

RESEARCH ARTICLE OPEN ACCESS

Posidonia oceanica Leaves as a Natural Filler for Poly(Butylene Succinate-Co-Adipate) Composites: Characterization and Biodegradation Assessment in Seawater

Chiara Pedrotti¹  | Damiano Rossi¹  | Giovanni Raimondi² | Riccardo Gallorini¹  | Miriam Cappello¹  | Irene Anguillesi¹  | Marco Sandroni¹  | Patrizia Cinelli¹  | Maurizia Seggiani¹ 

¹Department of Civil and Industrial Engineering, University of Pisa, Pisa, Italy | ²Aquarium of Livorno, Livorno, Italy

Correspondence: Damiano Rossi (damiano.rossi@unipi.it)

Received: 14 January 2026 | **Revised:** 24 April 2026 | **Accepted:** 29 April 2026

Keywords: biocomposites | biodegradability | filler | PBSA | poly(butylene succinate-co-adipate) | *Posidonia oceanica* | seawater

ABSTRACT

This study investigates the use of *Posidonia oceanica* (PO) leaves, derived from coastal banquette residues, as a filler in PBSA, aiming to valorise a natural biomass typically treated as waste within a circular materials approach, while reducing production costs and enhancing seawater biodegradation of the polymer matrix. Composites containing 5 and 10 wt.% of PO, with and without micro-talc, were prepared via extrusion and injection molding. Materials were characterized morphologically, chemically, thermally, and mechanically, while biodegradation was assessed in a controlled seawater environment. Thermal analyses confirm that both PO and micro-talc do not compromise PBSA thermal stability, with degradation temperatures remaining at approximately 403°C. Calorimetric results highlighted an increment of crystallization temperature and the appearance of a secondary melting peak at ~79°C, while the main PBSA melting peak remains at ~85°C and overall crystallinity degree is unchanged. Chemical and morphological analyses reveal no chemical interactions and confirm homogeneous filler dispersion with good interfacial adhesion. Mechanical testing shows a moderate stiffening and embrittling effect with the addition of PO, with elastic modulus increasing from 0.32 to 0.51 GPa and elongation at break decreasing from 450% to 325%, while impact toughness is markedly reduced at high filler loadings. Seawater exposure demonstrates accelerated biodegradation for *Posidonia*-filled composites, showing a ~20% mass loss after 4 months and rapid mechanical deterioration. The incorporation of PO fillers constitutes an effective approach to modulate and accelerate degradation under marine conditions.

1 | Introduction

Posidonia oceanica (PO) meadows, which cover approximately 1%–2% of the Mediterranean seabed, play a crucial role in sediment retention and stabilization, attenuating hydrodynamic forces, protecting coastlines from erosion, supporting biodiversity, and sustaining living resources [1–3]. However, these

meadows are highly vulnerable to ocean warming and to human activities, contributing to large-scale deposits of PO along the coasts [4]. In tourist areas, the accumulation of PO leaves (“banquettes”) is an environmental concern and a source of nuisance for tourism, often prompting their removal or relocation [1, 5–7]. At present, no specific regulatory framework governs the management of PO banquettes.

This is an open access article under the terms of the [Creative Commons Attribution](https://creativecommons.org/licenses/by/4.0/) License, which permits use, distribution and reproduction in any medium, provided the original work is properly cited.

© 2026 The Author(s). Journal of Vinyl & Additive Technology published by Wiley Periodicals LLC on behalf of Society of Plastics Engineers.

Highlights

- The influence of *Posidonia oceanica* powder and micro-talc on the thermal and mechanical stability was evaluated.
- Seawater biodegradation of PBSA-*P. oceanica* composites was simulated.
- *P. oceanica* can be effectively employed as a sustainable filler in PBSA.
- The incorporation of *P. oceanica* filler represents an effective strategy to accelerate degradation under marine conditions.

Another major challenge to marine environmental conservation is the increasing accumulation of non-biodegradable, petroleum-derived microplastics, persistent pollutants whose prevalence reflects the ongoing global rise in plastic production [8, 9]. As a result, an estimated 5–13 million tons of plastic waste enter aquatic environments each year [10, 11].

The widespread use and improper disposal of plastics undermine global sustainability goals, making the search for environmentally friendly alternatives increasingly urgent. Bioplastics offer a promising solution, as they can be bio-based, biodegradable, or both. Among these polymers, PLA, PHA, TPS, PBSA, bio-PET, and others are notable examples [12].

PBSA is an aliphatic co-polyester synthesized by polycondensation of two diacids (succinic acid, adipic acid) and a diol (1,4-butanediol). Formerly derived exclusively from fossil resources, these monomers can now be obtained through biorefinery processes using second-generation renewable feedstocks. Incorporating renewable succinic acid reduces the carbon footprint compared with petroleum-based analogues, while achieving superior biodegradability in various environments relative to other biodegradable polymers thanks to its lower crystallinity [13–15]. Combined with its favorable mechanical properties, PBSA shows strong potential to replace conventional plastics such as polypropylene (PP) and poly(ethylene terephthalate) (PET). Nevertheless, its market uptake remains limited, largely due to its comparatively high cost (5–6 EUR/kg) [16, 17].

One strategy to improve the economic competitiveness of PBSA involves the incorporation of low-cost, renewable fillers, which can simultaneously reduce material cost and enable the valorisation of biomass residues that would otherwise be discarded. Several studies have explored the development of PBSA-based composites containing fillers derived from agro-industrial by-products, including residues from wine and coffee production [18] and other lignocellulosic materials. These fillers are generally inexpensive, widely available, biodegradable, and non-abrasive, which also helps reduce tool wear during processing [19].

Among potential lignocellulosic fillers, PO biomass has attracted increasing attention. Different forms of this material, such as egagropili fibers and leaves, have been investigated as reinforcement in a variety of matrices, including cement,

ceramics, and polymeric systems, particularly to produce natural fiber-reinforced plastics (NFRPs) [20, 21].

Several studies have reported that the incorporation of PO can improve the mechanical performance of composite materials. For example, Ferrero et al. [20] observed significant increases in tensile and flexural stiffness when incorporating up to 40 wt.% PO fibers into a bio-based polyethylene matrix. Similarly, Khiari et al. [22] incorporated finely milled PO particles into potato starch/co-polyester films, obtaining promising mechanical properties. Other studies have explored the addition of milled PO leaves to PLA matrices, where increases in stiffness were observed despite reductions in strain at break [23, 24]. Seggiani et al. [24] also reported significant improvements in stiffness and impact resistance in PHA composites reinforced with PO fibers, reaching approximately an improvement of about 80% over the neat polymer.

In parallel, recent research on PBSA has increasingly focused on improving its environmental degradation behavior. Although composting and soil conditions have been identified as the most favorable environment for PBSA biodegradation [15, 22], its behavior in marine environments has also attracted growing attention. Several studies have highlighted the relatively high moisture absorption sensitivity of biocomposites, which is largely influenced by the presence of vegetal fibers and natural fillers, which are hydrophilic and more biodegradable, increasing the adhesion of microorganisms to the composite material and favour biofouling, as reported by Duigou et al. [25] and Brebu [26]. The incorporation of hydrophilic lignocellulosic fillers has been shown to promote water uptake and the formation of microvoids and cracks within the hydrophobic polymer matrix, facilitating fragmentation and biodegradation in aqueous environments [27]. For instance, Viel et al. [28] investigated the seawater degradation of PBSA films under different simulated marine conditions, observing relatively slow degradation rates, with weight loss generally below 20% after 1 year, except for samples buried in sand. Similarly, Strangis et al. [8] demonstrated that incorporating hygroscopic fillers such as wheat bran significantly enhanced the marine biodegradability of PBSA composites by promoting matrix disruption through swelling effects.

Despite the growing interest in PO biomass as a reinforcement material, its potential use as a filler in PBSA matrices remains largely unexplored, particularly with respect to its influence on processability, mechanical performance, and biodegradation behavior in marine environments.

Therefore, the aim of this study is to investigate the use of PO leaves as a lignocellulosic filler in PBSA to reduce production costs of polymer while maintaining adequate processability and mechanical performance and enhancing biodegradation in seawater. Composites containing 5 and 10 wt.% PO leaves were produced via melt extrusion and shaped by injection molding. Parallel composites with the same filler contents were also prepared with the addition of micro-talc to improve extrusion processability. This additive is known for its excellent blending properties, thermal and electrical resistance, chemical inertness, and ability to provide a smooth surface finish [29].

2 | Experimental

2.1 | Materials

Poly(butylene succinate-co-adipate) (PBSA), commercially marketed as BioPBS FD92PM, was supplied by Mitsubishi Chemical Corporation (MCCP, Tokyo, Japan). This compostable, food-contact-grade aliphatic polyester complies with EU Regulation No. 10/2011 and is available in pellet form. PBSA exhibits a melting temperature of 84°C and a density of 1.24 g cm⁻³. The melt flow rate is 4 g (10 min⁻¹), as determined according to ISO 1133 at 190°C under a load of 2.16 kg. The yield stress, tensile strength at break, and elongation at break are 17 MPa, 24 MPa, and 380%, respectively [30].

Biomass waste consisting of PO leaves was collected directly from the shoreline of St. Thomas Bay, Marsaskala (Malta) in December 2024.

To enhance processability during composite extrusion, a small fraction of micro-talc (IMERYS Jetfine 3C) was incorporated. Jetfine 3C is a hydrated magnesium silicate with the lamellar, plate-like crystal structure typical of talc minerals, represented as Mg₃Si₄O₁₀(OH)₂, and characterized by a median of 0.9 μm [31]. Due to its inherent biocompatibility, this additive also serves as an effective reinforcement, nucleating agent, and anti-agglomeration aid for PO filler, thereby improving the performance of rigid biopolymers, bioplastic films, and compostable plastics [32].

2.2 | Material Pre-Treatment

Dried PBSA pellets were further conditioned in a ventilated oven at 45°C for at least 24 h to ensure complete removal of residual moisture.

TABLE 1 | Composition of the prepared formulations.

Sample	Compositions		
	PBSA (wt.%)	Posidonia (wt.%)	Talc (wt.%)
P	100.0	—	—
PP5	95.0	5.0	—
PP10	90.0	10.0	—
PPT05	94.5	5.0	0.5
PPT1	89.0	10.0	1.0
PT05	99.5	—	0.5
PT1	99.0	—	1.0

TABLE 2 | Extrusion temperature to produce the tested composites.

Feed zone	Zone 1	Zone 2	Zone 3	Head zone
~30°C	120°C–140°C	140°C–150°C	155°C–160°C	145°C–155°C

PO leaves were thoroughly washed with fresh water to remove sand, salts, and organic debris. The leaves were then oven-dried at 140°C for 48 h, leading to a weight loss of approximately 74 wt.%. The dried material was manually pre-crushed and subsequently ground using a high-speed knife blender. The resulting fraction (average size ~3 mm) was sieved, and oversized particles, including residual egagropili and impurities, were discarded. The sieved material was then pulverized using an impact mill (Pulverisette 14, Fritsch, Germany) at 20,000 rpm for 1 h, yielding 700 g of fine brown powder. The final powder was stored in a natural-convection oven at 120°C.

Micro-talc was conditioned in a natural convection oven at 120°C for a minimum of 48 h to eliminate residual moisture.

2.3 | Composite Preparation

Two composite formulations were prepared using PBSA (P) as polymer matrix: one containing PO powder alone and the other incorporating both PO powder and micro-talc.

In the first formulation, PO powder was manually mixed with P granules at 5 wt.% and 10 wt.% (PP5 and PP10), immediately after removing the granules from a 120°C preheated oven. In the second formulation, the warm PO powder was blended with dry micro-talc with a weight ratio of 10:1.

The blending was carried out using a high-speed knife blender, yielding a denser, free-flowing powder. This talc-modified powder was then manually mixed with P granules at 5 wt.% and 10 wt.% PO (PPT05 and PPT1).

To investigate the effect of micro-talc, additional P samples containing 0.5 wt.% and 1 wt.% talc were prepared (PT05 and PT1).

Table 1 summarizes the composition and designation of all formulations. All mixtures were immediately transferred to the extruder feed system for composite production, as described in the following section.

2.4 | Composite Production

P granules containing either untreated or talc-treated PO were fed processed in a single-screw laboratory extruder (Brabender Plasti-Corder Lab-Station, Duisburg, Germany) equipped with a gravimetric side feeder. The extruder had a screw diameter of 19 mm and a length-to-diameter ratio of 25, an axle gauge of 16.5 mm and a flight depth of 3.75 mm. The extrusion temperatures applied for each composite formulation are listed in Table 2 and selected based on the information provided in the P datasheet [30], which recommends a processing temperature range of 130°C–155°C.

