A combined rheological and thermomechanical analysis approach for the assessment of pharmaceutical polymer blends

Mohammad Isreb[1], Marianiki Chalkia[1,2], Timothy Gough[3], Robert Thomas Forbes[1,4], Peter Timmins[5,6].

[1]School of Pharmacy and Medical Sciences, University of Bradford, Bradford, BD7 1DP, UK.

[2]Present address: Covance Clinical and Periapproval Services Ltd, Harrogate HG3 1PY, UK.

[3]School of Engineering and Informatics, University of Bradford, Bradford, BD7 1DP, UK.

[4]Present address: School of Pharmacy and Biomedical Sciences, University of Central Lancashire, Preston, PR1 2HE, UK.

[5]Drug Product Science and Technology, Bristol-Myers Squibb, Reeds Lane, Moreton, Merseyside, CH46 1QW, UK.

[6]Present address: Department of Pharmacy, University of Huddersfield, Huddersfield HD1 3DH, UK.

Abstract

The viscoelastic nature of polymeric formulations utilised in drug products imparts unique thermomechanical attributes during manufacturing and over the shelf life of the product. Nevertheless, it adds to the challenge of understanding the precise mechanistic behaviour of the product at the microscopic and macroscopic level during each step of the process. Current thermomechanical and rheological characterisation techniques are limited to assessing polymer performance to a single phase and are especially hindered when the polymers are undergoing thermomechanical transitions. Since pharmaceutical processing can occur at these transition conditions, this study successfully proposes a thermomechanical characterisation approach combining both mechanical and rheological data to construct a comprehensive profiling of polymeric materials spanning over both glassy and rubbery phases. This approach has been used in this study to assess the mechanical and rheological behaviour of a heterogenous polymer blends of hydroxypropyl cellulose (HPC) and hydroxypropyl methylcellulose (HPMC) over a shearing rate range of 0.1-100 s^{-1} and a temperature range of 30-150°C. This profiling provides a unique insight towards a quality-by-design approach to polymer-based solid dosage form design. The results indicates that HPC and HPMC do not appear to interact when mixing and that their mixture exhibits the mechanistic properties of the two individual polymers in accordance to their ratio in the mixture.

Introduction

Polymers find many uses in pharmaceutical products, including acting as bulking agents and binders to enable the granulation of powders, as disintegrants, aesthetic (non-functional) or

functional coatings for tablets and modifiers of drug release from dosage forms. They can also be employed as film forming agents for mouth dissolving medicated strips and as stabilisers of amorphous forms of normally crystalline drugs, to improve solubility and subsequently bioavailability (Liechty, Kryscio et al. 2010, Liu, Taylor et al. 2015, Debotton and Dahan 2017) . The reason for the popularity of polymers is that these unique materials offer a variety of favourable properties such as enhancing stability, providing protection from moisture and light, enabling processability and also being employed to modify pharmacokinetics and biological activity (Jones 2004). However, the recent growing demand to achieve quality by design means that the properties and performance of excipients must be adequately understood to help an informed approach to process development and to predict the short and long-term performance and stability of dosage forms.

Contrary to simple organic small molecule compounds (such as pharmaceutical drug substances), polymers are neither ideal elastic solids nor viscous liquids but are viscoelastic materials. This means that their behaviour is time, temperature and shear rate dependant. In other words, at any temperature, polymers may behave as solids or liquids depending on the time scale and /or the rate of chains deformation (strain rate) (Mezger 2006). A number of pharmaceutical processes that involve polymers, such as film coating, spray drying or hot melt extrusion, involve processing at elevated temperatures, therefore, it is essential to understand the behaviour of polymers at the temperature and shear rate used in such pharmaceutical processes. Moreover, it is vital to understand the chain relaxation dynamics of the polymer over the shelf life of the dosage form as these could potentially change the physical dimension of the product (e.g. die swelling, creep) (Larson 1999) and film defects such as bridging, its mechanical properties (e.g. tensile strength) as well as its stability and/or the release profile (e.g. film cracking) (Michael E. Aulton 1995)

Until recently, pharmaceutical processing parameters were usually selected using a trial and error approach without necessarily employing a comprehensive understanding of the viscoelastic behaviour of the polymer over the precise conditions of the various stages of the process. For a typical pharmaceutical process, polymers are required to exhibit various distinctive viscoelastic behaviour at each step. For example, in hot melt extrusion polymers are required to soften, coalesce and flow within the time, temperature and shear force exerted in the extruder. However, once outside the extruder the extrudates are expected to be tough and strong (i.e. predominantly elastic) to withstand further processing and handling stresses. Nevertheless, unlike other industries, most typical pharmaceutical processes are carried out at relatively low temperatures (below $200^{\circ}C$) to avoid the degradation of the active pharmaceutical ingredients (APIs). Meanwhile, most practical pharmaceutical polymers are amorphous thermoplastic materials. Therefore, they exhibit no sharp thermal transition from solid to liquid phases. Instead, the polymers transform gradually from predominantly elastic materials to materials dominated by their viscous moduli over a range of temperatures (the midpoint of which is referred to as the glass transition temperature, T_g) which vary significantly with shear rate and humidity(strobl 1997).

To impart desired handling properties that help process pharmaceutical polymer formulations in a relatively narrow temperature range, polymer blends and additives are frequently used in the pharmaceutical industry. The impact of these additives is usually characterised using a variety of thermal and mechanical techniques aimed at understanding polymer behaviour at very specific distinctive conditions. These techniques include intrinsic viscosity measurements (Entwistle 1979), mechanical strength testing (Aulton 1981, Gibson 1988, Bodmeier 1990), incidence of stress-induced defects (Rowe 1981), glass transition measurements (Entwistle 1979, Porter 1983), softening temperature (Masilungan 1984) and rheological properties (Paradkar, Kelly et al. 2009).

