBS) triblock copolymer. The dried solution was then roll-cast and demonstrated excellent in-plane orientation of the oblate MMT brushes within the SBS lamellar structure.

The inclusion of nanoparticles into BCP domains by grafted homopolymers has become commonplace in nanotechnology research however there are relatively few examples of BCP grafted directly to the nanoparticle surface. Li et al. produced silica particles with a difunctional initiator-terminated monolayer to produce spherical brushes. These “hairy” nanoparticles were produced by a combination of atom transfer radical polymerization (ATRP) and nitroxide mediated radical polymerization (NMRP). The resulting environmentally responsive nanoparticles formed colloidal suspensions in solvents as diverse as chloroform and methanol. Similar to Li et al. our research utilizes surface initiated ATRP, but instead of T-junction initiators we use end-tethered BCPs. The resulting MMT block copolymer brushes (MBBs) exhibit remarkable morphology phenomenon not seen with bulk BCPs, some of which has been published previously by our group. This article expands on that initial research to provide a comparison between n-butyl acrylate MBBs and tert-butyl acrylate MBBs and to produce a more complete morphological and mechanical analysis of the MBB phase diagram.

**Experimental Section**

**Functionalized montmorillonite.** Montmorillonite clay was generously supplied by Southern Clay Products Inc. Based on the ion exchange capacity, 92 mequiv/g, and measurements of the specific surface area, MMT contains $\approx 1 \text{ m}^2 \text{g}^{-1}$. Functionalized MMT (fMMT) was prepared via ion exchange of pristine MMT with a bromine terminated alkylammonium surfactant.

**MMT grafted homopolymer synthesis.** The surface-initiated ATRP (SI-ATRP) of homopolymers from
MMT can be summarized as the addition of fMMT, styrene (S), CuIBr, CuIIBr2, and PMDETA to a 50mL round bottom flask equipped with a 24/40 rubber septum and magnetic stir-bar under an N2 atmosphere. Chemicals were added in molar ratios relative to the amount of initiator used; 1 : 1000 : 1 : 0.06 : 1.06 respectively. A sonication probe was placed within one inch of the exterior of the reaction vessel for the duration of the experiment to maximize exfoliation without compromising the closed atmosphere. t-Butyl acrylate (tB) and n-butyl acrylate (nB) were also synthesized as homopolymers from the fMMT surface at 75°C and 65°C respectively. nB samples initially showed poor polydispersity (PDI), so they were allowed to sonicate for 6 hours at 30°C before reacting at 65°C.

Diblock synthesis from MMT homopolymer brushes. The subsequent polymerization of nB from MMT-polystyrene brushes was conducted by the addition of dry PS macroinitiators, nB, CuIBr, Cu0, and PMDETA to a 100mL round bottom flask equipped with a 24/40 rubber septum and magnetic stir-bar. Chemicals were added in molar ratios relative to amount of macroinitiator used; 1: 4000: 2 : 0.1: 2.1 respectively. The reaction flask was placed in an oil bath at 65°C with continuous stirring, and reaction time was varied to produce blocks of differing lengths. In the reverse reaction where PS was synthesized from nB macroinitiators a second equivalent amount of CuIBr, Cu0, and PMDETA were cannula transferred into the reaction vessel after 12h using additional styrene monomer as a solvent. MBBs are named based on MMT content(M), core block, corona block, and then number-average molecular weight ($M_n$) in kDa of each block respectively. Thus MnBS-60-63 (from Figure 5.5) is a MMT grafted (60 kDa)poly(n-butyl acrylate)-b-(63 kDa)polystyrene brush. Bulk BCPs of identical $M_n$ and composition were produced by reverse ion exchange of polymer from the clay surfaces via basic alumina flash chromatography.

Experimental Techniques. MBBs were annealed at 150°C in vacuo for over 96 hours prior to being steady shear processed at 0.5 s$^{-1}$ and 160°C for 20 minutes on a TA Instruments ARES-LS1 strain controlled rheometer in the parallel plate geometry under N2. Ultrathin (~100 nm) sections for electron microscopy were obtained at -100°C using a Leica Ultramicrotome Ultracut 125UCT with a Leica EM FCS cryo-stage. RuO4-stained sections were examined on a Tecnai G2 F20 scanning / transmission electron microscope at a high tension voltage of 200 kV.

Results

A library of samples containing styrene and either n- or tert-butyl acrylate MBBs has been synthesized with the intent of covering the entire phase diagram space for both tB (nB) core and tB (nB) corona MBBs. A complete list of examined tB-MBBs and nB-MBBs can be found in Tables 5.1 and 5.2 respectively. A great many more nB-MBBs were prepared than tB-MBBs owing to an increase in MBB synthesis expertise and the need to study
Table 5.1 Results of tB containing MBB polymerizations.

<table>
<thead>
<tr>
<th>Name-M_{n1}-M_{n2}</th>
<th>N</th>
<th>fS</th>
<th>PDI</th>
<th>Wt. Clay</th>
</tr>
</thead>
<tbody>
<tr>
<td>MStB-44-90</td>
<td>112.8</td>
<td>0.33</td>
<td>1.31</td>
<td>1.3%</td>
</tr>
<tr>
<td>MStB-44-64</td>
<td>92.4</td>
<td>0.40</td>
<td>1.17</td>
<td>1.6%</td>
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<tr>
<td>MStB-74-73</td>
<td>127.7</td>
<td>0.50</td>
<td>1.23</td>
<td>1.2%</td>
</tr>
<tr>
<td>MStB-44-36</td>
<td>70.1</td>
<td>0.55</td>
<td>1.14</td>
<td>2.1%</td>
</tr>
<tr>
<td>MStB-70-29</td>
<td>89.6</td>
<td>0.71</td>
<td>1.97</td>
<td>1.8%</td>
</tr>
<tr>
<td>MtBS-73-33</td>
<td>106.0</td>
<td>0.31</td>
<td>1.48</td>
<td>1.6%</td>
</tr>
<tr>
<td>MtBS-59-70</td>
<td>129.7</td>
<td>0.54</td>
<td>1.40</td>
<td>1.3%</td>
</tr>
<tr>
<td>MtBS-50-68</td>
<td>118.0</td>
<td>0.58</td>
<td>1.39</td>
<td>1.5%</td>
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<tr>
<td>MtBS-59-145</td>
<td>204.3</td>
<td>0.71</td>
<td>1.39</td>
<td>0.9%</td>
</tr>
<tr>
<td>MtBS-60-172</td>
<td>231.7</td>
<td>0.74</td>
<td>1.45</td>
<td>0.8%</td>
</tr>
</tbody>
</table>

MBB mechanical properties independent of tert-butyl cleavage. A detailed discussion on the molecular assembly leading to MBBs’ morphology has previously been published by this group; for this work it is sufficient to know that MBBs can be viewed as large pseudo-particles where intraparticle versus interparticle assembly is a critical parameter.  