The extrusion process was conducted at rotational speeds ranging from 60 to 70 rpm, with a torque of 21 ± 2 Nm and a mass flow rate of 3 ± 0.4 kg/h. The resulting filaments were then cooled in a water bath and pelletized using a Procut 3D cutter (Chinchio Sergio Srl, Brescia, Italy). The resulting pellets were dried at 80°C for at least 24 h and stored in vacuum-sealed bags prior to analysis and further injection molding.

2.5 | Lab-Scale Injection Molding

Dog-bone specimens (type 1BA) and Charpy specimens (80 mm × 10 mm × 4 mm) were fabricated via injection molding using a Megatech H22/50 press (Tecnica Duebi Srl, Italy), at an injection temperature of 170°C and a mold temperature of 45°C, applying a back pressure of 6 bar and a final holding pressure of 110 bar. The overall processing time was approximately 20 s.

2.6 | Characterization

To assess the influence of PO powders and micro-talc filler on the thermal stability of P, thermogravimetric analyses (TGA) were performed using an STA 2500 Regulus instrument (Netzsch, Germany). The onset degradation temperature (T_{onset}) was determined as the extrapolated temperature at the beginning of mass loss, while the degradation temperature (T_d) corresponded to the maximum of the differential thermogravimetric (DTG) curve. The residual mass was quantified as the remaining material at 800°C. Approximately 10–15 mg of each sample was placed in a platinum pan and heated from room temperature to 800°C at a rate of 10°C/min under a nitrogen atmosphere (20 mL/min).

The effects of PO powder and talc on P crystallization and melting behavior were investigated by Differential Scanning Calorimetry (DSC6000, PerkinElmer, USA). Samples of about 10–15 mg were sealed in aluminum pans and subjected to a thermal program under nitrogen flow (10 mL/min). The first heating scan, from –60°C to 120°C at 10°C/min, was used to erase the thermal history associated with processing. This was followed by cooling from 120°C to –60°C at 10°C/min, during which the crystallization temperature (T_c) and crystallization enthalpy (ΔH_c) were recorded. A second heating scan to 120°C at 10°C/min was then performed to determine the melting temperature (T_m) and melting enthalpy (ΔH_m).

The blend crystallinity percentage (X_c) was determined by Equation (1):

$$X_c(\%) = \frac{\Delta H_m}{\Delta H_m^0 \cdot f_w} \cdot 100 \quad (1)$$

where $\Delta H_m^0 = 142$ J/g [33] represents the melting enthalpy per gram of the 100% crystalline P polymer, and f_w the weight fraction of P in the blend.

Chemical interactions between the powder/filler and the P matrix were investigated using Fourier transform infrared spectroscopy (FTIR). Spectra were collected with a Cary 630 FTIR Agilent spectrometer (US) across a wavenumber range of 4000–450 cm^{–1} and a resolution of 4 cm^{–1}.

Dynamic mechanical analysis (DMA) was conducted on the gauge section (L_0) of 1BA dog-bone specimens using an Explorer 303 instrument (Netzsch, Germany). The samples were subjected to strain-controlled oscillatory loading at 0.03% strain across a temperature range from –65°C to 50°C, with a heating rate of 2°C/min and a frequency of 1 Hz under a nitrogen atmosphere. DMA measurements were used to determine the glass transition temperature (T_g), identified as the maximum of the damping factor ($\tan \delta$), as well as the storage modulus (E') at –65°C and 20°C. Mean values and standard deviations were calculated from five specimens per formulation to ensure statistical reliability.

Tensile tests were carried out in accordance with ISO 527-2 [34], using a Quasar 10 universal testing machine (Galdabini, Italy) equipped with a 1 kN load cell. The tests were conducted at a crosshead speed of 10 mm/min until specimen failure, measuring the yield stress (σ_y), stress at break (σ_b), and elongation at break (ϵ_b). To determine the Young's modulus (E), additional tests were performed at a lower strain rate of 1 mm/min using an extensometer MFA25 (Mess & Feinwerktechnik, Germany). Mean values and standard deviations were calculated from five specimens per formulation to ensure statistical reliability.

Charpy impact tests were performed in accordance with ISO 179-1 [35] on notched specimens with dimensions of 80 mm × 10 mm × 4 mm, featuring a centrally positioned 45° V-notch with a depth of 2 mm. The tests were conducted using an AMSE HIT 2492 Series impact tester (Torino, Italy) equipped with a 2 J pendulum. Mean values and standard deviations were calculated from five specimens per formulation to ensure statistical reliability.

Samples' morphology was examined using a COXEM EM-30N (Daejeon, Republic of Korea) Scanning Electron Microscope (SEM). Granules were cryo-fractured in liquid nitrogen, and the resulting surfaces were coated with a thin (5–6 nm) gold layer through an Edwards S150B Sputter Coater (UK).

The particle volume and number distributions of PO and micro-talc powders were determined using a Mastersizer 3000+ Malvern Panalytical (UK) under dry conditions, with air pressure and feeder vibration parameters adjusted to ensure optimal dispersion during the test. The median particle size (d_{50}) and the span, defined as the distance between the 10th and 90th percentile points, normalized by d_{50} , were determined. Three measurements for each material were evaluated.

A long-term seawater biodegradation test was carried out on dog-bone specimens of P and PP10 at the Aquarium of Livorno (Italy) to assess the influence of PO powder on the degradation behavior of P. The selection of PP10 was motivated by the fact that it represents the formulation with the highest filler content. Moreover, based on the material characterization results, no significant differences in thermo-mechanical properties were observed compared to PPT1. Additionally, PP10 was chosen due to the limited available space in the seawater tank, which constrained the number of samples that could be tested. The samples were buried at approximately 4 cm depth within natural calcareous sand placed in a tank continuously supplied with fresh seawater. The seawater was collected at 2 m depth and 50 m offshore, then filtered and UV-sterilized before entering the system (Figure 1). The sediment layer, about 6 cm

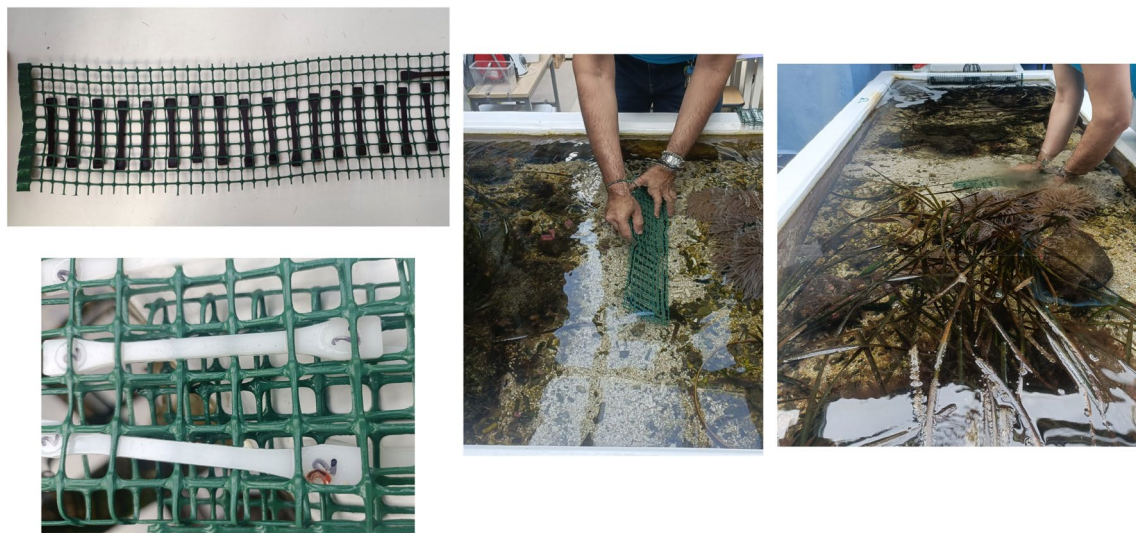


FIGURE 1 | From left to right: Arrangement of PP10 (top) and P (bottom) samples within the non-biodegradable net, later covered with sand in the seawater tank (center and right), where they are surrounded by microorganisms and algae.

thick, progressively became colonized by microorganisms and benthic fauna, including ophiuroids, polychaetes, and amphipods, feeding on food residues provided to the tank's resident organisms (sea urchins, anemones, sea stars, mollusks, and crustaceans). Due to the continuous exchange with the external marine environment, pH, oxygen concentration, and salinity remained comparable to natural seawater, while water temperature was primarily influenced by the indoor conditions of the aquarium facility.

The experiment lasted 4 months, during which environmental parameters were monitored daily. Temperature exhibited natural fluctuations over the course of the study, whereas pH, dissolved oxygen, and salinity remained relatively stable, with mean values of 7.8, 4.1, and 33.5 g/L, respectively. Oxygen was supplied naturally through algal photosynthesis and exchange with the surrounding environment, thereby eliminating the need for external aeration. Before immersion, all specimens were weighed.

At monthly intervals, four specimens of each material were retrieved, cleaned to remove adhering particles and surface moisture using absorbent paper, oven-dried at 58°C for 4 h, and reweighed. Weight loss relative to the initial mass was calculated and reported as mean values. The evolution of mechanical performance was also assessed through tensile testing of the retrieved specimens at each sampling time, performed at a cross-head speed of 1 mm/min until failure, with both mean values and standard deviations recorded.

3 | Results and Discussion

3.1 | Thermal Analysis

Figure 2 presents the TGA and DTG curves with the composites grouped based on the weight percentage of PO powder (5 wt.% in Figure 2a–c, and 10 wt.% Figure 2b–d), while Table 3 reports the corresponding values of T_{onset} , T_d , and residues at 800°C.

PO exhibits three main degradation peaks at 91.5°C, 316.3°C, and 671.0°C. These peaks correspond to moisture loss (91.5°C),

decomposition of the holocellulosic fraction (316.3°C), and combustion/volatilization of the char residue (671.0°C) [36–38]. The degradation extends across a broad temperature interval due to the heterogeneous thermal decomposition of lignin, which typically occurs over 190°C–900°C [39]. The high residual mass (47.0 wt.%) is associated with the substantial amounts of calcium carbonate (CaCO_3) and some metals residues absorbed in the biomass [39]. In this regard, Bertini et al. [40] reported the presence of trace heavy metals in PO leaves, including Pb, Cr, Cu, and Ni, with concentrations of approximately 2, 4, <2, 4, and <1 mg/kg, respectively, in samples collected from the Murelle site.

In the case of talc, no relevant degradation phenomena were observed within the selected temperature range. The material exhibited a high residual mass at 800°C (98.5%), confirming its notable thermal stability even at elevated temperatures.

The neat P shows thermal stability up to approximately 300°C, followed by a single, well-defined degradation stage with a maximum peak near 400°C. At 800°C, only a negligible residue of 1.75 wt.% was detected, attributable to the fixed carbon and inorganic matter, consistent with the observations reported by Seggiani et al. [41].

The introduction of micro-talc and PO powder does not markedly influence the characteristic degradation behavior of the polymer matrix, with T_{onset} and T_d varying by only 4°C–5°C compared to the values of neat P ($T_{\text{onset}} = 372.9^\circ\text{C}$, $T_d = 403.4^\circ\text{C}$). The degradation profiles of all composites display consistent, overlapping, and well-defined mass-loss steps, which could indicate a homogeneous distribution of the fillers within the matrix across the investigated loadings and the similarity of the T_d of P and PO particles. Regarding the residues, their amounts increase with the rising weight fraction of PO and micro-talc, consistent with the high intrinsic residue of these two components (47.0 wt.% and 98.5 wt.%, respectively).

The degradation processes remain outside the temperature window used in melt extrusion (Table 2), confirming the suitability of the composites for manufacturing without the occurrence of thermal degradation.