The aforementioned techniques are useful for the processes that take place in one phase (i.e. either above or below the T_g of the formulation) or at one constant shear regime. For example, the mechanical properties of polymers in their glassy and glass/rubber transition states are usually measured using solid mechanics instruments, such as uniaxial tensile machines or dynamic mechanical analysers (DMA) (Bashaiwoldu, Podczeck et al. 2004). Parameters such as tensile strength, elastic modulus and tan delta (defined as the loss modulus/storage modulus of the sample) are the main parameters usually measured in that case. On the other hand, shear and extensional rheometers are usually used to characterise polymeric formulations in their rubber/liquid and liquid phase (Paradkar, et al., 2009). Storage (elastic), loss moduli, tan delta and complex viscosity are the main measurements to be obtained and analysed. However, the transitional phases between rubber to rubber/liquid are on the borderline of the accuracy limits of the two techniques and are harder to measure accurately and thus be modelled.

Many pharmaceutical processes occur at conditions at or near the transitional phases of the polymers whilst operating at temperature that avoid thermal degradation of drugs. Therefore, it is hard to obtain a comprehensive picture of the polymer performance using a single technique from those mentioned above.

To our best knowledge, there has been no previous attempt to bridge the results from both solid mechanics and rheology into one continuous model. In fact, insufficient work has been carried out to model and understand the behaviour of pharmaceutical polymers and their blends in either the glass or the rubber phases. It is ,therefore, vital to build up such knowledge if quality-by-design formulations and processes are to be adopted for optimising pharmaceutical processes.

This study is aimed at investigating the feasibility of harmonising the mechanical and rheological data, obtained from DMA and shear rheometer over a wide range of temperature and shear forces typically encountered during the processing and shelf life of pharmaceutical products. This harmonisation will help produce a continuous and comprehensive master profile to predict the behaviour of two cellulose-based polymers commonly used in the pharmaceutical industry, namely hydroxypropyl cellulose (HPC) and hydroxypropyl methylcellulose (HPMC). Moreover, the effect of polymer blending on the chain dynamics of these polymers will be investigated over the whole range of temperature and shear forces. This will help in the understanding of the extent of interaction between the two polymers and the overall effect of their blend on future solid dosage form design. HPC/ HPMC mixtures have been used in hot melt extrusion ((Repka, Gutta et al. 2005, Wilson, Jones et al. 2017) and matrix tablets (Vueba, Batista de Carvalho et al. 2013) to exploit the thermoplasticity of HPC and the added functionality of HPMC (modify drug release, increase tensile strength etc). However, a proper understanding of the effect of blending these polymers has rarely been investigated beyond the T_g value. Indeed, changes in Tg values and comparing the calculated solubility parameters could be used to determine the miscibility of two polymers. Mixing two polymers could lead to the formation of a product with multi heterogenous domains, rich in one of the two polymers, and subsequently impacting the manufacturing process (e.g. heterogenous flow, heat transfer, drug solubility) and final product attributes (e.g. mechanical properties and drug loading, distribution and release profile)

Materials and Methods

Materials

Hydroxypropyl cellulose (HPC), 850,000 Daltons molecular weight, viscosity of 2% aqueous solution 4000-6500 mPa (Klucel™ MF, Ashland, Inc., Covington, KY, USA), and hypromellose

2208 (hydroxypropyl methylcellulose, HPMC), (Methocel™ K100M, Colorcon, Orpington, UK), were acquired from Dow, UK. Ethanol (reagent grade) was purchased from Fisher Chemical, UK and ultra-purified water (18.2MΩ·cm) was produced using a Triplered Laboratory Technology filtering system (Avidity Science, UK).

Preparation of free films

Free films of the polymers and their blends were prepared to enable characterisation. Two methods were used in the preparation of the polymers for characterisation.

Compressed discs preparation

Compressed discs of HPC and HPMC and their blends (table 1) were prepared using a thermal hydraulic press preheated to $210^{o}C$. A sample of 25g of the polymer powder (or a pre-blended mixture of the two) were oven dried at $60^{o}C$ for 2 hours to help reduce risk of film cracking due to moisture evaporation during compression. The dried powder was placed on the surface of the plates in a thickness of approximately 2 mm and then compressed by applying a 20,000 kg force for 10 min using a 0.5mm spacer. The hydraulic pressure was then released gradually over one minute to avoid cracking due to a sudden elastic die swelling. The final thickness was 0.5 ± 0.05mm

Casting polymer from solution

In order to assess the effect of film compactibility on the rheological measurements value, cast films of both pure HPC and HPMC films were prepared. Blends of polymers were difficult to prepare as cast films as it was challenging to dissolve both HPC and HPMC in a common solvent with a reasonable concentration of both polymers for casting. Hence for cast films, only pure single polymers were evaluated. HPC solution was prepared using ethanol: water mixture (85:15 v/v) at a solid ratio of 2% w/v. Polymer powder was added gradually to the solvent mixture under constant

stirring to avoid agglomeration. Films were then cast onto an acetate sheet placed over levelled glass Petri-dishes and left to dry in a fume cupboard. Successive additions of polymer solutions were needed to build up the thickness of the dry film to 0.5mm (±0.5mm). The drying of the film was controlled by covering the petri dish with aluminium foil with pin-holed pores to reduce the evaporation rate over the first 24h to avoid the wrinkling of the film surface and the build-up of internal stress.