Looking first at the tB-MBBs, since it is the smaller sample set, there appears to be five distinct morphologies: worm-like cylinders (W), interpenetrating networks (IPN), oblate spheres (OS), isolated discoids (ID), and toroids (T). Figure 5.1 is a phase diagram of all the tB-MBBs examined with representative micrographs to demonstrate each phase; it is important to note that the phase lines are only approximations intended to guide the eye. The complex morphologies seen in this figure appear to follow more of a continuum rather than the rigid boundaries observed in tradition bulk BCPs morphologies. As can be observed in the series of micrographs 5.1g → 5.1h → 5.1f → 5.1c, a barely ordered ID morphology (5.1g) transitions to a more regular ID pattern (5.1h) that fills the entire void space. Then there is a more dramatic transition to OS (5.1f), but there still remains similarities to the previous ID morphology. Finally, a fully developed OS morphology forms (5.1c) that is clearly distinct from the ID morphology. This series of micrographs also demonstrates that block sequence (tB-PS versus PS-tB) has little to no impact on the MBB morphology formed. Sequence independence is corroborated by the nB-MBBs (Figure 5.2) where the lamellar (5.2c-d, 5.2f-g) and disordered lamellar (5.2b, 5.2e) phases are represented by both interior and exterior nB-MBBs.  

It was hypothesized that the large BCP polydispersities (PDI > 1.2) observed in tB-MBBs could be contributing to the indistinct morphologies developed, therefore a literature survey was conducted to evaluate this possibility. In a very methodical study of PDI effects in BCP morphology formation, Lynd and Hillmyer inves-
Table 5.2 Results of nB containing MBB polymerizations.

<table>
<thead>
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tigated the selective increase of PDI of the \(_{DL}\)-lactide block of a poly(ethylene-\(alt\)-propylene)-\(b\)-poly(\(_{DL}\)-lactide) diblock copolymer.\(^{141}\) Lynd and Hillmyer observed a linear increase in lamellar domain spacing with increasing PDI resulting from large chains filling the void space between domains more easily, thus eliminating the need for unfavorable stretching. When non-lamellar samples were examined the asymmetric increase in PDI of the minority block induced phase transitions towards increased mean interfacial curvature (i.e. lamellar became gyroid etc.), conversely increasing PDI in the majority block decreased mean interfacial curvature. The change in interfacial curvature was attributed to the movement of the interface towards the block that previously had the larger stretching penalty. Thus it seems unlikely that PDI had any significant impact on the indistinct morphologies observed in tb-MBBs but it may contribute to the large domains spacings observed.

The larger nBA-MBB sample set displays a great deal more symmetry than the tb-MMBs and follows more traditional BCP morphologies: perforated lamellae (\(PL\)), disordered lamellae (\(LD\)), and lamellae (\(L\)). Figure 5.2 is an nB-MBBs phase diagram with select micrographs to demonstrate each phase; it is important to remind the reader that the phase lines are only approximations intended to guide the eye. Samples that fall outside of the approximate phase boundaries typically demonstrated incomplete phase separation as shown in Figure 5.2f. This inability to fully phase separate is attributed to the constraint imposed on mean interfacial curvature by the large flat clay platelets.

Understanding MBB phase behavior is important for designing materials with desirable anisotropic properties, equally important is the melt state behavior as measured by rheology. Figure 5.3 presents four time-temperature superposition (TTS) master curves for both grafted and liberated homopolymers/BCPs. The MMT grafted samples are the shaded in symbols while the bulk BCP samples are the open symbols. Terminal relaxation of BCPs has been used in the past as a facile method for screening BCP phase behavior.\(^{142}\) However, before extracting any inferences about MBB structure from rheological data it is important to establish baseline behavior of disordered materials. The easiest way to assure a disordered material is to examine free homopolymer as shown in Figure 5.3a, its terminal behavior, \(G' \propto \omega^{1.82}\), is in close agreement with the liquid-like behavior of previously reported disordered homopolymers (\(G' \propto \omega^{2}\)).\(^{142}\) The MMT grafted homopolymer on the other hand displays a terminal regime that reflects the influence of chain tethering, \(G' \propto \omega^{0.56}\). Figure 5.3b-d are rheology comparisons of MBBs to their bulk BCP counterparts. The behavior of grafted and free samples of the same composition are remarkably similar with nearly identical terminal behavior. Looking at the series based on composition it becomes apparent that the MBB sample shifts from having a higher modulus than the bulk (5.3b), to nearly equal (5.3c), to significantly lower (5.3d). Given how close MSnB-66-36 and MSnB-66-30 are in total molecular weight and composition it seems obvious that an entirely different mechanism is responsible for this shift in modulus intensity. Based on the rheology results of additional samples (including nB interior MBBs),
Figure 5.1  Representative TEM micrographs of different observed morphologies in PS/PtBA MBBs: worm-like cylinders (W), interpenetrating networks (IPN), oblate spheres (OS), isolated discoids (ID), and toroid (T). Phase lines are approximations intended to guide the eye. Triangles (△) refer to PS core MBBs and squares (□) refer to PtBA core MBBs. Scale bar for (e) is 100nm otherwise scale bar is 50nm.
Figure 5.2 Representative TEM micrographs of different observed morphologies in PS/PnBA MBBs: perforated lamellae (PL), disordered lamellae (LD), and lamellae (L). Phase lines are approximations intended to guide the eye. Diamonds (○) refer to PS core MBBs and circles (●) refer to PnBA core MBBs. Scale bar is 50nm.
it appears that morphology is the governing factor for relative intensities of MBB modulus versus bulk BCP modulus. In the case of the lamellar morphology MBBs and bulk BCPs appear to have nearly identical moduli (5.3c), while lower \( f_S \) morphologies result in superior MBB moduli (5.3b), and higher \( f_S \) morphologies result in superior bulk BCP moduli (5.3d). Although these results show promise for designing a stronger or tougher MBB, more research is needed before any conclusive statements can be made.