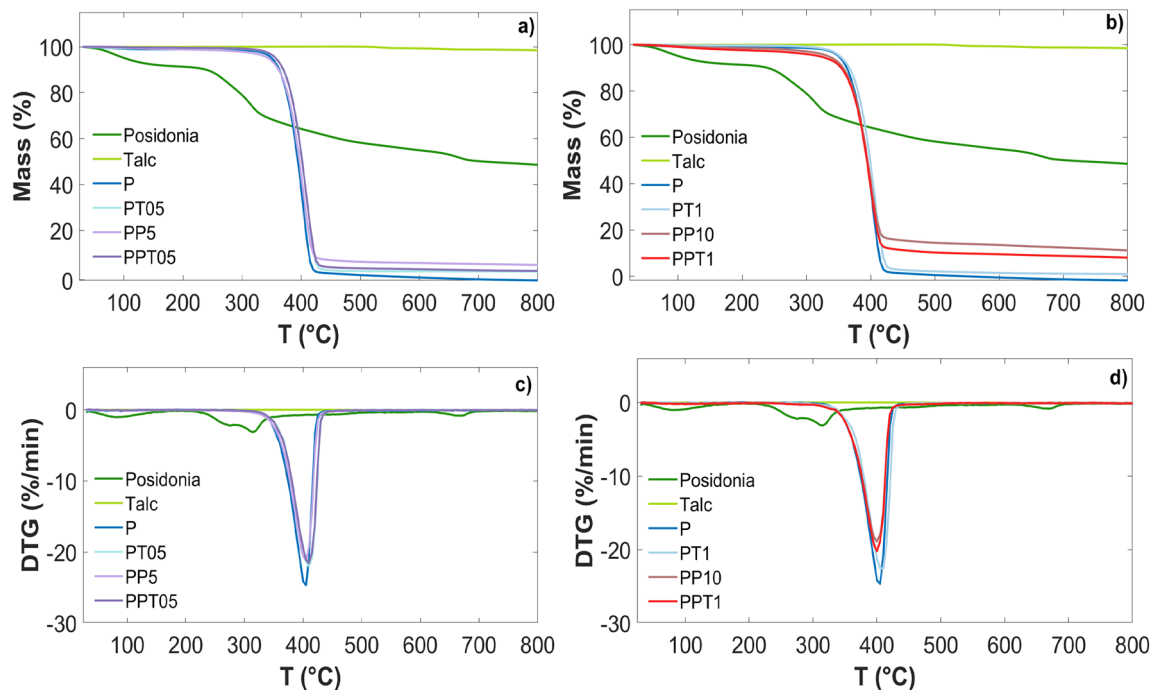


FIGURE 2 | TGA and DTG curves of PO, Talc, P, PT05, PP5, PPT05 (a, c) and PO, Talc, P, PT1, PP10, and PPT1 (b, d).

TABLE 3 | T_{onset} , T_d , and residue from TGA and DTG curves.

SAMPLE	T_{onset} (°C)	T_d (°C)	Residue (%)
Posidonia	67.0	91.5	47.0
	264.7	316.3	
	659.7	671.0	
Talc	503.6	543.2	98.5
	608.4	644.7	
P	372.9	403.4	1.75
PT05	378.2	408.2	2.38
PT1	377.4	408.6	2.59
PP5	373.4	404.3	4.95
PPT05	371.7	400.4	6.64
PP10	369.8	399.2	11.2
PPT1	370.6	400.0	8.06

Figure 3 and Table 4 report the DSC profiles of the various samples, enabling the identification of cooling and melting temperatures and the corresponding enthalpy variations.

Figure 3a highlights the distinct cooling behavior of the neat P matrix, which is characterized by a very low crystallization temperature ($T_c = 15.4^\circ\text{C}$). This behavior is attributed to the presence of succinate-co-butylene adipate segments, which increase chain flexibility and thereby facilitate rapid chain rearrangement and the formation of well-developed crystalline structures during crystallization [42]. The incorporation of micro-talc and PO powder shifted the crystallization process to higher temperatures ($T_c \approx 53.0^\circ\text{C}$) due to the nucleating effect of the fillers,

which also resulted in a slight increase in the crystallization enthalpy for all composites, from 42Jg^{-1} for neat P to approximately 44Jg^{-1} for the filled systems. This behavior is consistent with previous studies on talc-filled systems reported by Várdai et al. [43] and Castillo et al. [44].

In particular, talc's nucleating efficiency is enhanced by its lamellar morphology, consisting of small, thin, and irregular platelets that facilitate the formation of heterogeneous nucleation sites within the polymer matrix, especially along the flow direction during injection molding. Likewise, the nucleating activity of PO, and of lignocellulosic fillers more broadly, has been highlighted in DSC analyses by Cappello et al. [16] and Strangis et al. [8].

This behavior is further confirmed during the second heating cycle (Figure 3b), where a second melting peak appears in addition to the single melting peak of neat P. A broad endothermic peak with a maximum at 85.5°C is observed, corresponding to the melting point of the material and resulting in a crystallinity degree (χ_c) of 33.3%. Comparable results have been reported in other studies employing the same type of material [16, 24].

As expected, the melting thermograms of the composites indicate that the incorporation of micro-talc and PO powder alters the crystallization behavior of P, as evidenced by the appearance of a low-temperature shoulder at approximately 79°C , while the characteristic broad melting peak of neat P is retained at around 85°C . This phenomenon can be attributed to the melting of newly formed, smaller crystalline domains within the P matrix, induced by the nucleating action of the fillers and possibly enhanced by the relatively high cooling/heating rates employed during the DSC measurements. An additional consequence of the formation of smaller crystalline domains is a slight

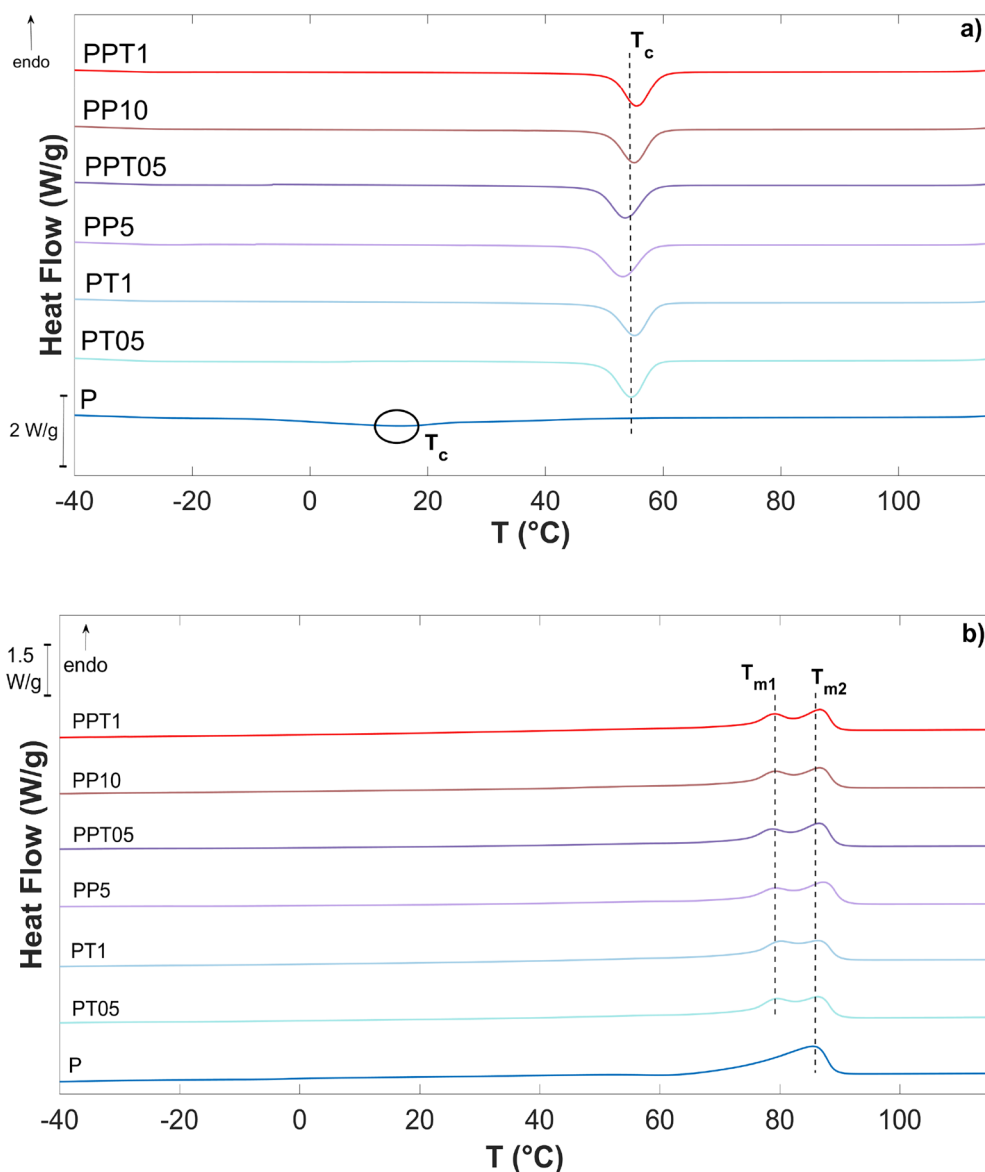


FIGURE 3 | Thermogram of the (a) cooling and (b) second heating cycle of the composites obtained in DSC analysis.

reduction in the melting enthalpy of the composites compared to neat P (47.3 J g^{-1}), with values ranging from 43 J g^{-1} for PT05 to 39 J g^{-1} for PPT1. This reduction suggests a decreased ability to form highly ordered or branched crystalline lattices, thereby requiring less energy for melting. Nevertheless, no significant differences in the degree of crystallinity (χ_c) are observed, with values ranging from 31.0% to 33.3%, irrespective of filler type or content.

3.2 | Chemical Analysis (FTIR)

Figure 4 reports the FTIR spectra of the raw materials and the composites (Figure 4a), along with a magnified view of the region between 1500 and 450 cm^{-1} (Figure 4b).

The spectra of PO exhibit weak and broad bands at 3332 , 1615 , 1049 , and 872 cm^{-1} , which can be attributed to functional groups associated with cellulose, hemicelluloses, and lignin, representing the main structural constituents of the seagrass [45].

In particular, the broad band at 3332 cm^{-1} is attributed to the O—H stretching vibration of phenolic or hydroxyl groups in lignin and hemicellulose [46]. The sharp peak at 1615 cm^{-1} can be attributed to the asymmetric stretching of the —COO^- group. The broad peak at 1424 cm^{-1} corresponds to the deformation vibrations of C—OH and phenolic O—H groups, while the band at 1049 cm^{-1} is associated with the C—O stretching vibration typically observed in carboxylic acids and ethers [45, 47, 48]. For micro-talc, the peak at 3675 cm^{-1} is associated with the O—H stretching vibration of the octahedral Mg—O—H unit. The Mg—O—H unit also gives rise to a peak at 665 cm^{-1} , corresponding to its bending [49]. In addition, the Si—O stretching vibration is observed at 980 cm^{-1} [50].

For P, the characteristic absorption peak at 1710 cm^{-1} is attributed to the C=O stretching vibration of ester groups, whereas the band at 1149 cm^{-1} arises from the stretching of —C—O—C— bonds in the ester linkages. Additionally, the peak at 2948 cm^{-1} is associated with the —CH_3 stretching vibration [33, 51].

The spectra of the P composites exhibited only the characteristic peaks of neat P, indicating the absence of chemical interactions between the matrix and the fillers. This behavior can

TABLE 4 | Cooling temperatures (T_c), crystallization enthalpy (ΔH_c), melting temperatures (T_m) and relative enthalpy (ΔH_m) with crystallinity degree (χ_c) of the composites.

Sample	T_c (°C)	ΔH_c (J/g)	T_{m1} (°C)	T_{m2} (°C)	ΔH_m (J/g)	χ_c (%)
P	15.4	42.3	—	85.5	49.1	33.3
PT05	54.5	44.8	79.4	86.2	43.5	30.8
PT1	55.0	45.0	80.2	86.2	42.7	30.4
PP5	53.2	43.4	79.2	87.2	44.8	33.2
PPT05	53.6	43.7	78.7	86.5	41.2	30.7
PP10	55.2	41.3	79.2	85.5	41.0	31.3
PPT1	55.4	43.6	79.3	86.6	39.2	31.0

be attributed to the low filler content and its homogeneous dispersion within the polymer matrix, which together prevent any significant contribution to molecular interactions. In addition, it is known that talc is chemically inert; therefore, no chemical interactions are expected or detectable in the FTIR spectra. Its role during extrusion is purely physical, to reduce friction between the PO particles and the polymer matrix. Similarly, untreated PO does not exhibit any chemical interaction with the P matrix, underlining that the interaction between PO and the polymer matrix is therefore primarily physical. Consequently, no PO-specific peaks are observed in the FTIR spectra. This is not unexpected and can be attributed to the small particle size (predominantly around 1 μm), which leads to their effective incorporation within the polymer matrix, thereby preventing any distinct spectral contribution from being detected. This interpretation is further supported by SEM images (Figure 10), which show a well-integrated filler/matrix interface, suggesting good interfacial adhesion and rather good compatibility.