HPMC solution was prepared in water at a solid ratio of 1.5% w/v, using water at 80°C when introducing the polymer. HPMC is practically insoluble in hot water therefore it is less likely to agglomerate. Once the polymer was dispersed in the hot water, the temperature is reduced gradually under constant stirring until all the polymer was dissolved. The petri dish was stored in vacuum oven at 30°C and at a partial vacuum pressure of 0.3 bar overnight. The films were then left to equilibrate at room conditions (20°C and 65% RH) along the rest of the films. Film thickness was 0.5±0.05mm.

Thermal gravimetric analysis

Thermal gravimetric analysis (TGA) test was carried out using a Q500, Discovery series instrument (TA Instruments, New Castle, De, USA*)* to measure the moisture content and the decomposition temperature of the starting materials. Compressed polymer samples (weighing 15±1 mg) were cut for analysis using a cork borer size 2 (5.8 mm in diameter) and evaluated at a heating rate of 10 °C/min. using nitrogen gas atmosphere. The temperature range used was 30- 400 °C and the gas flow was set to 25 ml/min. All measurements were carried out in duplicate and analysed using TRIOS software (TA Instruments, New Castle, De, USA). The determination of the decomposition onset value is essential to identify temperature range to be used in differential

scanning calorimetry (DSC) test. This helps in avoiding sample decomposition during DSC tests. The degradation temperature was measured just before the onset of the transition peak.

Differential scanning calorimetry

DSC was performed using a Q2000, Discovery series calorimeter (TA Instruments, New Castle, De, USA) with a refrigerated cooling system. Compressed polymer samples (weighting 15±1 mg) were cut for analysis using a cork borer size 2 (5.8 mm diameter) and enclosed in an aluminium pan with a pin-holed lid to allow moisture evaporation. A heat/cool/heat cycle was carried out for all the samples at a ramping rate of 10 ºC/min heating and 5 ºC/min for the cooling cycle.

Dynamic mechanical analysis

DMA (Q 800, TA Instruments, Newcastle, DE, USA) was used in this study to perform two separate sets of experiments. Firstly, DMA was used to determine the T_g of the compressed films. Secondly, the mechanical properties and the dynamic moduli of the polymer or polymer blend preparations were measured. In both tests film strips cut from both compressed and cast films (12mm x 3mm x 0.5mm) were evaluated using an 8mm long dual cantilever clamp.

The details of the two methodologies used to measure the samples are presented below

A. DMA methodology for T_g measurements of the films:

Thermal transitions were measured using a temperature sweep test at a fixed frequency of 1 Hz, 0.1% strain and a temperature ramp rate of 3°C/min. Loss and storage moduli were measured. The results were analysed using "Universal Analysis 2000 version 4.3A"

software from TA instruments. Moreover, the ratio of viscous modulus to elastic modulus at a given frequency (known as tan delta) was calculated using the following equation:

$\tan(\delta) = G''/G'$ Equation 1

Tan delta is widely used to quantify the balance between energy loss and storage in the material. When $\tan(\delta)$ is greater than unity, the material has more viscous properties or plasticity, whereas $\tan(\delta)$ value lower than unity means more elastic properties, regardless of the viscosity (Mezger 2006).

B. DMA methodology for dynamic moduli measurements:

In order to mimic the measurements of complex viscosity in shear rheometer, DMA was used in a frequency sweep mode to measure the storage and loss moduli using dual cantilever clamp. Frequency sweep tests were carried out for both compressed and cast films. Initially, a strain sweep test was carried out to determine the linear viscoelastic region of the strain range. Afterwards, frequency sweep tests were carried out using a constant strain value of 0.1%. The temperature was adjusted to 30°C, held isothermally for 3 min, and a frequency sweep was carried out over a frequency range of 0.1-100 Hz. The temperature was then raised by 10°C and the cycle was repeated until the sample deformed and the test was disrupted. The loss modulus, storage modulus, and damping factor were recorded and complex viscosity was calculated according to equation 2 (Larson 1999):

$$\eta^* = \frac{G''}{\omega} + i\frac{G'}{\omega}$$ Equation 2

Where:

$\eta^* =$ complex viscosity

$G' =$ Storage or elastic modulus

$G'' =$ Loss modulus

$\omega =$ Angular frequency

$i =$ square root of -1

Shear rheometer procedure

Oscillatory strain amplitude test was carried out using a Physica MCR 501" rheometer (Anton Paar, Graz, Austria). Parallel plate (diameter = 25mm) configuration was used to perform all the tests. The gap between the two plates was adjusted to 0.55±0.5mm (depending on the film thickness). Strain amplitude tests (not shown here) were carried out to determine the linear viscoelastic region (LVR) in order to select the strain value to be used subsequently. Frequency sweep tests were performed using 3% strain and a frequency range of 0.01-100 rad/s. compressed and cast films were measured at a temperature range of 140-230°C. Storage, loss and complex viscosities were recorded.

Combining DMA and PPSR data

Viscosity, referring to the resistance of the material to flow, is usually measured by rheometers and is used to describe the viscoelasticity of polymers in their melt state (above their T_g). On the other hand, DMA is usually used to evaluate the thermomechanical properties of the material at a given temperature below and around its T_g. Compared to other thermal analysis methods, such as DSC, DMA provides information about major transitions as well as the more subtle events caused by a side group's rotation and/or stretching (Cassel 1991). However, since pharmaceutical

processes occur mostly at the transitional boundaries between the two phases in which the polymer transforms between its glassy and rubbery states, this study is investigating the possibility of using complex viscosity to express the viscoelasticity of the polymer over both the elastic and the rubber phases. To do this, some assumptions have to be made to simplify the process and allow for the data harmonisation. Firstly, this study assumed that thermoplastic polymers in the glassy state are very viscous liquids that will flow over an infinite time scale. Moreover, although the bending movement of the DMA is not a pure shear process, it was assumed to be directly proportional to the shearing movement in the PPSR. These assumptions allow for a direct comparison between complex viscosity values calculated from the dynamic moduli measured in both the PPSR and DMA techniques and therefore a plot of continuous viscosity versus temperature can be constructed across a temperature range spanning from room temperature to well above the T_g of both polymers. The frequency of oscillation in both tests was expressed in terms of linear frequency according to equation 3:

$$\boldsymbol{\omega = 2\pi f}\qquad\qquad \text{Equation 3}$$

where $\boldsymbol{\omega}$ is the angular frequency in rad/s and $\boldsymbol{f}$ is the regular frequency and is given by s^{-1}