In addition to the melt state rheological behavior of MBBs we have also studied the solid state tensile behavior. Table 5.3 displays a comparison of MBB ultimate stress, \( \sigma \), and Young’s modulus, \( E \), for MBBs and their corresponding bulk BCPs. Three factors have been observed to noticeably impact \( \sigma \) and \( E \): MBB core block chemistry, fraction \( S \), and morphology. MBBs with a S core demonstrate moderate elevations in \( \sigma \) and \( E \) relative to bulk while the nB core MBBs demonstrate moderate reductions. The fraction \( S \), \( f_S \), in MBBs and bulk BCPs is by far the principle factor contributing to overall \( \sigma \) and \( E \), however it is surprising that \( f_S \) has a much larger impact than total \( M_n \). The impact from \( f_S \) is especially noticeable when comparing S core MBBs versus nB core MBBs, where MSnB-59-22 clearly has superior tensile properties relative to MnBS-41-144 even though its S block is almost 100kDa smaller. Location on the phase diagram also plays a strong role in MBB tensile properties, specifically looking at the center of the phase diagram. Sample MSnB-62-77 displays nearly an order of magnitude increase in \( E \) and a 3 fold increase in \( \sigma \) relative to its bulk analog, while sample MnBS-60-65 presents nearly the exact opposite behavior where \( E_{\text{bulk}} > E_{\text{MBB}} \) by 100 MPa and \( \sigma_{\text{bulk}} > 3^*\sigma_{\text{MBB}} \). Beyond the quantitative tensile properties of MBBs, is the qualitative shape of the material deformation. Figure 5.4 presents two stress strain curves which exemplify the two-stage yield behavior observed for many MBB samples. The exact mechanism for the MBB two-stage yielding is not yet known but we believe that the first yield point is related to chain disentanglement while the second yield point arises from the difficulty of large MBB pseudo-particles to translate past one another.

**Discussion**

Given the incredibly complex nature of examining silicate nanocomposites it is important to reexamine MBBs in respect to orientational control of the MMT filler. Looking first at the tB-MBBs, it is apparent that morphological control is difficult at best and unreliable at worst. The cleavage of t-butyl groups creates hard to reproduce circumstances and is known to cause a shift in bulk BCP morphology equal to the compositional shift of removing the mass of the t-butyl groups. However much more complex and large range morphologies are forming than can be accounted for by simply shifting along a typical BCP phase diagram. The charge of the remaining acrylic acid groups and the inherent anionic charge of the clay likely are playing a significant role in creating these novel structure formations. These samples have overcome the clay curvature limitation
Figure 5.3  Time-temperature superposition master curves for both grafted and liberated (a) homopolymer and (b-d) block copolymers. MMT grafted samples are the shaded symbols and bulk BCP samples are the open symbols, where diamonds (⋄) represent $G'$ and circles (◦) represent $G''$. The sample $M_n$'s and styrene fraction are as follows (a) MS-35; (b) MSnB-68-78, $f = 0.46$; (c) MSnB-66-36, $f = 0.65$; (d) MSnB-66-30, $f = 0.69$
Table 5.3 MBB/bulk BCP tensile testing comparison

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Figure 5.4 Tensile testing stress strain curves for MSnB-62-77 ($f = 0.45$) and MSnB-66-44 ($f = 0.60$) demonstrating a two-stage yield behavior which was observed for many of the MBB samples.
by strongly stretching the tethered brushes to increase domain sizes. Many of the samples display elliptical structures measuring tens to hundreds of nanometers across, which is commensurate with the dimensions of a highly bent MMT platelet.\textsuperscript{143}

In order to achieve a fundamental understanding of MMT/BCP interactions our work shifted to focus on nB-MBBs where cleavage and charged polymer effects are no longer a concern. What we found is a strong tendency to form lamellar structures and a strong resistance to forming other morphologies. The incorporation of clay particles in the lamellar morphology promotes the formation of defect-free areas by creating a high free energy penalty for misaligning the large MMT sheets away from lamellar domains. Figure 5.5 is a micrograph of MnBS-60-63 demonstrating a 4x4µm square of defect free lamellae. Also contributing to defect-free lamellae is an effect similar to the open-ended channels of Ruiz et al., where MBBs offer a terminal zone of increased polymer mobility such that defects can migrate to the edge of the MMT platelets. Even though MMT introduces a free energy of bending which encourages an enthalpic trend towards lamellar formation, it is surprising that the observed lamellar window of 0.45 < \( f_{St} \) < 0.7 is comparable in size to that observed by Bates and Fredrickson when studying polystyrene-\( b \)-polysisoprene BCPs.\textsuperscript{144}

In the case where BCP composition is just too asymmetric to form lamellae the MMT bending energy becomes more apparent. The relatively large MMT platelets appears to inhibit the formation of morphologies that requires high mean curvature. The semi-rigid platelet resists the deformation necessary to accommodate the non-lamellar BCP morphologies. The best example of this phenomenon is the partially segregated cylindrical morphology in Figure 5.2f. It is possible that a fully developed cylindrical morphology is still forming and has not yet reached equilibrium as asymmetric BCPs microphase segregate more slowly than their symmetric counterparts.\textsuperscript{145}

**Conclusions**

The synthesis of a library of tB-MBBs and nB-MBBs was accomplished and allowed for the comparison of phase diagrams for two nearly identical materials. Some remarkable morphologies were realized in terms of domain period, complex structure, and long-range order. The development of an extremely well-ordered lamellar structure is of particular interest to microelectronic applications. Furthermore the demonstration of well ordered and aligned MMT filler provides the opportunity to test computational small molecule penetrant models with a experimental system.

The nB-MBBs were then further studied for mechanical properties of MBBs in the melt and solid state. A comparison to bulk BCPs of identical molecular weight and composition was provided, however no dramatic improvements in mechanical properties were observed except in a few rare instances. It is possible that the
Figure 5.5 TEM micrograph of sample MnBS-60-63, demonstrating a 4x4μm square of defect free lamellae. Inset shows incredible regularity of lamellae. Scale bar is 1 μm.
reverse ion exchange process is unreliable resulting in a combination of BCPs and MBBs during mechanical testing. This area certainly merits further research to optimize MBB mechanical properties.

Acknowledgements

Acknowledgment for support of this research is made to the Donors of the American Chemical Society Petroleum Research Fund (PRF #48399-G7), the Camille and Henry Dreyfus New Faculty Award Program, and the National Science Foundation, DMR-0847515.