This result is consistent with the thermal analysis, which revealed no variations in the main thermal properties.

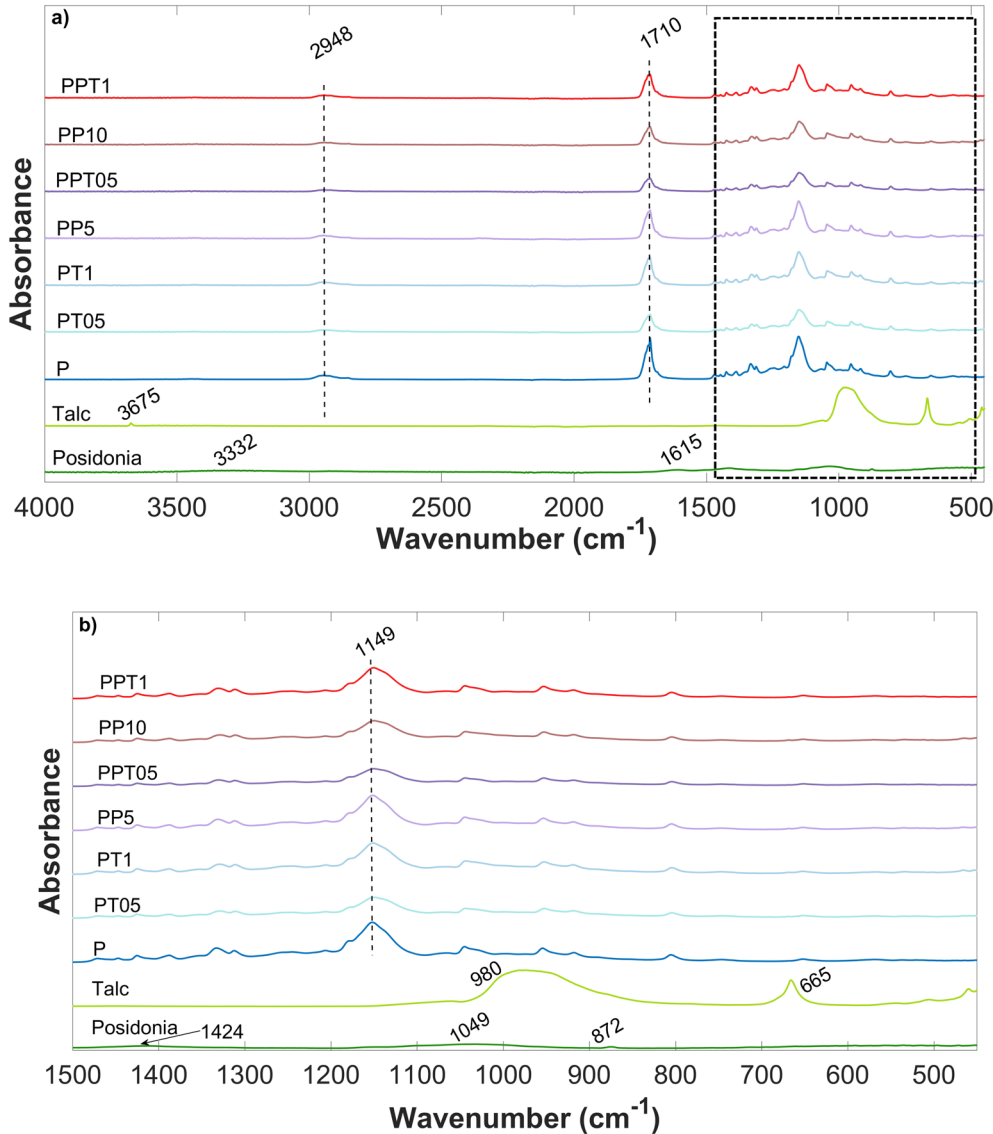


FIGURE 4 | FTIR spectra of the composites (a), with an enlarged view of the wavenumber region 1500–500 cm^{-1} (b).

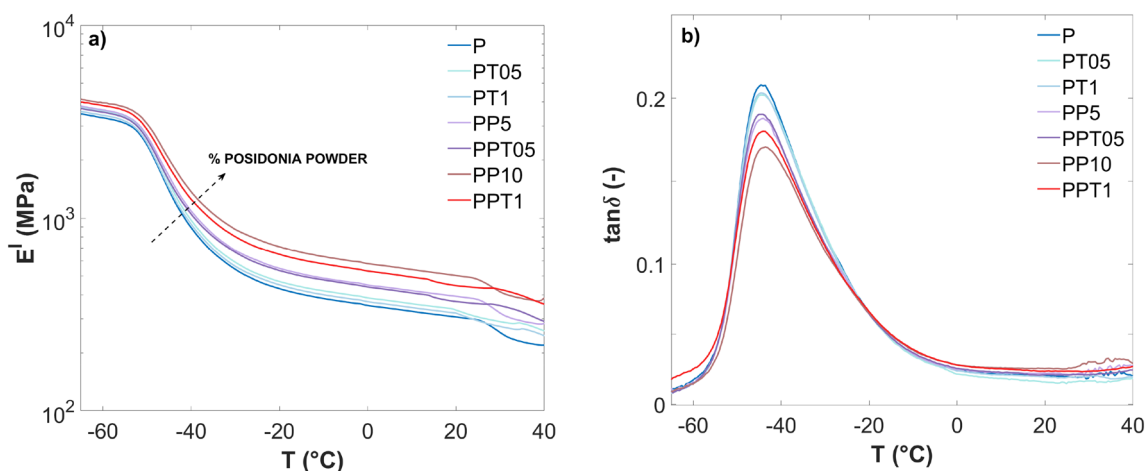


FIGURE 5 | Average storage modulus E' (a) and damping factor $\tan \delta$ (b) of the composites as a function of temperature, measured from -65°C to 40°C at a heating rate of $2^{\circ}\text{C}/\text{min}$ and a frequency of 1 Hz.

3.3 | Mechanical Analysis

Figure 5 and Table 5 present the DMA results (T_g and E') of the P/PO composites between -65°C and 40°C .

At low temperatures ($< -55^{\circ}\text{C}$), the P matrix and all composites remain in the glassy region, characterized by high E' values resulting from restricted molecular motion. As the materials approach their T_g , E' shows a pronounced decline, indicating the transition toward a prevailing rubbery plateau (from -40°C to 40°C), with each sample displaying an approximate one-order-of-magnitude reduction. For instance, the P sample decreases from 3.44 to 0.30 GPa.

Increasing the weight percentage of PO powder results in an upward shift of the E' curve, with mean values increasing from 3.44 GPa for P, to 3.77 GPa for PP5, and finally to 4.08 GPa for PP10. This behavior can be attributed to the reduced mobility of polymer chains caused by interfacial interactions between the filler and the polymer matrix, a mechanism also described by Qi et al. [52], who reported comparable effects when incorporating inorganic attapulgite nanoparticles into a PBSA matrix. These findings further support the observations of Khiari et al. [22], emphasizing the reinforcing role of lignocellulosic fibers and inorganic particles, which are widely employed in composite materials due to their low cost and environmentally sustainable nature. These fillers act as reinforcing agents by restricting the mobility of polymer chains in their vicinity and hindering chain disentanglement during deformation, thereby limiting chain sliding and increasing the composite's stiffness. This effect is evidenced by an increase in elastic modulus accompanied by a reduction in elongation at break.

The effect is slightly attenuated in the presence of talc (PPT05: 3.71 GPa; PPT1: 4.04 GPa), likely due to the specific arrangement of the organic–inorganic interface within the composite [53] and to the lamellar morphology of talc. In particular, talc platelets tend to align along the flow direction during injection molding, thereby not significantly hindering polymer chain mobility [43]. This interpretation is further supported by the comparable E' values measured for PT05 and PT1, equal to 3.63 GPa and 3.53 GPa, respectively, at -65°C .

TABLE 5 | T_g and average value of E' at -65°C and 20°C of the composites.

Sample	T_g ($^{\circ}\text{C}$)	E' (GPa)	
		-65°C	20°C
P	-44.63	3.44 ± 0.03	0.30 ± 0.01
PT05	-44.64	3.63 ± 0.18	0.31 ± 0.02
PT1	-44.60	3.53 ± 0.06	0.31 ± 0.01
PP5	-44.61	3.77 ± 0.05	0.38 ± 0.01
PPT05	-44.63	3.71 ± 0.04	0.37 ± 0.01
PP10	-44.12	4.08 ± 0.09	0.49 ± 0.01
PPT1	-43.61	4.04 ± 0.06	0.45 ± 0.01

The same trend observed at low temperatures is preserved in the rubbery plateau region, where an increase in E' is recorded with increasing PO content, rising from an average value of 0.30 GPa for neat P to 0.38 GPa for PP5 and 0.49 GPa for PP10. The addition of talc slightly mitigates the embrittlement of the composites, owing to its lamellar morphology and alignment along the processing direction, which contributes to a more balanced viscoelastic response. Furthermore, the homogeneous dispersion of both fillers within the polymer matrix prevents any broadening of the $\tan \delta$ curve and maintains a constant peak value, corresponding to a T_g of approximately -44.60°C [52].

Figure 6 presents the experimental data obtained from the tensile tests. As shown in Figure 6e P exhibits the typical behavior of a highly ductile polymer, characterized by relatively low values of σ_y and σ_b (16.02 MPa and 25.33 MPa, respectively), but an exceptionally large plastic deformation (450.27%), which is in accordance also with the technical data sheet of the material [30]. The deformation initiates with the necking of the 1BA specimen once the yielding stress is reached, a phenomenon facilitated by the high chain mobility above the glass transition temperature (-44°C) [54].

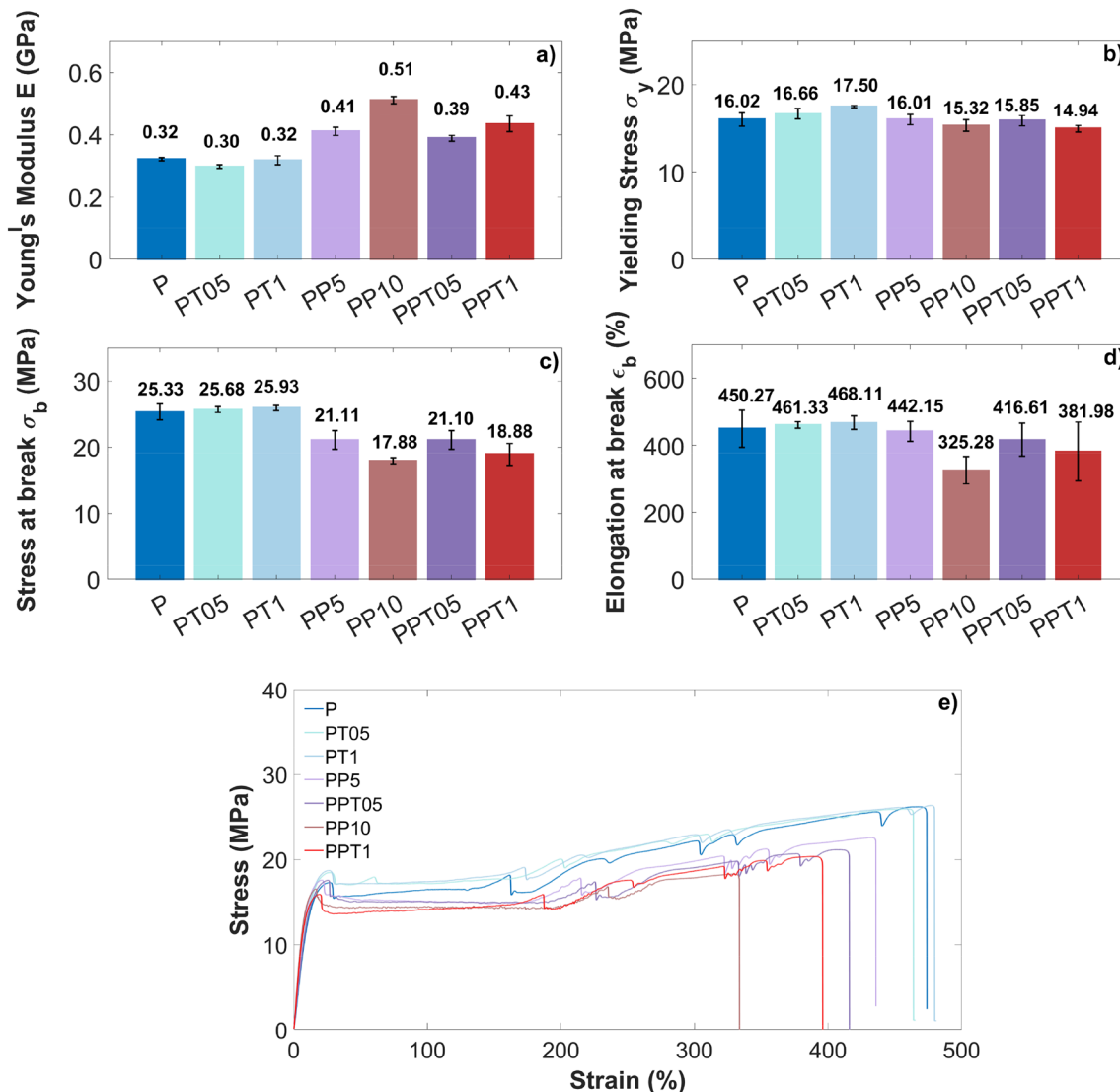


FIGURE 6 | Average values with respective standard deviation of E (Figure 6a), σ_y (Figure 6b), σ_b (Figure 6c), and ϵ_b (Figure 6d) of the composites; average stress–strain curves of the composites (Figure 6e).