In order to reduce the data in two simplified 2D graphs, Power law model parameters were used to compare the complex viscosity data of polymers (Paradkar, Kelly et al. 2009). The advantage of using Power law is that it reduces the information in the complex viscosity graph over a range of frequency to two numbers: K and n. Therefore, it was decided to adopt these parameters to allow for the comparison between the data measured by the DMA and PPSR techniques over the whole temperature range. The complex viscosity data from both the DMA and PPSR therefore

were fitted to a power law trend line. The power law formulation can be represented by the equation 4:

$$\eta = K\dot{\gamma}^{n-1} \qquad \text{Equation 4}$$

where η is the complex viscosity of the sample, $\dot{\gamma}$ is the shear rate, K is the consistency coefficient which describes complex viscosity value (calculated from the fitted line) at a frequency value of 1 s^{-1}, n is the Power Law Index which decreases as the shear sensitivity of the polymer increases. Therefore, K and n could describe the complex viscosity of the material over the measured frequency range at specific temperature.

Results and discussion

Thermal and rheological measurements were used to analyse the effect of blending HPC and HPMC as well as to establish the feasibility of reconciling DMA and PPSR data

Thermal Gravimetric analysis (TGA) of the polymers

The moisture content and the thermal degradation of the compressed films were studied using thermal gravemitric analysis (TGA) under nitrogen atmosphere. The results are shown in figure 1. The thermograms of both polymers exhibit an initial drop in the weight due to the evaporation of the residual moisture. Interestingly the moisture evaporation seems to extend to around 140°C for HPC films and to around 200°C for HPMC. The moisture content seems to be around 3% for HPC films and around 4% for HPMC films. It is worth noting that this is the equilibruim moisture content of the films rather than the powder as the powder was dried in oven at 80°C for 2 hours before presseing the films (Repka, Gutta et al. 2005). Compressing the powder without the drying

step caused cracking in the films. The thermal degradation seems to start around 300°C under nitrogen atmosphere.

Thermal analysis of the films using DSC and DMA

Thermal analysis of HPMC and HPC films and their blends has been conducted using DSC and DMA and the results are presented in figure 2 and 3.

The first heating cycle of both polymers reveals a wide thermal transition between 50-150°C which is consistent with the moisture loss in TGA thermogram. The second run reveals a T_g value of around 187°C for HPMC and 193°C for HPC films.

DMA temperature sweep tan delta graphs of the films are presented in figure 3 and table 2. The figure reveals that for HPC the ratio of the viscous to elastic moduli (tan delta) peaks at around 140°C just before the loss of moisture. The ratio drops afterwards (indicating an increase in overall elastic behaviour of the sample) before peaking again at 215°C. On the other hand, it seems that the moisture content of the HPMC film is not playing a noticeable plasticising effect. The tan delta signal for HPMC only peaks at around 216 °C. The mixtures are behaving according to the ratio of HPC to HPMC. The mixture with 80% HPC shows the same peak around 140°C as pure HPC films. The value of this peak however decreases in the other mixtures as the ratio of HPC decreases. It seems that the mixtures are retaining the characteristic peaks of both polymers and the overall behaviour depends on the ratio of both polymers.

In summary the T_g values measured using the DSC are different to the transitions registered by the DMA. This can be explained by the fact that the T_g was measured in the second cycle of heating in the DSC (i.e. after the evaporation of the moisture and with loss of its plasticising effect during

the first cycle) whereas the DMA measured the T_g of the sample while heating it for the first time (not enough time for the moisture to evaporate fully). Moreover, the samples are larger in the DMA (has smaller surface area and therefore water evaporation is slower). It is simply not feasible to measure the Tg of the polymers in the DSC using the first heating cycle as the moisture evaporation endothermic peak is significantly more prominent and will obscure the subtle Tg step.

It is worth noting that, while the DMA data shows mixing between the two polymers in the blends was such that domains of each individual polymer existed, it reveals little information about the overall behaviour of the film below or above the single or double transitions noticed in the films. The powder of the two polymers were mixed prior to compressing the films. However, an ideal mixture was unlikely to have been achieved. Therefore, the presence of individual polymer phases could be also partially explained by lack of ideal mixing. The samples, however, are more realistic representation of the mixing that can be expected in pharmaceutical processing such as granulation or extrusion and the results therefore are invariably representative of the interaction that would take place during such processes

Rheological results

Amplitude sweep test was carried out to determine the strain range through which a linear response is maintained (known as LVE range). In this region the sample deformation is reversible and the sample behave in repeatable manner. From the results, it was decided to use a strain value of 3% for all oscillatory tests in the rheometer and 0.1% strain was used in the DMA experiments.

Complex viscosity data from both the DMA and PPSR were measured and Power law model was used to model the complex viscosity graphs. Power law parameters: the consistency coeffecient K

and the shear sensitivity index (n) were calculated from the equation and plotted against the temperature for each film in figures 4 and 5.