Bibliography


CHAPTER 6  FUTURE WORK

Post-Synthesis Property Evaluation

Nanocomposite’s potential property enhancements arise from synergistic effects generated by high interfacial contact between polymer and nano-sized fillers. Therefore it would seem that high graft density BCP brushes attached to large discoidal silicates would exhibit substantial mechanical and barrier property enhancements. While barrier properties have not yet been tested by this group, melt and solid state properties have and at best only a nominal improvement was observed. Since the free BCPs were produced by reverse ion exchange from the MMT-grafted-BCPs, three possible explanations exist for this nominal difference in mechanical properties. One, the reverse ion exchange process did not effectively de-graft the BCP chains so it is not truly free polymer; two, the charged end-groups on the liberated BCPs produce similar properties to grafted chains; or three, there is no difference in BCP and MBB mechanical properties.

Figure 6.1 compares TTS for MBB and de-grafted BCP (a)MnBS-66-44, \( f_S = 0.60 \) alongside a MBB and bulk synthesized BCP (b)MStB-54-49 \( f_S = 0.52 \), BCP-62-46 \( f_S = 0.57 \). The important factor here is that in Figure 6.1b the BCP was synthesized separately so a true comparison of the de-grafting process can be made with the obvious result is that there is almost no difference in the MBB and the bulk BCP. Although this lends strong credibility to the third scenario, that there is no difference in BCP and MBB mechanical properties, it does not guarantee that chains are truly de-grafted. The fact that none of the de-grafted TEM micrographs demonstrated clear microphase separation is disconcerting at the least.

In addition to further TEM examination of de-grafted BCPs to confirm microphase separation a further TEM study of non-lamellar MBBs and post tensile testing samples is needed. Figure 6.2 is a TTS curve of MnBS-60-35 \( f_S = 0.36 \) which based on the rheological data suggests that it should be demonstrating cylindrical morphology. However based on the micrograph of MnBS-60-35 (Figure 6.2 inset) it is only showing incomplete microphase separation at best. A study of post tensile testing samples would be beneficial to confirm the cause of the two-stage yielding mechanism and to evaluate the impact of morphology on tensile properties.
Figure 6.1  Time-temperature superposition master curves comparing relative moduli for de-grafted BCPs versus bulk BCPs. (a) TTS for MBB and de-grafted BCP: MnBS-66-44, $f_S = 0.60$. (b) TTS for MStB-54-49 $f_S = 0.52$ and BCP-62-46 $f_S = 0.57$. MMT grafted samples are the shaded in symbols and bulk BCP samples are the open symbols, where diamonds (●) represent $G'$ and circles (○) represent $G''$.

Figure 6.2  Time-temperature superposition master curves for both grafted and liberated MnBS-60-35, $f_S = 0.36$. MMT grafted samples are the shaded in symbols and bulk BCP samples are the open symbols, where diamonds (●) represent $G'$ and circles (○) represent $G''$. The inset is a TEM micrograph of the MBB sample.
Novel Synthesis Routes

The synthesis of nB-MBBs radically altered our view of MBB morphologies, both in terms of BCP arrangement but also in terms of domain size. It was believed that the graft-from synthesis of BCPs from MMT was responsible for generating the morphologies realized in Chapter 3, however now it seems more plausible that the loss of tert-butyl groups was responsible for the large domains sizes observed in tB-MBBs. The loss of the tert-butyl groups would created poly(acrylic acid) (AA) MBBs, which given their negative charge would repel from the anionic clay surface. Obviously it was not strongly charged enough to cause de-grafting of the quaternary amine end-groups or else no microphase separation would have been observed. Furthermore, the most extreme morphologies observed (oblate spheres and toroids) were from tB-MBBs with S core blocks \( \geq 70\text{kDa} \). Thus moderately charged polymers attached to large (>100kDa) S-MMT brushes are strong candidates to produce large BCP domain sizes for non-linear optics. One potential polymer to explore the effect of charge on BCP domain size would be poly(di-ethyl amino ethyl methacrylate) (PDEAEMA) as it can be treated with acid to swell the PDEAEMA layer. PDEAEMA makes a good starting polymer since expertise to synthesize it already exists in this group and it is positively charged, thus helping to establish whether charge is directly responsible or not.

In addition to the production of novel MBBs, some research into improved removal of CuI X and CuII X\(_2\) after synthesis would be beneficial. Beyond eliminating the tainted color from post-processing samples, removal of copper would be beneficial if any biological application from ATRP are desirable. A group at the USAF has been studying the removal of Grubb’s catalyst from polymers used for light emitting diodes. They found a reduction from \( >200\text{ppm} \) of Grubb’s catalyst to \( <15\text{ppm} \) by adding 300 molar equivalents of the Grubb’s ligand to the post-synthesis flask prior to precipitation of the polymer. Therefore it can be expected that a similar addition of 300 molar equivalents of ATRP ligand would dramatically enhances copper removal during precipitation also. This is entirely practical as well since PMDETA is an industrially produced ligand, thus very inexpensive.

Bibliography

APPENDIX A  COMPLEX POLYMER ARCHITECTURES FOR DRUG DELIVERY: FROM THERMOGELS TO POLYMERSOMES

Modified from a paper to be submitted to Journal of Controlled Release.

Ross E. Behling\textsuperscript{1,2} and Balaji Narasimhan\textsuperscript{1,3}

Abstract

The use of advance architecture copolymers in drug delivery applications has grown from the narrow subset of polymers approved by the Food & Drug Administration (FDA) to a wide variety of micelles and polymersomes. The necessity for a strong understanding of how polymer chemistry interacts with targeted therapeutic agents and how it can be modified to provide better delivery methods is becoming readily apparent as the low hanging fruit offered by pre-approved FDA polymers runs out.

Preamble

The stringent restrictions of the Food and Drug Administration (FDA) limit the number of pre-approved monomer species available for biopolymer scientists to choose from to quickly validate CPA’s drug carrier capabilities. Poly(lactic acid) (PLA), poly(ε-caprolactone) (PCL), and poly(glycolic acid) are three of the most frequently used FDA approved hydrophobic CPA components due to their inherent biocompatibility and biodegradability.\textsuperscript{147} Unfortunately this narrow focus on only a few monomer species forces biopolymer scientists to work from a top down approach and chose non-ideal chemistry, when a bottom-up approach would much more quickly benefit the medical community.