Upon incorporation of micro-talc, the tensile behaviour of the talc/P composite remains largely comparable to that of neat P. Similar or slightly enhanced values of E , σ_y , σ_b , and ϵ_b are observed. This response can be attributed to the lamellar morphology of talc and its preferential alignment along the injection-moulding flow direction, as also reported by Post et al. [55] for polypropylene-based composite. Since the primary objective of talc addition was to improve extrusion processability, the preservation of mechanical performance indicates that the filler does not negatively affect the composite's structural integrity.

The incorporation of PO powder leads to a noticeable reduction in σ_y , σ_b , and ϵ_b for PP5 and PP10 samples. This decline becomes more pronounced with increasing filler content, generally reaching the lowest values in the PP10 sample, which recorded 15.32 MPa for σ_y , 17.88 MPa for σ_b , and 325.28% for ϵ_b . Conversely, a slight increase in E is observed with increasing filler loading, indicating that the lignocellulosic powder acts as a stiffening and mildly embrittling agent. This trend is consistent with the results obtained from DMA analysis and with the findings reported by Strangis et al. [8], who observed an increase in the elastic modulus with increasing wt.%

of the lignocellulosic filler, accompanied by a non-significant decrease in both stress at break and elongation at break. Nevertheless, owing to the good dispersion of the filler and its satisfactory interfacial adhesion with the polymer matrix, the embrittlement effect remains limited.

Finally, consistent with the above discussion, the tensile properties of the ternary composites PPT05 and PPT1 are not significantly affected by the addition of micro-talc when compared to PP5 and PP10. This behavior is likely related to the lamellar morphology of talc, which allows it to orient along the processing direction, thereby preserving mechanical properties within the standard deviation range of the corresponding binary composites.

Figure 7 presents the impact energy values obtained for the different samples, highlighting the significant effect of PO powders and micro-talc within the polymer matrix.

P exhibits optimal impact energy absorption (53.42 kJ/m²), as demonstrated by Post et al. [55]. This enhanced toughness is

primarily associated with its relatively low elastic modulus, which allows significant deformation under stress, and the high segmental mobility of its polymer chains. Since the mechanical tests were performed at room temperature, well above the glass transition temperature (T_g) of P, the material displays a ductile response, enabling efficient dissipation of impact energy.

The incorporation of micro-talc and PO powder results in a marked reduction in impact energy. In particular, the absorbed energy is reduced to approximately half for the PPT05 and PPT1 samples. In these composites, the lamellar morphology of the filler does not contribute to energy dissipation but instead acts as an embrittling factor.

A substantial reduction in impact energy is also observed in composites containing only PO powder, with values of 9.61 kJ m^{-2} for PP5 and 6.88 kJ m^{-2} for PP10. As previously highlighted in the tensile test analysis, PO powder primarily behaves as a stiffening filler with a mild embrittling effect on the P matrix, which likely contributes to the observed decrease in impact resistance.

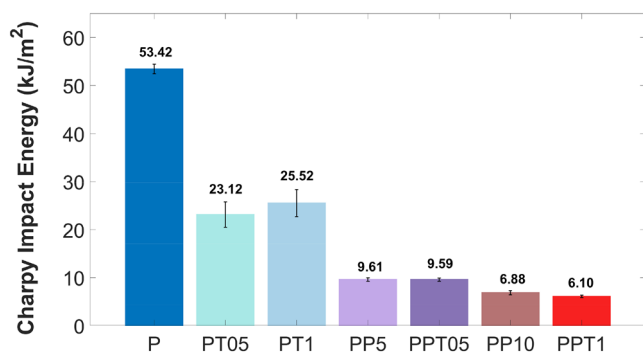


FIGURE 7 | Average Charpy impact energy values with corresponding standard deviations for the composites.

When micro-talc and PO powder are combined, the resulting impact energy values remain comparable to those of the composites filled exclusively with PO, indicating that the simultaneous presence of both fillers does not mitigate the embrittling tendency but instead maintains a similar mechanical response.

Overall, the reduction in impact energy can be attributed to the relatively high aspect ratio of both fillers, which are non-spherical in shape and therefore act as stress concentrators within the polymer matrix. As the particle edges become sharper, the local stress concentration intensifies, leading to premature failure, as demonstrated by Ananthapadmanabha et al. [56] through a direct comparison between spherical calcium carbonate (CaCO_3) fillers and platelet-shaped talc. Furthermore, the tendency of the fillers to form aggregates, as evidenced in Figure 9c, further contributes to the deterioration of impact strength by generating additional stress concentration sites within the composites.

Based on the results obtained thus far, a model (Figure 8) describing the interactions among the composite components can be proposed. As reported in the sub-paragraph “Composite Preparation” (Section 2.5), micro-talc and PO powder were premixed prior to extrusion to enhance processability and minimize PO particle agglomeration. At this stage, talc lamellae shield and cover the PO fibers, promoting the physical separation of PO-talc particles within the polymer matrix.

Talc exhibits a layered crystalline structure consisting of an octahedral magnesium brucite sheet sandwiched between two tetrahedral silica layers. Its surface is intrinsically hydrophobic, owing to the absence of hydrogen-bonding sites, thereby limiting moisture-induced agglomeration, reducing dispersion in aqueous media, and conferring chemical inertness [57, 58]. Furthermore, the talc layers are held together by weak van der Waals interactions, which enable interlayer sliding during composite extrusion and deformation. This

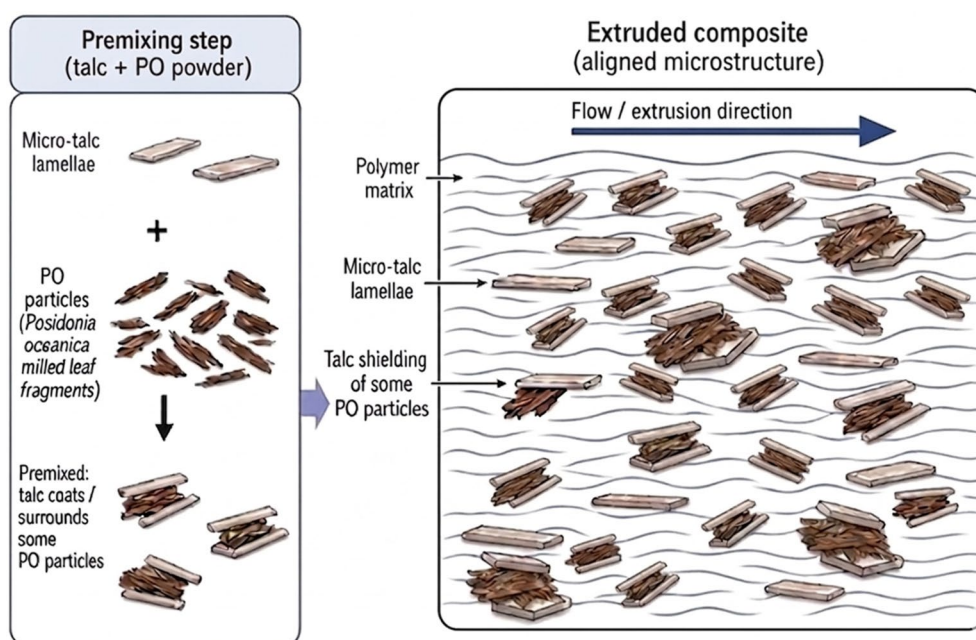


FIGURE 8 | Schematic model of interaction among PO, micro-talc and polymer matrix Morphological analysis.

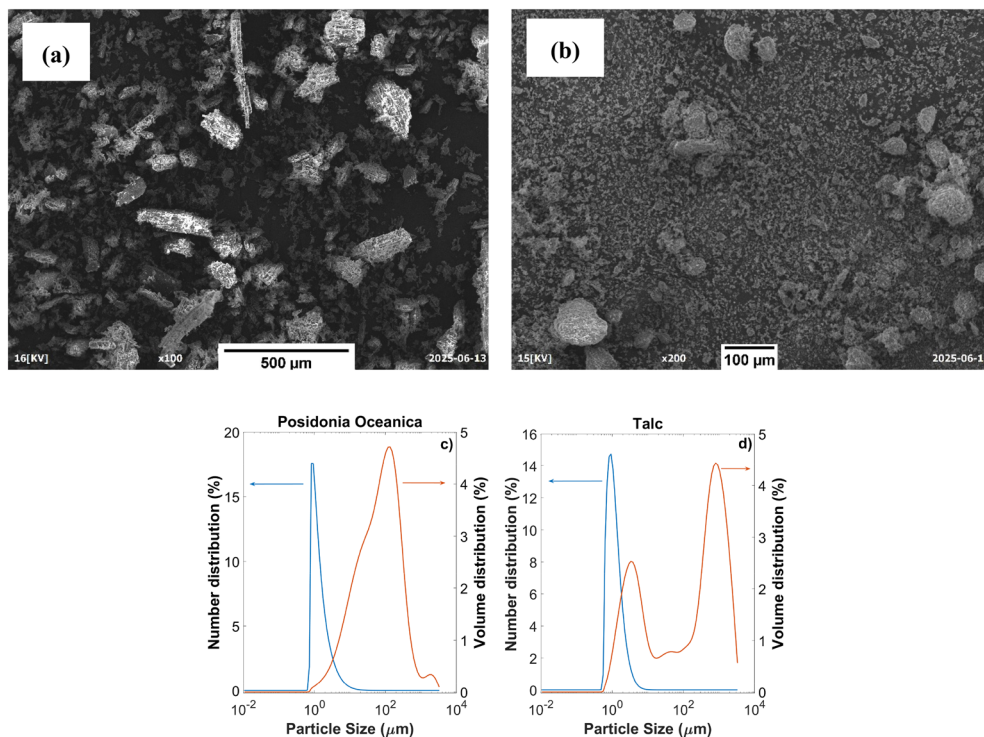


FIGURE 9 | SEM images with magnification of (a) $\times 100$ of PO and (b) $\times 200$ of micro-talc and number particle distribution of (c) PO powders and (d) Talc.

behavior facilitates chain alignment along the flow direction and reduces friction between the PO-talc particles and the polymer matrix. Overall, this mechanism is consistent with the absence of strong chemical interactions among the three components of the composite.

Figure 9 presents the SEM images of PO and talc powder, together with their corresponding number distributions and the value of the particle size distribution d50 and span.

At $100\times$ magnification, the PO powder exhibits a clear non-homogeneous morphology, as further evidenced by the relatively high span value (4.21), which reflects the pronounced particle polydispersity typically associated with mechanically ground materials. Despite this variability, the median particle size (d50) remains at $1.28\mu\text{m}$. In the case of talc, the micrographs reveal the presence of aggregates, most likely induced by moisture absorption, leading the volumetric span at 4.98; nevertheless, its number-based particle size distribution is characterized by a d50 of $1.04\mu\text{m}$.