The K value (figure 4) represents the value of the average complex viscosity at 1 htz. As expected, the figure reveals that, as the HPC ratio in the film drops, the complex viscosity increases. This is expected as HPC has more pronounced thermoplastic behaviour than HPMC (Repka, 2005). However, it is interesting to note that the viscosity of HPC compressed films starts to drop significantly faster than all other films above 80°C. The rate of reduction in the viscosity value slows down closer to 140°C which is the temperature at which the first transition was noticed in DMA temperature sweep. Importantly, each film exhibits a unique behaviour that cannot be inferred from the Tg value presented in the DSC or the DMA. This illustrates the importance of measuring the mechanical and rheological behaviour of the film at the required shear range at each temperature. Since no single instrument can perform this measurement over the whole temperature range, it is essential to combine the data from more than one technique as it was attempted in this study.

The shear sensitivity (figure 5) of all the films decreases sharply (higher n value) with increasing temperature up until 140°C. The exception is the film with $\leq 20\%$ HPC content. This is in line with the Tg data from the DMA. Moreover, all the films reveal another reduction in their shear sensitivity around 180°C as HPMC component soften as temperature approaches its Tg value (figure 5). This again support the suggestion that the two polymers are not completely mixed with each other as the characteristic of each polymer is visible in the mixture.

Most importantly the complex viscosity data from the DMA and PPSR seems to be following the same trend with a correlation factor of 12 for K values of HPC rich films and 1.5 for n values of all films (Table 3).

The difference could be due to the fact that both instruments are at the limit of their ability to reliably make measurements for the samples. In DMA samples start to soften and sag between the suspension points around the Tg value which means the measurements are no longer within the LVE region. Likewise, the PPSR starts to struggle in controlling the amplitude of the rotational shearing as the sample harden. Moreover, the difference might also be due to the fact that the polymer particles have not fully coalesced when they were compressed to make the film. If this assumption is correct then the DMA measurements would not affect the degree of coalescence because it operates below the Tg value of the polymers. Meanwhile the PPSR is expected to promote particles coalescence and chain entanglement through the shearing force and the fact that the film is in the melt state.

To test this hypothesis cast films of HPC and HPMC were prepared and tested using the same methodology. The comparison between cast and compressed film results were demonstrated in figures 6 and 7. It is clear that PPSR data has not been affected by the preparation methodology while as predicted DMA data has changed. As predicted the viscosity values measured using the compressed films were higher than those measured from the cast film in the case of HPC. This might be due to the friction forces between the incompletely coalesced particles of the powder. The DMA dual cantilever exerts a complex stress profile combining compression, tensile and shearing elements that is different to the shearing forces in the PPSR.

In the case of HPMC, the difference between the DMA and PPSR measurements for the cast film was reduced to 3 (compared to a difference ratio of 9 in the case of compressed films). The remaining discrepancy might be due to the highly elastic nature of HPMC compared to HPC and the difficulty that this elasticity imposes on the PPSR technique. In fact, the PPSR file shows that the error in the repeated measurements was "higher than normal". The film sample of HPC

softened when measured with PPSR allowing the smooth surface of the film to adhere to the plats and thus eliminating gaps between the film and the plates and providing appropriate grip for the shearing force to take place. The HPMC film however was still elastic and therefore the cast film with the smooth surface was even more slippery that the rougher surface compressed powder sample and the measured valued were lower than that of the cast film.

Conclusion

This study demonstrates for the first time the feasibility of reconciling the complex viscosity data measured using DMA dual cantilever and PPSR. Up to our best knowledge this is the first study to hypothesise the possibility of using complex viscosity equation below the T_g of amorphous polymers to gain an extra insight into their performance. In this study, it was successfully used to evaluate the miscibility and overall effect of blending HPMC and HPC. However, it can also be used to optimise various polymeric solid dosage forms in terms of composition and process conditions; for example during hot melt extrusion or coating processes. While the use of complex viscosity might not be the ideal function to represent the behaviour of the material in the glassy phase, it has been demonstrated here to be a reliable metric to span over the various phases. In future other rheological and mechanical functions could provide a more meaningful Combining the thermal and rheological data reveals that both polymers contain approximately 3-5% moisture content. This water evaporates upon heating to 140°C for HPC and 200°C for HPMC. The moisture content plays a significant plasticising effect on HPC but very little effect is noticed on HPMC viscosity profile. The polymers do not seem to be fully miscible in the mixture. The mixtures indeed exhibit the individual thermal transitions behaviour of both polymers in a pattern consistent

with the fraction of each polymer in the mixture. The overall viscosity profile is though unique for each mixture.

From the master curves of complex viscosity values (k vs temperature) it can be concluded that increasing the temperature causes a tangible reduction in the viscosity of both HPC and HPC:HPMC 80:20 w/w films up until 150°C. On the contrary, films containing 50% w/w or more of HPMC do not show a significant reduction in their viscosity with temperature increase compared to HPC. In fact increasing the temperature from 120°C to 170°C for the HPC:HPMC 20:80 causes the viscosity to decrease by 50%.

On the other hand, HPC:HPMC 20:80 films show greater shear sensitivity suggesting that increasing the shearing rate for this blend might have more significant effect of reducing the viscosity than increasing the temperature during extrusion.

Finally, the study indicates that depending solely on the T_g value of the polymers using either DMA or DSC is not enough indication to the behaviour of the polymers at each temperature point before and after the T_g. This approach offers instead an enhanced insight into the behaviour of the polymer over a wider range making it easier to determine the ideal operating temperature and shear force range for a process.

Acknowledgement.

Composition fraction (%w/w)	
HPC	HPMC
100	0
80	20
50	50
20	80
0	100

Table 1. Composition of polymer mixtures in the compressed films in terms of the weight ratio of HPC and HPMC.