This review is designed to expose biopolymer chemists to the broad range of successful work that is being done with complex polymer architectures (CPAs) for drug delivery with an emphasis on the types of polymers

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\textsuperscript{2}Primary author.
\textsuperscript{3}Author for correspondence.
used and the associated successful applications of those architectures. This review is arranged to first exam CPAs from a synthesis and characterization perspective and then reexamine the same CPAs in terms of applications and drug release. The intended goal of this review is to encourage biopolymer chemists to design new and improved polymer delivery devices by exploring alternate polymer chemistries. Finally, several areas where advances in CPA design that would greatly benefit the medical community are highlighted.

**Polymer Synthesis and Architecture**

**Introduction to Polymer Architecture**

Polymer architecture is a description of the ordering of monomer species in terms of sequence on a single chain and pendant chain spatial orientation relative to the main chain backbone. Figure A.1 displays a number of polymer topologies, compositions, and functionalities that can be achieved through controlled polymerization methods and carefully designed reaction conditions. Of greatest interest to this review is block copolymers, random copolymers, and graft copolymers. In brief, block copolymers (either diblock or triblock) are predominantly used to form micelles and vesicles, random copolymers are used to form bioerodible microspheres, and graft-copolymers are used for forming thermoresponsive gels.

Figure A.1  A schematic representation of how new polymers and materials can be prepared from a few monomers using controlled/living polymerizations. Reproduced from Ref. 148.
Table A.1 Overview of complex polymer architectures and chemistries.

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</tbody>
</table>

For those unfamiliar with polymer architecture a brief overview follows. Block copolymers are essentially two homopolymer chains that are covalent bonded at the middle. Each polymer block has a strong enthalpic preference to aggregate with its own block, but is unable to separate completely because of the central covalent bond creating microphase separation. Microphase separation leads to macroscale morphologies which are governed by the length of the A and B blocks and the Flory interaction parameter, $\chi$, which is the degree to which the A and B blocks “want” to be separated from each other. Random copolymers, often called statistical copolymers, are synthesized by the concurrent polymerization of two different species to produce a random sequence distribution of both monomer species. Altering the final copolymer composition is easily done by either changing the initial monomer concentrations or adjusting the monomer ratio during the reaction to change the final composition. In the most extreme case of starting a reaction composed entirely of species A and switching the reaction composition to entirely species B a gradient copolymer is formed, where the far ends of each chain is a homopolymer of either A or B. Graft copolymers are similar to block copolymers in that they have two different homopolymer species covalently bonded together, however graft copolymers have a single main chain back bone that is entirely one polymer species and then the pendant chains are covalently bonded polymers of a different species. The pendant chains can either be \emph{grafted-from} polymers at all pendant locations (i.e. being synthesized directly from the polymer back bone) or they can be synthesized separately and then \emph{grafted-to} site-specific functional locations on the back bone. Table A.1 provides a summary of some of the different polymer structures and chemistries used to produce complex polymer architectures.
Block Copolymers

The most extensively studied block copolymer (BCP) systems have been micelle forming amphiphilic BCPs, which were first examined to encapsulate highly hydrophobic therapeutic drugs. Typical micelle sizes (10-100 nm) are easily tuned via molecular weight control, such that their uptake by the mononuclear phagocyte system is minimized. A subgroup of BCPs that have received ever increasing attention as a new drug delivery vehicle is amphiphilic block copolymers (ABCP). ABCPs form stable micelles in aqueous solutions with a hydrophobic core, ideal for encapsulating water insoluble drugs. The corona block of ABCPs is typically poly(ethylene glycol) (PEG) or a random copolymer containing PEG; it should be mentioned that PEG is frequently referred to in polymer literature as poly(ethylene oxide) (PEO). The transition from PEG to PEO seems to occur when its molecular weight exceeds 10,000 g/mol ($M_n > 10$ kDa), although there is no agreed upon limit. It was previously stated that micelle BCPs resist uptake by the mononuclear phagocyte system; PEG further reduces ABCPs elimination by preventing the reticuloendothelial system from recognizing micelles in the bloodstream. PEG is also preferred as it has excellent water solubility, high resistance to protein adsorption, and low cellular adhesion. All of these factors contribute to long circulation times which is ideal for a prolonged delivery period.

The synthesis and micellization of an amphiphilic ABA triblock copolymer of poly(sebacic anhydride) (SA) and poly(ethylene glycol) (PEG) was prepared by Zhang et al. via melt polycondensation of PEG macroinitiators. Low molecular weight ($M_n < 5.1$ kDa) polymer was produced with modest polydispersities ($1.20 \leq M_w/M_n \leq 1.35$) and then converted to micelles via the precipitation/solvent evaporation method. Micelles ranged from 122-148nm in diameter which was easily tuned by controlling the dropping rate (mg/s).

Poly(ethylene oxide)-block-poly(isoprene) (PEO-b-PI) micelles were synthesized by nitroxide-mediated free radical polymerization (NMFR) by Wegrzyn et al. PEO macroinitiators ($M_n \sim 5.2$ kDa) were functionalized with an alkoxyamine terminal group, which was used for NMFR polymerization of poly(isoprene). Diblock molecular weights of almost 20kDa were achieved with polydispersities as low as $M_w/M_n \leq 1.1$. Resulting micelles were capable of undergoing PI crosslinking and retaining structural integrity upon dehydration as confirmed by atomic force microscopy (AFM).

The ring opening polymerization of $DL$-Lactide and glycolide (PLGA) from a difunctional PEG macroinitiator to produce ABA triblock copolymers of PLGA-PEG-PLGA were synthesized by Zentner et al. The PLGA second block synthesis was carried out at 155°C and resulted in modest polydispersities of $M_w/M_n \sim 1.3$. The resulting product has been patented and approved by the FDA under the ReGel® trademark and has been used for the delivery of proteins and hydrophobic drugs. A similar ABA triblock synthesis has been carried out by Jo et al. where the $L$-lactide was copolymerized with $\epsilon$-caprolactone (PLLACL) to produce PLLACL-PEG-PLLACL. Re-
actions were conducted at 80°C and resulted comparable molecular weights and superior polydispersities $M_m/M_n \sim 1.13 - 1.25$. Both PLGA and PLLACL triblocks formed thermoreversible gels, however the PLGA samples had release profiles on the order of 40-50 days while the PLLACL samples released in 10-15 days.