Figure 10 shows the fracture surfaces of neat P and the corresponding P-based composites. At a magnification of $\times 2000$, no significant morphological differences are observed in the brittle fracture surfaces of the P/talc composites, indicating a homogeneous dispersion of the filler within the polymer matrix. In the case of talc incorporation, the identification of voids is particularly challenging, owing to the good interfacial adhesion between the platelet-shaped talc particles and the polymer matrix, as well as to their large contact surface area [59].

Upon incorporation of PO powder in the PP5 and PP10 samples (Figure 10d,f), a generally homogeneous filler distribution is

also observed. However, as indicated by the particle size distribution analysis, the presence of relatively large particles cannot be excluded. This is clearly evidenced in Figure 10d, where the characteristic porous morphology of PO can be distinguished, with particle diameters of approximately $20\mu\text{m}$. As a result, the fracture surfaces of these composites appear more irregular compared to neat P, likely due to the presence of larger particles acting as stress concentration sites.

In ternary composite PPT05 and PPT1 (Figure 10e,g), both PO and talc powders appeared well dispersed within the polymer matrix, without visible agglomeration or voids. The homogeneous distribution of the fillers, together with the absence of defects, indicates good compatibility and efficient processing conditions.

In conclusion, composites containing higher amounts of PO do not exhibit clearly visible powder particles, likely due to their particle size distribution. The number-based distribution indicates an average particle size of approximately $1\mu\text{m}$, without significant peaks at larger sizes, whereas the volume-based distribution shows a broad peak around $10^2\mu\text{m}$. This suggests that large individual particles are generally absent; however, occasional agglomerates may still be present, possibly as a result of moisture-induced aggregation.

This morphological evidence provides further support for the tensile test results, as the mechanical properties of these composites are comparable to those of the neat P matrix. Nevertheless, similar to the PP5 and PP10 systems, the occasional presence of larger PO particles with a porous morphology cannot be entirely excluded; these particles may induce local heterogeneities in the polymer matrix and function as stress concentration points, potentially compromising the mechanical integrity of the composite.

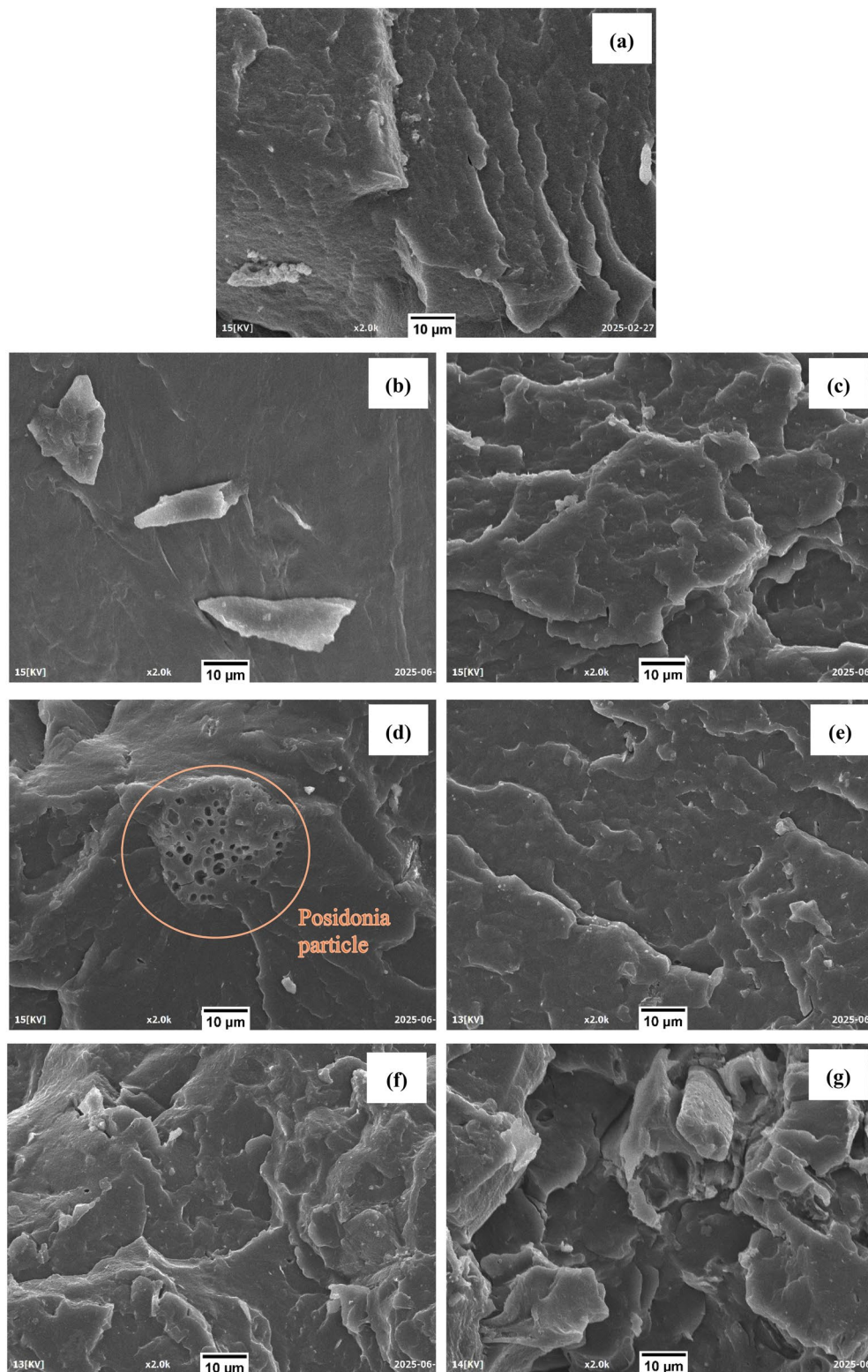


FIGURE 10 | SEM images of $\times 2000$ magnification of (a) P, (b) PT05, (c) PT1, (d) PP5, (e) PPT05, (f) PP10, and (g) PPT1.

In conclusion, the SEM analysis supports the evidence that the PPT1 formulation may represent the threshold beyond which the mechanical properties deteriorate. This behaviour can be attributed to the excessive fracture surface generated by the non-uniform dispersion of the filler within the polymer matrix.

3.4 | Biodegradation Test

This section reports the mechanical properties and the biodegradation progress in seawater of the dog-bone specimens of P and PP10, respectively.

Figure 11 illustrates the condition of P and PP10 specimens over the months of immersion in seawater tanks. It can be immediately observed that P specimens remain intact, showing no defects throughout the examination period. In contrast, PP10 exhibits signs of degradation after only 2 months, with some specimens partially degraded only after 3 months.

This phenomenon can be attributed to the incorporation of powder as a filler within the polymer matrix, which enhances the biodegradation rate of P by promoting composite fragmentation and thereby initiating a mechanically driven degradation process. The hydrophilic nature of the PO powder likely contributes to this effect, as the swelling of the flakes in the aqueous medium facilitates specimen fragmentation, resulting in an increased surface area available for microbial colonization and subsequent degradation, as it is possible to observe from the SEM image in the zoomed Figure 11c. Similar effects have been reported with the incorporation of bran into P-based composites, as described by Seggiani et al. [60].

Figure 12a shows the percentage weight loss over time, enabling quantification of the effect of PO particles on specimen mass loss and comparison with neat P. PP10 samples degrade faster than P samples, which exhibit a slower, linear trend. This accelerated degradation is attributed to the hydrophilic nature of PO powder, which absorbs water, promotes material fragmentation, and enhances the activity of microorganisms and benthic organisms. Similar results are also reported in Strangis et al. [8], when comparing the same exposure period, P shows a weight loss of less than 5% after 4 months, whereas the composite material reaches approximately 20% weight loss.

Consequently, the mechanical properties of the composites are drastically reduced even after 2 months, with the material transitioning from the typical ductile behavior of P to a brittle one. While P samples maintain their ductile behavior throughout the entire testing period, showing no significant influence of the seawater environment. In fact, the P samples reach optimal performance after approximately 3 months of seawater exposure.

In contrast, PP10 exhibits a markedly different response. Seawater exposure induces pronounced embrittlement, with a drastic reduction in elongation at break (ϵ_b) from 325.28% to 19.04% within 3 months, and the disappearance of yield stress (σ_y) after only 2 months. By the fourth month, PP10 samples are no longer suitable for mechanical testing due to degradation-induced defects (Figure 12b–e).

As observed, the biodegradation of P in seawater is less favorable than in compost. This difference may be attributed to intrinsic factors such as the polymer's breaking point, chemical structure, and degree of polymerization, as well as to environmental conditions including temperature, pH, and the presence of catalysts or enzymes [61, 62]. In fact, hydrolytic degradation constitutes a critical pathway for the breakdown of aliphatic polyesters such as P.

Among these factors, crystallinity plays a central role in determining the polymer's susceptibility to degradation. With a χ_c of approximately 33%, the crystalline regions are densely packed, limiting water penetration and consequently slowing both hydrolytic and microbial processes [63]. Moreover, elevated temperatures can markedly accelerate hydrolytic reactions even

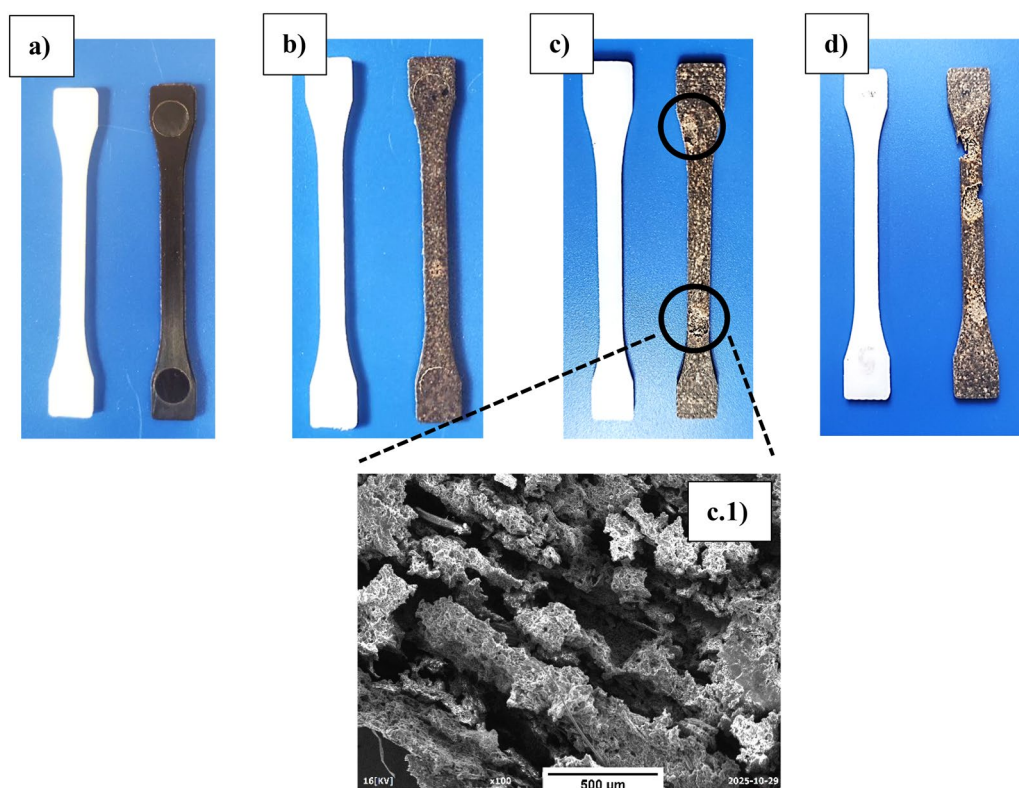


FIGURE 11 | Dog-bone specimen for the biodegradation test in sea water after month: 0 (a), 2 (b), 3 (c) and 4 (d). In (c.1), a detail of the dog-bone specimen after 3 months is shown by an SEM micrograph at 100× magnification.