HPC:HPMC ratio (w/w)	Transition temperature(s) (°C)
1:0	138.7±0.5 , 215±1.1
8:2	140.3±0.4, 223.5±1.5
5:5	209.6±1.2
2:8	217.6±1.4
0:1	216.9±3.2

Table 2 Glass transition temperature of the HPC/ HPMC compressed films as measured by the DMA.

HPC:HPMC (w/w)	Correlation factor for K value K_{DMA}/K_{PPSR}	Correlation factor for n value n_{DMA}/n_{PPSR}
1:0	12	1.5
8:2	12	1.5
5:5	12	1.5
2:8	4	1.5
0:1	9	1.5

Table 3: The ratio of the Power law parameters for the compressed polymer mixtures calculated from the complex viscosity data measured by both DMA and PPSR
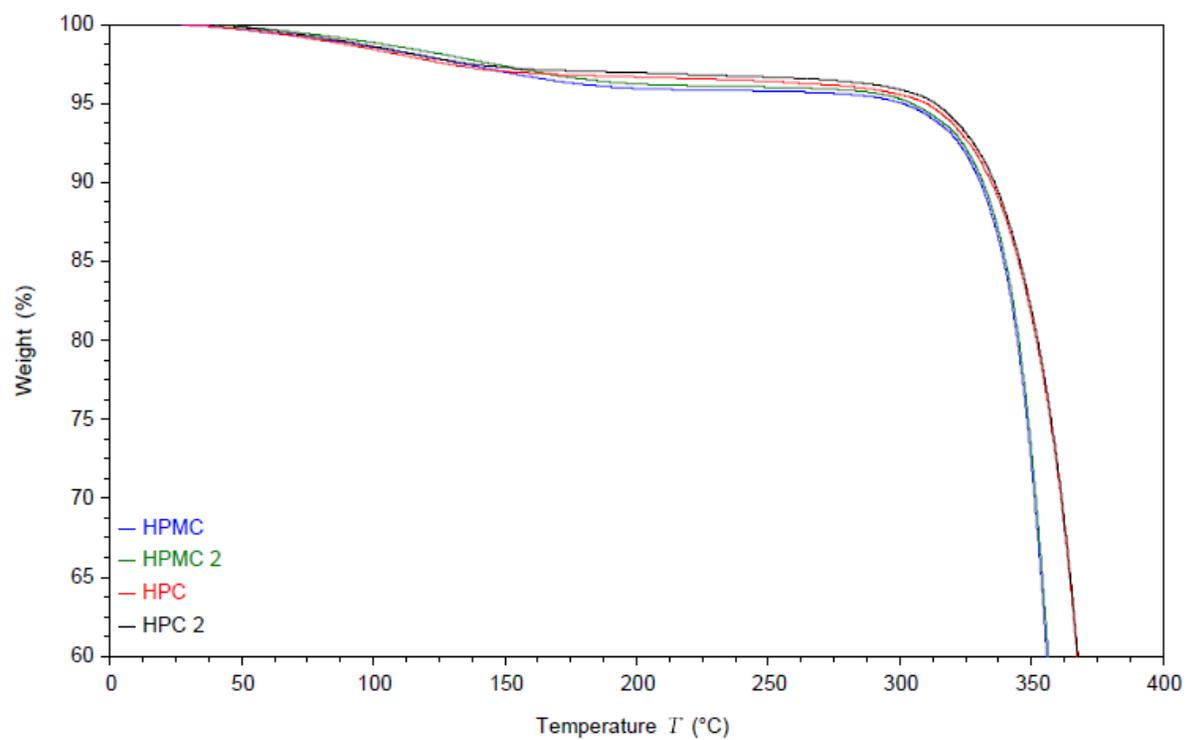
Figure 1. TGA thermogram of two HPC and two HPMC films and the powders. Samples named HPMC and HPC are measured from the powder while HPMC2 and HPC 2 are measured from the compressed films.
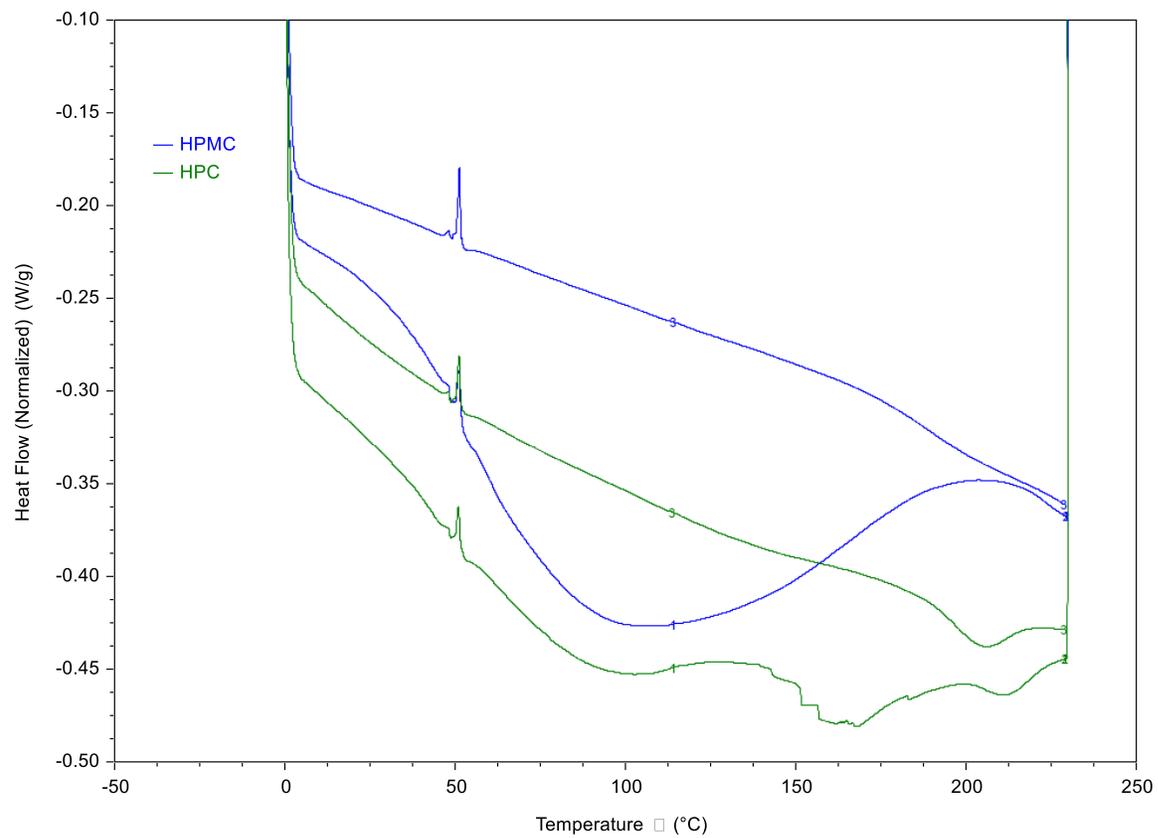
Figure 2 DSC thermograms of HPC and HPMC compressed samples using heat-cool heat cycle. The figure shows the two heating cycles. The cooling cycle is not shown.
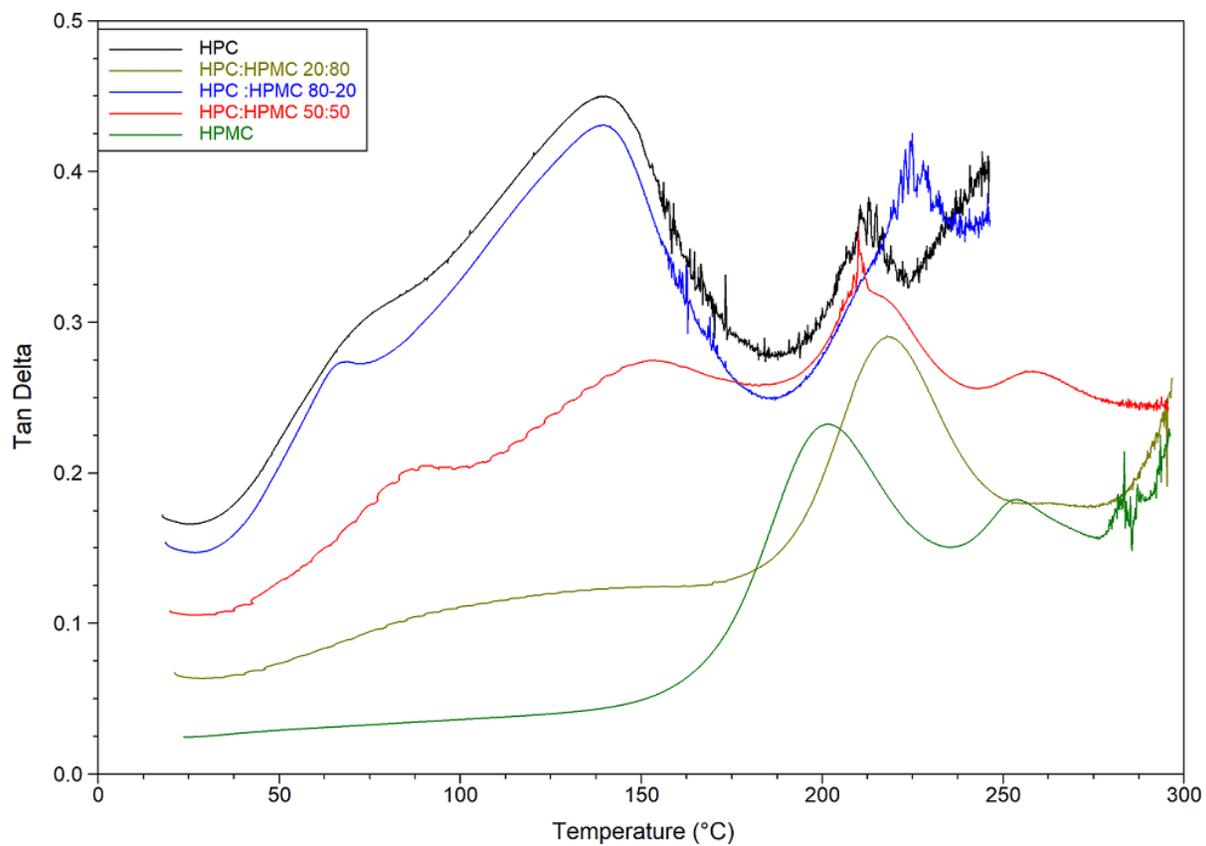
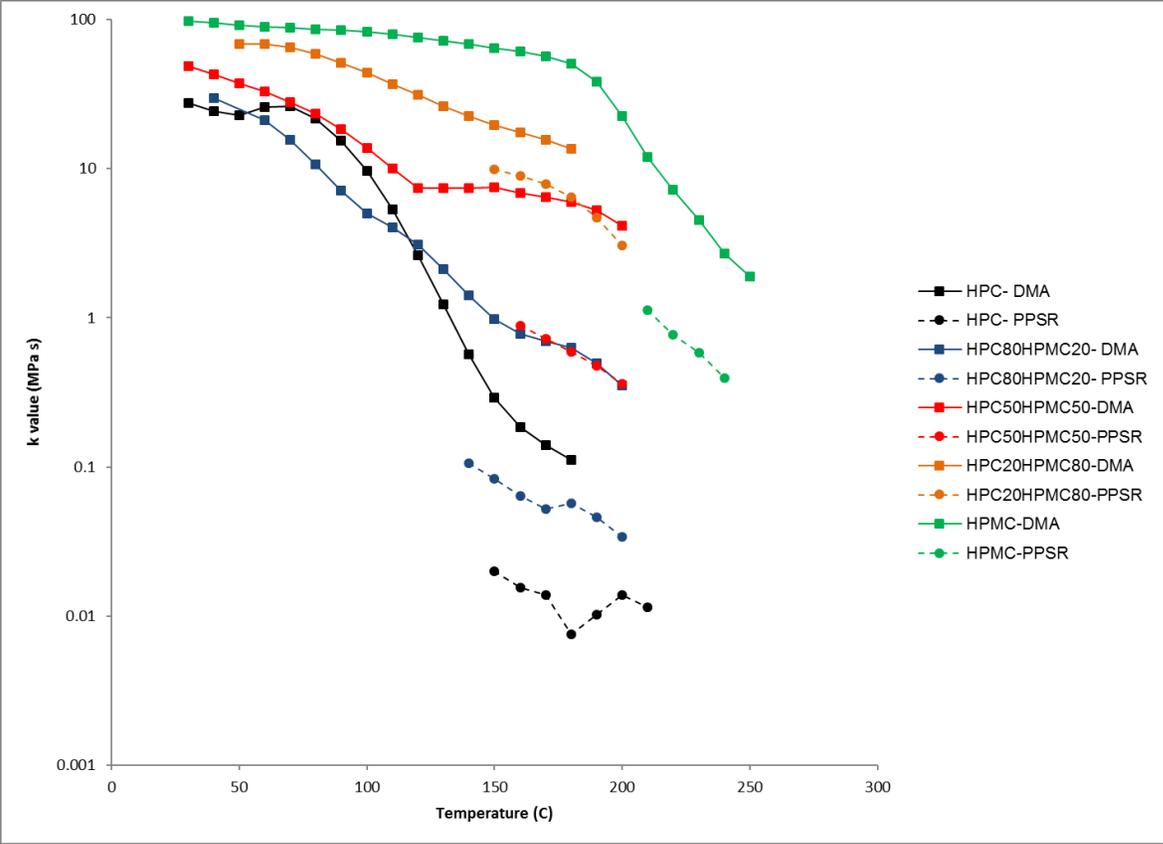Figure 3. Temperature sweep tan delta graphs of HPC, HPMC and their mixtures