The synthesis of thermosensitive star block copolymers consisting of three poly(N-isopropylacrylamide) (PNIPAAm) arms and one poly(methyl methacrylate) (PMMA) arm was accomplished by Wei et al. using reverse addition-fragmentation chain transfer polymerization (RAFT). The synthesis of PNIPAAm and PMMA macroinitiators was accomplished with straightforward RAFT synthesis, using 3-mercaptopropionic acid (MPA) as a chain transfer agent. An amino-terminated PNIPAAm was synthesized using 2-amino ethanol hydrochloride as a chain transfer agent, then combined with pristine PNIPAAm and N-acroylosuccinimide to produce a three armed PNIPAAm homopolymer. A carboxyl terminal group was then added to the PMMA macroinitiator via MPA catalyzed RAFT polymerization and the PMMA-COOH was combined with the three arm PNIPAAm to produce the final thermosensitive star block copolymer.

**Random Copolymers**

The rapid synthesis of polyanhydrides by microwave radiation was conducted by Vogel et al. to produce poly(sebacic anhydride) (SA) and poly(1,6-bis(p-carboxyphenoxy)hexane) (CPH) homo- and co-polymers. Poly(SA) samples 10-15 kDa in size were polymerized in as little as 5 minutes and poly(CPH-rand-SA) samples up to 16 kDa were synthesized in 15-20 minutes. The remarkable rate of polymerization is unfortunately limited to small batch sizes (< 1g). More conventional polyanhydride synthesis via melt polycondensation reactions have previously been reported in the Narasimhan group. Irrespective of polymerization mechanism, polyanhydrides are easily formed into surface bioerodible microspheres of considerable interest due to their easily tunable release profiles.

**Graft Copolymers**

The synthesis of bioerodible pH sensitive graft copolymer thermogels of polyacetal-graft-poly(betraylene glycol) and polyacetal-co-poly(ortho ester)-graft-poly(betraylene glycol) were prepared by Schacht et al. using a polycondensation reaction mechanism. A primary amine was protected until after the synthesis of the back bone chain and then a NHS ester reaction was utilized to attach the betylene glycol side chains to the active amine sites. The resulting graft copolymers had tuneable release profiles from 14-80 days depending on polyacetal composition and pH.
Preparation & Characterization of Materials

Micelle preparation

The preparation of polymer micelles from highly hydrophilic low molecular weight ABCPs can be as simple as mixing the polymer with a hydrophobic drug in water and allowing equilibrium to form loaded micelles. This usually results in poor drug loading and is obviously a very specific case of ABCPs, therefore the techniques of dialysis, solution casting, oil-water (O/W) emulsions, and freeze-drying have been applied to encapsulate drugs based on the relative hydrophobicities of the ABCPs and the drug. For ABCPs that are water insoluble an organic solvent that co-dissolves the ABCPs and the drug moiety is chosen. If the organic solvent is miscible in water then dialysis is used to slowly remove the organic phase until micelle formation is induced. A slightly less water miscible organic solvent will often be used to solution cast the combined drug/ABCP to encourage drug and hydrophobic block interactions, followed by rehydration in an aqueous solvent. If the organic solvent used is completely immiscible in water then an O/W emulsion can be formed to produce drug-loaded micelles. Freeze-drying can be used for any of the aforementioned methods as a means to extend the shelf-life of loaded micelles.

The chosen preparation method plays a large role in final micelle size, size distribution, and stability. It is important to carefully control these quantities as circulation times, distribution in the body, and release profiles are strongly dependent on micelle structure. In order to optimize micelle preparation, Vangeyte et al.
conducted a comparison of dialysis, rapid water addition to organic solutions, and rapid organic addition to water (all organic solvents examined were highly miscible in water). ABCPs composed of PEO-\(b\)-PCL and PEO-\(b\)-poly(\(\gamma\)-methyl-\(\varepsilon\)-caprolactone) (PMCL) were studied representing a crystalline and amorphous polymer respectively and they were both found to form large particles (~1 \(\mu\)m) with high polydispersity from dialysis. The rapid addition of water to organic solvent or vice versa resulted in much better micelles with diameters 30–100 nm depending on solvent type. Water addition showed no dependence on polymer concentration, however, organic solvent addition resulted in nearly a doubling of diameter across an order of magnitude decrease in concentration. Lastly, Vangeyte et al. examined the contribution of the hydrophobic block length to micelle formation and it had no effect except when the organic solvent was dimethylsulfoxide and then it was a significant factor. Overall Vangeyte et al. present an excellent summary of how many factors contribute to the final micelle size and size distribution and thus must be carefully controlled to optimize micelle preparation.

At present the race to produce BCPs that are competitive with more conventional delivery methods is ongoing with BCPs faltering along several fronts; most notably micelle structural stability, drug loading retention, and controlled release rates.

**Polymersomes**

The endeavor to explore CPA drug delivery beyond FDA approved polymer species has been begun by a number of groups; most notable of these efforts is those involving larger self-assembled structures which offer “cell-sized” delivery vehicles, namely polymer vesicles and polymersomes. Polymersomes offer increased drug loading capacities and the validation that other monomeric species may be used without cytotoxic effects. Polymer vesicles and polymersomes are composed of monodisperse block copolymers which form a polymer membrane around an enclosed hydrophobic environment. While micelles typically produce self-assemblies on the order of 30–40 nm, polymersomes assemble up to 100–200 nm in diameter, and vesicles have been documented to self-assemble up to 25 \(\mu\)m in diameter. The key distinction between micelles and polymersomes (other than size) is the capability of polymersomes to encapsulate a secondary solution phase. One of the major hurdles for polymersomes is the concern that vinyl-based polymers are not biodegradable but by choosing an appropriate biodegradable diblock the vinyl polymer blocks can fall below the renal excretion molecular weight limit of ~40 kDa and be removed from the body. An additional concern of vinyl-based polymers is their cytotoxicity but several polymers (e.g. polyethylene and polybutadiene) have already been shown to be bioinert.

Polymersomes are a large molecular weight (MW) subset of amphiphiles which self-assemble into polymer vesicle shells. Polymersomes are closely related to liposomes as both are constructed by two interfaces of a bilayer membrane, but polymersomes have larger MWs, are typically composed of synthetic polymers, and
can achieve an areal strain nearly an order of magnitude higher than lipid membranes.\textsuperscript{167} The effective design of polymersomes requires careful consideration of the hydrophilic fraction $f$, overall MW, and relative block volume fractions $\phi$. The importance of each of these design parameters has been explored in a series of papers by Discher et al. where $f$ is predominantly responsible for aggregate stability, MW for shell thickness $d$, and $\phi$ for morphology.\textsuperscript{166,168,169} Figure A.3 depicts the contribution from $f$ on vesicle (25-40%), worm-like micelle (40-50%), and micelle (>50%) formation.