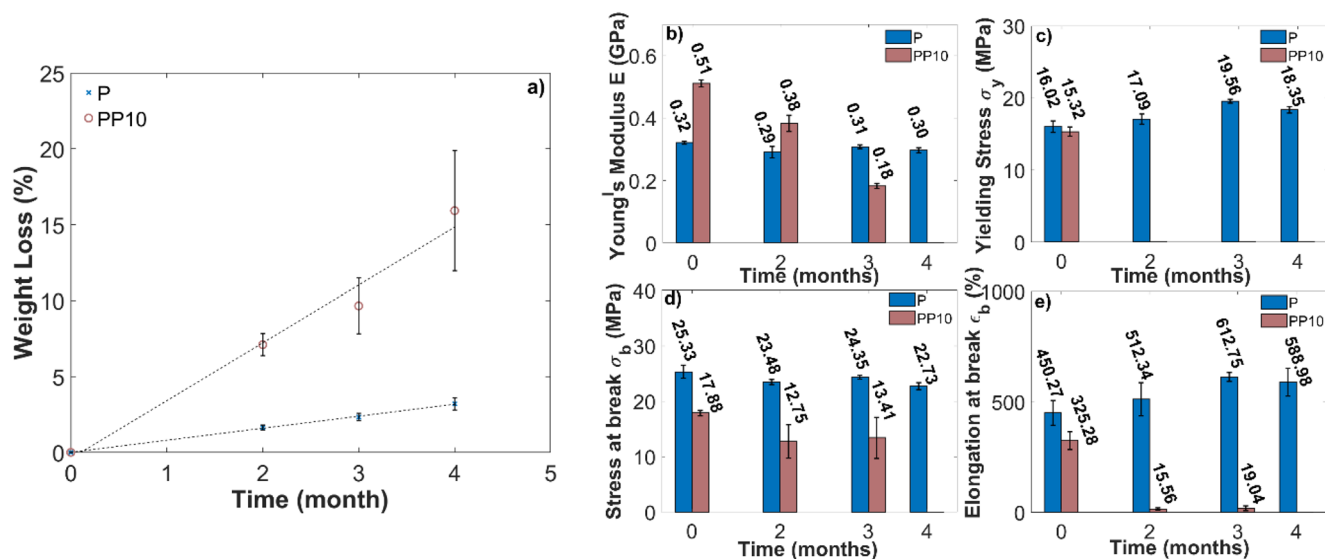


FIGURE 12 | (a) Percentage weight loss of P and PP10 after 4 months of deposition in seawater tank and (b) relative average value of E , δ_y , δ_b and ϵ_b compared during the test period.

in the absence of enzymatic activity; however, this effect is not relevant in the present study, as the temperature remained constant at approximately 20°C–21°C.

4 | Conclusion

This study demonstrates the feasibility of producing biobased P composites reinforced with PO powder obtained from coastal banquette residues, valorizing a natural biomass typically treated as waste within a circular material approach. The addition of small amounts of micro-talc improved extrusion processability without negatively affecting the overall material performance.

Thermal analyses show that the incorporation of talc and PO does not affect the intrinsic thermal stability of P, with degradation temperatures remaining around 403°C. However, DSC results reveal changes in crystallization behavior due to the nucleating effect of the two additives, leading to an increased crystallization temperature and the appearance of a secondary melting peak at about 79°C, while the overall crystallinity remains nearly unchanged.

FTIR analyses indicate the absence of significant chemical interactions between P and the additives. SEM observations reveal a generally homogeneous dispersion of both talc and PO particles within the matrix, along with good interfacial adhesion. However, PO particles exhibit a tendency to agglomerate due to moisture, as corroborated by the results of granulometric analysis.

As a result of this compatibility, the tensile and thermo-mechanical properties of the composites containing micro-talc remain comparable to those of neat P, although PO induces slight stiffening and mild embrittlement. DMA results demonstrate an increase in storage modulus across the investigated temperature range with increasing PO content, while the glass transition temperature remains essentially unchanged ($\approx -44.6^\circ\text{C}$) for all samples. Tensile tests confirm an increase in elastic modulus accompanied by reductions in tensile strength and elongation at

break for the PO/P composites, whereas the presence of micro-talc does not significantly affect the tensile response. Impact tests, however, reveal a marked reduction in toughness for filled systems, particularly at higher filler loadings.

Seawater exposure tests further show that PO significantly accelerates P biodegradation, promoting water uptake, fragmentation, and microbial colonization, which leads to early mass loss and a rapid deterioration of mechanical properties.

Overall, these results demonstrate that PO powder combined with micro-talc can be effectively used as a sustainable filler in biodegradable P composites, enabling controlled crystallization behaviour and accelerated environmental degradation while maintaining adequate processability and mechanical performance.

Author Contributions

Chiara Pedrotti: validation, formal analysis, investigation, writing – original draft, writing – review and editing, visualization. **Damiano Rossi:** methodology, formal analysis, investigation, writing – original draft, writing – review and editing, supervision, data curation. **Giovanni Raimondi:** investigation, resources. **Riccardo Gallorini:** investigation, resources. **Miriam Cappello:** investigation, resources. **Irene Anguillesi:** formal analysis, investigation. **Marco Sandroni:** formal analysis, investigation. **Patrizia Cinelli:** conceptualization, resources, writing – original draft, funding acquisition. **Maurizia Seggiani:** conceptualization, resources, writing – original draft, project administration, funding acquisition.

Acknowledgments

The authors would like to express their gratitude to Perla Galatolo and Marco Galatolo for their support in providing the *P. oceanica* leaves. Open access publishing facilitated by Università degli Studi di Pisa, as part of the Wiley – CRUI-CARE agreement.

Conflicts of Interest

The authors declare no conflicts of interest.

Data Availability Statement

The data that support the findings of this study are available from the corresponding author upon reasonable request.

References

1. I. Dentamare, L. Capasso, E. Chianese, et al., “Ecological Implications of *Posidonia oceanica* Banquette Removal: Potential Loss of Natural Capital and Ecosystem Services,” *Water (Basel)* 17, no. 9 (2025): 1362, <https://doi.org/10.3390/w17091362>.
2. S. Simeone, A. G. L. Palombo, F. Antognarelli, W. Brambilla, A. Conforti, and G. de Falco, “Sediment Budget Implications From *Posidonia oceanica* Banquette Removal in a Starved Beach System,” *Water (Basel)* 14, no. 15 (2022): 2411, <https://doi.org/10.3390/w14152411>.
3. N. Marbà, E. Díaz-Almela, and C. M. Duarte, “Mediterranean Seagrass (*Posidonia oceanica*) Loss Between 1842 and 2009,” *Biological Conservation* 176 (2014): 183–190, <https://doi.org/10.1016/j.biocon.2014.05.024>.
4. C. M. Duarte, “The Future of Seagrass Meadows,” *Environmental Conservation* 29, no. 2 (2002): 192–206, <https://doi.org/10.1017/S0376892902000127>.
5. C. Astudillo-Gutierrez, J. P. Sierra, C. Möso, and V. Gracia, “Impact of Banquette Structures on a Sandy Beach Profile: Laboratory Experiments,” *Coastal Engineering* 202 (2025): 104815, <https://doi.org/10.1016/j.coastaleng.2025.104815>.
6. L. Manfra, S. Chiesa, S. Simeone, et al., “Towards Sustainable Management of Beach-Cast Seagrass in Mediterranean Coastal Areas,” *Sustainability* 16, no. 2 (2024): 756, <https://doi.org/10.3390/su16020756>.
7. M. Á. Mateo, J. L. Sánchez-Lizaso, and J. Romero, “*Posidonia oceanica* ‘Banquettes’: A Preliminary Assessment of the Relevance for Meadow Carbon and Nutrients Budget,” *Estuarine, Coastal and Shelf Science* 56, no. 1 (2003): 85–90, [https://doi.org/10.1016/S0272-7714\(02\)00123-3](https://doi.org/10.1016/S0272-7714(02)00123-3).
8. G. Strangis, D. Rossi, P. Cinelli, and M. Seggiani, “Seawater Biodegradable Poly(Butylene Succinate-Co-Adipate)—Wheat Bran Biocomposites,” *Materials* 16, no. 7 (2023): 2593, <https://doi.org/10.3390/ma16072593>.
9. M. Chanda and J. R. Bathi, “Transport and Retention of Polyethylene Microplastics in Saturated Porous Media: Effect of Physicochemical Properties,” *Environmental Pollution and Management* 3 (2026): 52–58, <https://doi.org/10.1016/j.epm.2025.09.005>.
10. G. Fredi and A. Dorigato, “Recycling of Bioplastic Waste: A Review,” *Advanced Industrial and Engineering Polymer Research* 4, no. 3 (2021): 159–177, <https://doi.org/10.1016/j.aiepr.2021.06.006>.
11. A. Boldrini, N. Gaggelli, F. Falcai, et al., “Emerging Contaminants From Bioplastic Pollution in Marine Waters,” *Water (Basel)* 16, no. 24 (2024): 3676, <https://doi.org/10.3390/w16243676>.
12. D. D. Kapoor, P. Madaan, J. Kumar, S. K. Tiwari, K. K. Gupta, and R. K. Gupta, “Transition Towards Renewable and Biodegradable Polymers: A Comprehensive Review,” *Journal of Polymer Research* 32, no. 10 (2025): 345, <https://doi.org/10.1007/s10965-025-04524-8>.
13. F. Jahangiri, D. Nath, E. Pesaranhajiabbas, et al., “Study of Biodegradable Poly(Butylene Succinate-Co-Adipate) (PBSA) Maleation: Analysis of Grafting, Thermal, and Rheological Behaviour,” *Canadian Journal of Chemical Engineering* 5 (2025): 5357–5372, <https://doi.org/10.1002/cjce.25703>.
14. M. Salomez, M. George, P. Fabre, et al., “A Comparative Study of Degradation Mechanisms of PHBV and PBSA Under Laboratory-Scale Composting Conditions,” *Polymer Degradation and Stability* 167 (2019): 102–113, <https://doi.org/10.1016/j.polymdegradstab.2019.06.025>.
15. M. Puchalski, G. Szparaga, T. Biela, A. Gutowska, S. Sztajnowski, and I. Krucińska, “Molecular and Supramolecular Changes in Polybutylene Succinate (PBS) and Polybutylene Succinate Adipate (PBSA) Copolymer During Degradation in Various Environmental Conditions,” *Polymers (Basel)* 10, no. 3 (2018): 251, <https://doi.org/10.3390/polym10030251>.
16. M. Cappello, D. Rossi, S. Filippi, P. Cinelli, and M. Seggiani, “Wood Residue-Derived Biochar as a Low-Cost, Lubricating Filler in Poly(Butylene Succinate-Co-Adipate) Biocomposites,” *Materials* 16, no. 2 (2023): 570, <https://doi.org/10.3390/ma16020570>.
17. I. Delidovich, P. J. C. Hausoul, L. Deng, R. Pfützenreuter, M. Rose, and R. Palkovits, “Alternative Monomers Based on Lignocellulose and Their Use for Polymer Production,” *Chemical Reviews* 116, no. 3 (2016): 1540–1599, <https://doi.org/10.1021/acs.chemrev.5b00354>.
18. E. Pagliarini, C. Minichiello, L. Sisti, et al., “From Food Waste to Eco-Friendly Functionalized Polymer Composites: Investigation of Orange Peels as Active Filler,” *New Biotechnology* 80 (2024): 37–45, <https://doi.org/10.1016/j.nbt.2024.01.001>.
19. M. A. Masanabo, J. T. Keränen, S. S. Ray, and M. N. Emmambux, “Effect of Cowpea Lignocellulosic Fibers as a Low-Value Reinforcing Filler on the Properties of Poly(Butylene Succinate-Co-Adipate) Bio-Composite Foams,” *Macromolecular Materials and Engineering* 310, no. 7 (2025): 2400369, <https://doi.org/10.1002/mame.202400369>.
20. B. Ferrero, V. Fombuena, O. Fenollar, T. Boronat, and R. Balart, “Development of Natural Fiber-Reinforced Plastics (NFRP) Based on Biobased Polyethylene and Waste Fibers From *Posidonia oceanica* Seaweed,” *Polymer Composites* 36, no. 8 (2015): 1378–1385, <https://doi.org/10.1002/pc.23042>.
21. C. Fragassa, A. Pesic, S. Mattiello, A. Pavlovic, and C. Santulli, “Exploring the Potential of *Posidonia oceanica* Fibers in Eco-Friendly Composite Materials: A Review,” *Journal of Marine Science and Engineering* 13, no. 1 (2025): 177, <https://doi.org/10.3390/jmse13010177>.
22. R. Khiari, Z. Marrakchi, M. N. Belgacem, E. Mauret, and F. Mhenni, “New Lignocellulosic Fibres-Reinforced Composite Materials: A Step-forward in the Valorisation of the *Posidonia oceanica* Balls,” *Composites Science and Technology* 71, no. 16 (2011): 1867–1872, <https://doi.org/10.1016/j.compscitech.2011.08.022>.
23. R. Scaffaro, F. Lopresti, and L. Botta, “PLA Based Biocomposites Reinforced With *Posidonia oceanica* Leaves,” *Composites Part B: Engineering* 139 (2018): 1–11, <https://doi.org/10.1016/j.compositesb.2017.11.048>.
24. M. Seggiani, R. Altieri, P. Cinelli, A. Esposito, and A. Lazzeri, “Thermoplastic Blends Based on Poly(Butylene Succinate-Co-Adipate) and Different Collagen Hydrolysates From Tanning Industry: I—Processing and Thermo-Mechanical Properties,” *Journal of Polymers and the Environment* 29, no. 2 (2021): 392–403, <https://doi.org/10.1007/s10924-020-01880-y>.
25. A. Le Duigou, A. Bourmaud, P. Davies, and C. Baley, “Long Term Immersion in Natural Seawater of Flax/PLA Biocomposite,” *Ocean Engineering* 90 (2014): 140–148, <https://doi.org/10.1016/j.oceaneng.2014.07.021>.
26. M. Brebu, “Environmental Degradation of Plastic Composites With Natural Fillers—A Review,” *Polymers (Basel)* 12, no. 1 (2020): 166, <https://doi.org/10.3390/polym12010166>.
27. C. J. Tsenoglou, S. Pavlidou, and C. D. Papaspyrides, “Evaluation of Interfacial Relaxation due to Water Absorption in Fiber-Polymer Composites,” *Composites Science and Technology* 66, no. 15 (2006): 2855–2864, <https://doi.org/10.1016/j.compscitech.2006.02.022>.
28. T. Viel, I. Liotta, R. Avolio, et al., “The Fate of Biodegradable Polyesters in the Marine Environment,” *Polymer Degradation and Stability* 241 (2025): 111539, <https://doi.org/10.1016/j.polymdegradstab.2025.111539>.
29. Bausano, “Extrusion Instinct. Talc Powder” (2025), <https://www.bausano.com/en/extrusion-materials/additives-fillers/talc-powder-extrusion>.
30. Mitsubishi Chemical Europe GmbH, “BioPBS” (2025), <https://www.mcgp-global.com/en/mcgp-europe/products/brand/biopbsTM>.