### Applications of CPAs

**Introduction to CPA Applications**

The use of CPAs for pharmaceutical applications is a relatively young field with great potential to enhance therapeutic drug delivery. The benefits of CPAs arise from the interaction of the drug with the encapsulating polymer in terms of polymer microstructure, compatibility of the drug with the polymer, and solubility of the drug in the polymer.\textsuperscript{170} The intelligent design of drug delivery devices for biomedical applications necessitates a detailed understanding of the drug interactions with the host polymer as well as the release mechanism of the delivery device. This is especially true for biodegradable polymers where the diffusion of drugs through the polymer, the polymer degradation rate, and diffusion of degraded monomer away from the bulk surface all play key roles in determining drug release rates.\textsuperscript{171}
Micelles

Using micelles for drug delivery offered many advantages over more conventional drug encapsulation methods. The hydrophobic core easily associated with water insoluble drugs, resulting in high loading capacities, and the hydrophilic corona easily solubilized the micelles. Additionally, micelles possess a passive targeting mechanism whereby they pass through leaky vasculature and aggregate in tumors and inflamed tissue. The micelles’ small size allows them to accumulate at damaged tissue, which is known as the enhanced permeability and retention effect (EPR); this aspect of micelle drug delivery is particularly promising for the targeting of solid-tumors. Figure A.4 illustrates the differences in healthy vasculature versus that found near solid-tumors and how low molecular weight drugs interact at these sites versus nano-scale micelles.

ABCPs for drug therapy often contain a degradable component to adjust the time and duration of drug release, known as temporal control. The ability to control release rates and profiles is very important for pharmaceuticals as many drugs require a minimum dose levels to achieve effective treatment responses. Although temporal control is extremely important for treatment it still requires micelles that can keep their spherical capsule structure even at low concentrations. A method for insuring that ABCPs do not break apart in the body before reaching the target location was developed and is known as micelle stability, where stable ABCPs must have a
micelle concentration above the critical micelle concentration (CMC). One method for preventing ABCPs disassociation is by crosslinking part of the system, be it the core, the corona, or a surface functionalized crosslinking agent. Core crosslinking has the disadvantage of reducing free volume and hence hydrophobic drug loading. Corona and surface crosslinking however avoid this pitfall and additionally allow for fine tuning of the drug release rate since the degree of crosslinking will affect the corona permeability.\textsuperscript{155}

The first reported instance of micelle corona crosslinking was by Wooley et al. wherein polystyrene-\textit{b}-poly(4-vinyl pyridine) (PS-\textit{b}-PVP) micelles were initially formed.\textsuperscript{172} The PVP amine was then quaternized with \textit{p}-(chloromethyl) styrene to produce a crosslinking agent which was then polymerized using a photo-initiator. The final micelles formed monodisperse spheres with little to no interparticle crosslinking as confirmed by AFM. In a follow-up study, Wooley et al. pursued micelle corona crosslinking to produce hollow-sphere nanocages.\textsuperscript{173} These nanocages were synthesized from much more typical ABCPs species consisting of polyisoprene-\textit{b}-poly(acrylic acid) (PI-\textit{b}-PAA). After crosslinking the corona with a PEO-diamine, the PI core was removed via ozonolysis and the remaining PAA nanocages were re-solvated via dialysis. The final nanocages exhibited significant solvent swelling (> 100% increase in diameter) as a result of the nanocage’s high porosity, reminiscent of hydrogels.

\textbf{Micelle in vivo studies}

The selection process for any novel pharmaceutical application undergoes several levels of analysis before even being considered for clinical trials. The lowest level being general chemical considerations and targeting chemical functionalities, the next being the slightly more rigorous \textit{in vitro} studies for successful treatment and toxicity, and the last being the \textit{in vivo} study of the therapeutic agent, typically in rodents. The following is a discussion of some of the more successful research thrusts in \textit{in vivo} studies and beyond.

Among the first examples of micelle \textit{in vivo} studies was conducted by Kim et al. using monomethoxy poly(ethylene glycol)-\textit{b}-poly(D,L-lactide)(mPEG-PDLLA) micelles to release paclitaxel for antitumor treatment.\textsuperscript{174} Kim et al. tested the efficacy and maximum tolerated dose (MTD) of their packed micelles versus a current clinical formulation Taxol\textsuperscript{®}. They found tumor volume reduction to be significantly greater for the mPEG-PDLLA micelles than that of Taxol\textsuperscript{®}, with a MTD of 205-222 mg/kg versus 8-9 mg/kg respectively. Additionally, the use of polymer micelles in this instance was particularly beneficial since it eliminated the need to use Cremophor EL solution as part of the formulation, a poorly tolerated solubilizing agent that may lead to hypersensitivity reactions in some patients.
Biodegradation of polymersomes

Polymersomes stability is a key trait for drug delivery applications to lengthen circulation times and reduce leakage. However, an increase in stability makes disassembly for drug release difficult, so many polymersomes have been designed for biodegradation to improve payload delivery or to facilitate burst release. The use of PLA or PCL in conjunction with PEO allows for stable polymersomes which can undergo polyester hydrolysis; further control of capsule release rates can be gained by blending PEO with inert polymers.\textsuperscript{166} Alternatively, thiolytic cleavages have been used to destabilized agglomerated liposomes.\textsuperscript{175}

Ahmed and Discher investigated PEO-\textit{b}-PLA and PEO-\textit{b}-PCL polymersomes and found circulation times of several days.\textsuperscript{176} The hydrolysis of the PLA (PCL) resulted in poration of the membrane within several hours of \textit{in vivo} injection for a gradual release followed by complete membrane degradation after several days. When studied \textit{in vitro} via phase contrast microscopy it was found that loaded micelles perforated and unloaded their entire payload relatively quickly but still remained stable for several more days. The mean vesicle pore size was determined by fluorescent measurements of mass transport by dextrans of varied radii of gyration and found to be $\sim 5$ nm.

The initial research in thiolytic cleavage of liposomes by Annapragada et al. utilized 55+\% 1,2-distearoyl-sn-glycero-3-phosphatidyethanolamine (DSPE), 40\% cholesterol, and 2-5\% DSPE-\textit{b}-PEG-NH$_2$ where the terminal amine served as a ligand cross-linking site.\textsuperscript{175} Agglomerates were successfully prepared and tested well in aerodynamic nebulizer studies for pulmonary exposure. \textit{In vitro} thiolytic cleavage was accomplished with dithiothreitol (DTT) in Survanta\textsuperscript{®} (an Abbott Laboratories product used to simulate the lung environment), however it was noted that DTT is unsuitable for \textit{in vivo} use. Overall this study gave strong indication that a successful modulated drug delivery method was possible with agglomerated liposomes. Two years after this study Annapragada et al. released an improved thiolytic cleavage method that did not rely on DTT but rather cysteine, a harmless amino acid commonly used in medicine to break down mucus.\textsuperscript{177} Liposomes agglomerates proved incredibly stable, maintaining their size distribution for at least 3 days at 4\degree C and after incubation for 12 h at 37\degree C. Final release profiles were also studied \textit{in vitro} and proved well-controlled.

Extended CPA Applications

Oral bioavailability

A great deal of micelle research has been singularly focused on taking advantage of the EPR effect for tumor targeting, however several groups have had the vision to see other potentially beneficial applications of polymer micelles. In a pair of papers by Sant et al. the preparation, the characterization, and the evaluation
of improving the oral bioavailability of poorly water soluble drugs was described. Many pharmaceutical products have high clinical efficacy and gastrointestinal permeability but poor solubility due to their hydrophobic nature. The traditional means for compatibilizing these chemicals is through surfactant molecules which improve their solubility. Good solubility becomes appreciably more important when the chemical are prepared for oral administration which improves ease of use and patient compliance. Sant el al. designed a amphiphilic micelle which form supramolecular assemblies arising from intermicellar association. They used an oil/water emulsion to embed micelles with progesterone (a poorly water-soluble model drug) and measured its pH-dependent release rate. Upon increasing pH from 1.2 to 7.2 there was a significant increase in progesterone release resulting in near complete drug release.

Polymersomes in vivo studies

Conclusions

Bibliography


APPENDIX B PH SENSITIVE POLYMERS FOR QUANTUM DOT FORSTER RESONANCE ENERGY TRANSFER

Project Objective

This project was intended to synthesize a pH sensitive polymer with a terminal fluorescent tag for coating quantum dots (QD) used for Forster resonance energy transfer (FRET). The ultimate objective being the selective intracellular fluorescence of QDs based on organelle pH. The polymers selected for this reaction, shown in Figure B.1, were poly((ethylene glycol) methyl ether methacrylate) (PEGMEMA) for its water solubility and poly(2-(diethylamino)ethyl methacrylate) (PDEAEMA) for its pH sensitivity.

Experimental Materials

Ethyl-2-bromopropionate (EBP), 2-methyl-2-bromo propionic acid (MBPA), copper(II) chloride, sodium azide (NaN₃), 1,3-diaminopropane, and N,N,N',N''',N''''-pentamethyldiethylenetriamine (PMDETA) were purchased from the Aldrich Chemical Co. and used without further purification. Poly(ethylene glycol) methyl ether methacrylate (PEGMEMA) and 2-(diethylamino)ethyl methacrylate (DEAEMA) were purchased from the Aldrich Chemical Co. and purified with basic alumina flash chromatography and degassed prior to use. Copper(I) chloride (99.999%) was purchased from the Aldrich Chemical Co. and purified with glacial acetic acid, rinsed with absolute alcohol and ethyl ether, dried under vacuum, and stored under nitrogen prior to use. Toluene was purchased from the Fisher Scientific Co. and degassed prior to use.

Synthesis of PEGMEMA-ran-PDEAEMA

To a 50mL round bottom flask equipped with a 24/40 rubber septum and magnetic stir-bar were added PEGMEMA, DEAEMA, CuCl, CuCl₂, and PMDETA. Chemicals were added in molar ratios relative to amount of initiator used; 15 : 35 : 1 : 0.1 : 1.1 respectively. Additionally toluene was added as a solution solvent in a 3 : 1 volume ratio to total monomer content (the complete reaction scheme is shown in Figure B.2). To a separate 50mL round bottom flask equipped with a 24/40 rubber septum and magnetic stir-bar were added MBPA and
Figure B.1 Schematic of poly((ethylene glycol) methyl ether methacrylate) (PEG-MEMA) and poly(2-(diethylamino)ethyl methacrylate) (PDEAEMA).

toluene in a 10 : 1 volume ratio. All samples were prepared in a nitrogen glove box and pressurized with 5 psi argon during reaction. The sealed round bottom was removed from the glove box and placed in an ice bath at constant temperature with continuous stirring. After 5 minutes in the ice bath MBPA was cannula transferred to reaction round bottom using 5 psi argon, resulting in an immediate color change indicating polymerization initiation. Reactions at 0.4°C were found to produce narrow polydispersities (Mw/Mn ∼ 1.2) and have relatively fast kinetics, reaching ∼90% conversion after only 35 minutes. Samples were further dissolved in toluene and passed over activated basic alumina to remove copper catalyst. Aliquots were removed from polymerized samples and dried. The molecular weights and molecular weight distributions of samples were determined via gel permeation chromatography with respect to poly(methyl methacrylate) standards by a Waters 717 autosampler and Waters 515 HPLC system. Sodium azide was then added in a 10 : 1 molar ratio relative to polymer molecular weight to replace the terminal halogen with an active azide. Further purification over activated basic alumina was used to remove excess NaN₃ and sodium salts formed from reaction. Azide terminated polymer was then purged with argon gas and added dropwise to 20x excess 1,3-diaminopropane in 50mL of toluene at 110°C. This produces an amide bond between the diamine and polymer leaving a primary amine available for further reactions. This step must follow the azide addition because the diamine would also react with the terminal halide. While in toluene, the polymer solution was centrifuged for 6min to remove polymer aggregates and to dilute the supernatant. The supernatant was then dried and re-solvated with nanopure water. A second centrifugation was then conducted for 20 minutes to further purify the polymer in the supernatant. The resulting solution was used for the addition of fluorescent dye to the primary amine.
**Fluorescent dye addition**

0.700 mL of polymer solution was pipetted into a clean reaction vial and then an additional 1.30 mL nanopure water was added to bring reaction solution to 2.0 mL. Dye was solvated with 0.500 mL nanopure water and transferred to the reaction vial. The dye cuvette was then rinsed with 0.500 mL of the reaction solution, and 0.500 mL nanopure water which were both added to the reaction vial bringing the total reaction volume to 3 mL. The reaction vial was then wrapped in aluminum foil to limit light exposure and after 45 minutes the sample was moved to refrigerated storage for 1 hour. After 1 hour 1.500 mL of product was dried under an air stream to concentrate it. The resulting fluorescent functionalized polymer was added to quantum dots for pH sensitive fluorescence measurements.