31. VHM Company, "North America, Imerys Talc for Plastics," (2018), <https://www.vanhornmetz.com/function/talc-plastics/>.
32. Imerys, "Jetfine" (2025), <https://www.imerys.com/product-ranges/jetfine>.
33. D. Rossi, M. Cappello, S. Filippi, P. Cinelli, and M. Seggiani, "Potential of Lauryl Gallate as Stability and Recyclability Improver of Poly (Butylene Succinate-Co-Adipate)," *Journal of Polymers and the Environment* 32, no. 3 (2024): 1202–1216, <https://doi.org/10.1007/s10924-023-03017-3>.
34. UNI EN ISO, "ISO 527-2: Plastics—Determination of Tensile" (2012), <https://cdn.standards.iteh.ai/samples/56046/c5c3ec20cc0741289bf6def861e0a40b/ISO-527-2-2012.pdf>.
35. UNI EN ISO, "ISO 179-1: Plastics—Determination of Charpy Impact Properties" (2023), <https://cdn.standards.iteh.ai/samples/84393/c5581746f5fa4df4830d0b7257272d21/ISO-179-1-2023.pdf>.
36. M. R. Ricciardi and V. Antonucci, "Thermal Degradation and Fire Behavior of *Posidonia oceanica* Epoxy Composites," *Journal of Composites Science* 9, no. 7 (2025): 349, <https://doi.org/10.3390/jcs9070349>.
37. A. Plis, J. A. Lasek, J. Zuwała, C. C. Yu, and A. Iluk, "Combustion Performance Evaluation of *Posidonia oceanica* Using TGA and Bubbling Fluidized-Bed Combustor (Batch Reactor)," *Journal of Sustainable Mining* 15, no. 4 (2016): 181–190, <https://doi.org/10.1016/j.jsm.2017.03.006>.
38. A. Plis, J. Lasek, A. Skawińska, and M. Kopczyński, "Thermo-Chemical Properties of Biomass From *Posidonia oceanica*," *Chemical Papers* 68, no. 7 (2014): 879–889, <https://doi.org/10.2478/s11696-013-0532-4>.
39. V. Chiodo, G. Zafarana, S. Maisano, S. Freni, and F. Urbani, "Pyrolysis of Different Biomass: Direct Comparison Among *Posidonia oceanica*, Lacustrine Alga and White-Pine," *Fuel* 164 (2016): 220–227, <https://doi.org/10.1016/j.fuel.2015.09.093>.
40. L. Bertini, F. Focaracci, S. Proietti, P. Papetti, and C. Caruso, "Physiological Response of *Posidonia oceanica* to Heavy Metal Pollution Along the Tyrrhenian Coast," *Functional Plant Biology* 46, no. 10 (2019): 933–941, <https://doi.org/10.1071/FP18303>.
41. M. Seggiani, V. Gigante, P. Cinelli, et al., "Processing and Mechanical Performances of Poly(Butylene Succinate–Co–Adipate) (PBSA) and Raw Hydrolyzed Collagen (HC) Thermoplastic Blends," *Polymer Testing* 77 (2019): 105900, <https://doi.org/10.1016/j.polymertesting.2019.105900>.
42. G. Jiang, H. Wang, L. Yu, and H. Li, "Improving Crystallization Properties of PBSA by Blending PBS as a Polymeric Nucleating Agent to Prepare High-Performance PPC/PBSA/AX8900 Blown Films," *Polymer Engineering and Science* 62, no. 4 (2022): 1166–1177, <https://doi.org/10.1002/pen.25915>.
43. R. Várdai, Á. Schäffer, M. Ferdinánd, et al., "Crystalline Structure and Reinforcement in Hybrid PP Composites," *Journal of Thermal Analysis and Calorimetry* 147, no. 1 (2022): 145–154, <https://doi.org/10.1007/s10973-021-11053-1>.
44. L. A. Castillo and S. E. Barbosa, "Comparative Analysis of Crystallization Behavior Induced by Different Mineral Fillers in Polypropylene Nanocomposites," *Nanomaterials and Nanotechnology* 10 (2020): 184798042092275, <https://doi.org/10.1177/1847980420922752>.
45. S. Boubakri, M. A. Djebbi, Z. Bouaziz, et al., "Nanoscale Zero-Valent Iron Functionalized *Posidonia oceanica* Marine Biomass for Heavy Metal Removal From Water," *Environmental Science and Pollution Research* 24, no. 36 (2017): 27879–27896, <https://doi.org/10.1007/s11356-017-0247-0>.
46. S. Chen, T. Shao, J. Ouyang, et al., "Optimizing Fermentation Quality via Lignocellulose Degradation: Synergistic Effects of Fibrolytic Additives and Lactiplantibacillus Plantarum on *Elymus dahuricus* Silage," *Chemical and Biological Technologies in Agriculture* 12, no. 1 (2025): 70, <https://doi.org/10.1186/s40538-025-00795-x>.
47. L. Yang and J. P. Chen, "Biosorption of Hexavalent Chromium Onto Raw and Chemically Modified *Sargassum* sp.," *Bioresource Technology* 99, no. 2 (2008): 297–307, <https://doi.org/10.1016/j.biortech.2006.12.021>.
48. V. F. Meseguer, J. F. Ortuño, M. I. Aguilar, et al., "Biosorption of Cadmium (II) From Aqueous Solutions by Natural and Modified Non-Living Leaves of *Posidonia oceanica*," *Environmental Science and Pollution Research* 23, no. 23 (2016): 24032–24046, <https://doi.org/10.1007/s11356-016-7625-x>.
49. P. Schroeder, "Infrared Spectroscopy in Clay Science," *Teaching Clay Science* 11 (2002): 181–206.
50. V. Čavajda, P. Uhlík, A. Derkowski, et al., "Influence of Grinding and Sonication on the Crystal Structure of Talc," *Clays and Clay Minerals* 63, no. 4 (2015): 311–327, <https://doi.org/10.1346/CCMN.2015.0630405>.
51. Y. Wang, Y. Zhong, Q. Shi, and S. Guo, "Study of the Preparation and Properties of TPS/PBSA/PLA Biodegradable Composites," *Journal of Composites Science* 5, no. 2 (2021): 48, <https://doi.org/10.3390/jcs5020048>.
52. Z. Qi, H. Ye, J. Xu, J. Chen, and B. Guo, "Improved the Thermal and Mechanical Properties of Poly(Butylene Succinate-Co-Butylene Adipate) by Forming Nanocomposites With Attapulgite," *Colloids and Surfaces A: Physicochemical and Engineering Aspects* 421 (2013): 109–117, <https://doi.org/10.1016/j.colsurfa.2012.12.051>.
53. R. Wang, S. Wang, and Y. Zhang, "Morphology, Mechanical Properties, and Thermal Stability of Poly(L-Lactic Acid)/poly(Butylene Succinate-Co-Adipate)/silicon Dioxide Composites," *Journal of Applied Polymer Science* 113, no. 6 (2009): 3630–3637, <https://doi.org/10.1002/app.30353>.
54. D. Lascano, L. Quiles-Carrillo, R. Balart, T. Boronat, and N. Montanes, "Toughened Poly (Lactic Acid)—PLA Formulations by Binary Blends With Poly(Butylene Succinate-Co-Adipate)—PBSA and Their Shape Memory Behaviour," *Materials* 12, no. 4 (2019): 622, <https://doi.org/10.3390/ma12040622>.
55. W. Post, L. J. Kuijpers, M. Zijlstra, M. van der Zee, and K. Molenveld, "Effect of Mineral Fillers on the Mechanical Properties of Commercially Available Biodegradable Polymers," *Polymers (Basel)* 13, no. 3 (2021): 394, <https://doi.org/10.3390/polym13030394>.
56. G. S. Ananthapadmanabha and V. Deshpande, "Influence of Aspect Ratio of Fillers on the Properties of Acrylonitrile Butadiene Styrene Composites," *Journal of Applied Polymer Science* 135, no. 11 (2018): 46023, <https://doi.org/10.1002/app.46023>.
57. V. Atluri, J. Jin, K. Shrimali, L. Dang, X. Wang, and J. D. Miller, "The Hydrophobic Surface State of Talc as Influenced by Aluminum Substitution in the Tetrahedral Layer," *Journal of Colloid and Interface Science* 536 (2019): 737–748, <https://doi.org/10.1016/j.jcis.2018.10.085>.
58. E. Sasimowski, Ł. Majewski, and M. Grochowicz, "Influence of the Conditions of Corotating Twin-Screw Extrusion for Talc-Filled Polypropylene on Selected Properties of the Extrudate," *Polymers (Basel)* 11, no. 9 (2019): 1460, <https://doi.org/10.3390/polym11091460>.
59. D. C. McNeill, A. K. Pal, A. K. Mohanty, and M. Misra, "Injection Molding of Biodegradable Polyester Blends Filled With Mineral and Sustainable Fillers: Performance Evaluation," *Journal of Applied Polymer Science* 141, no. 13 (2024): e55166, <https://doi.org/10.1002/app.55166>.
60. M. Seggiani, P. Cinelli, N. Mallegni, et al., "New Bio-Composites Based on Polyhydroxyalkanoates and *Posidonia oceanica* Fibres for Applications in a Marine Environment," *Materials* 10, no. 4 (2017): 326, <https://doi.org/10.3390/ma10040326>.
61. M. Zumstein, G. Battagliarin, A. Kuenkel, and M. Sander, "Environmental Biodegradation of Water-Soluble Polymers: Key Considerations and Ways Forward," *Accounts of Chemical Research* 55, no. 16 (2022): 2163–2167, <https://doi.org/10.1021/acs.accounts.2c00232>.

62. R. A. Pérez-Camargo, G. Liu, D. Cavallo, D. Wang, and A. J. Müller, "Effect of the Crystallization Conditions on the Exclusion/Inclusion Balance in Biodegradable Poly(Butylene Succinate- *Ran* -Butylene Adipate) Copolymers," *Biomacromolecules* 21, no. 8 (2020): 3420–3435, <https://doi.org/10.1021/acs.biomac.0c00847>.
63. A. Svarcova, M. Dvorackova, and P. Svoboda, "Influence of Hydrolysis on Non-Isothermal Crystallization of Poly(Butylene Succinate-Co-Adipate) (PBSA)," *Molecules* 30, no. 11 (2025): 2252, <https://doi.org/10.3390/molecules30112252>.