Figure 4. Temperature dependency of the k value of the Power law model of the complex viscosity of the HPC and HPMC mixtures measured by DMA and PPSR
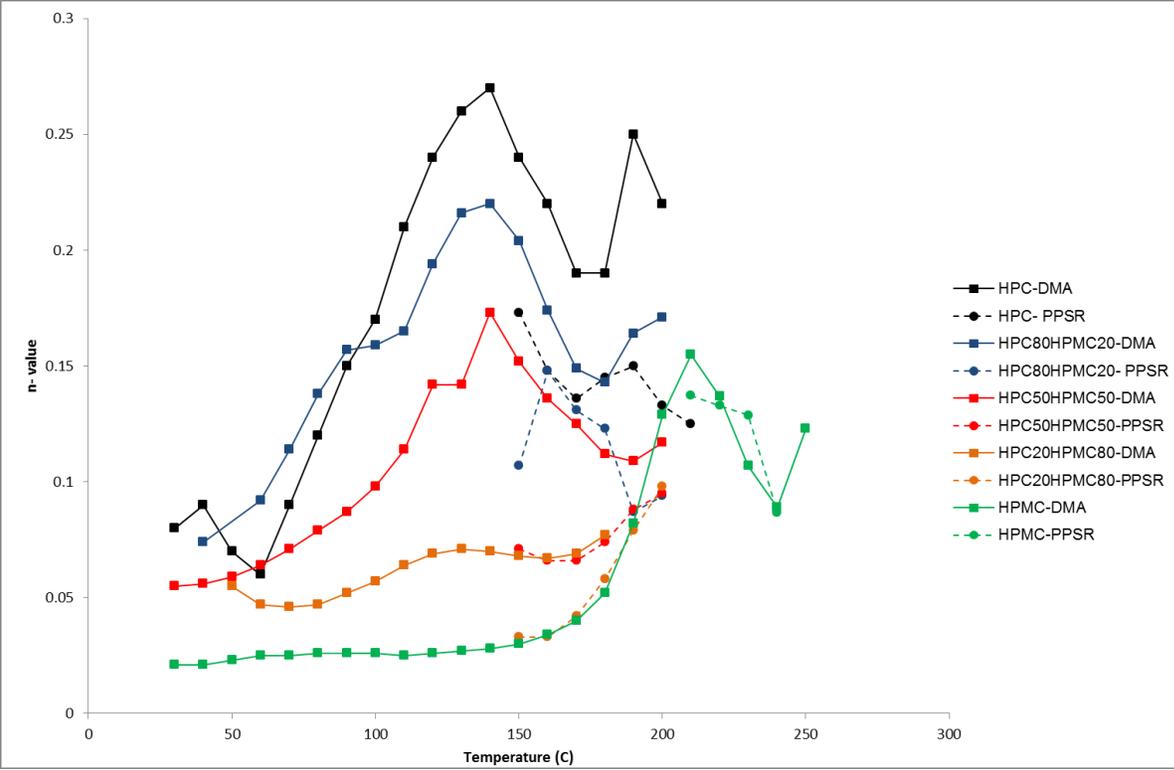
Figure 5. Temperature dependency of n value (shear sensitivity index) of the Power law model for HPC and HPMC mixtures measured by DMA and PPSR.
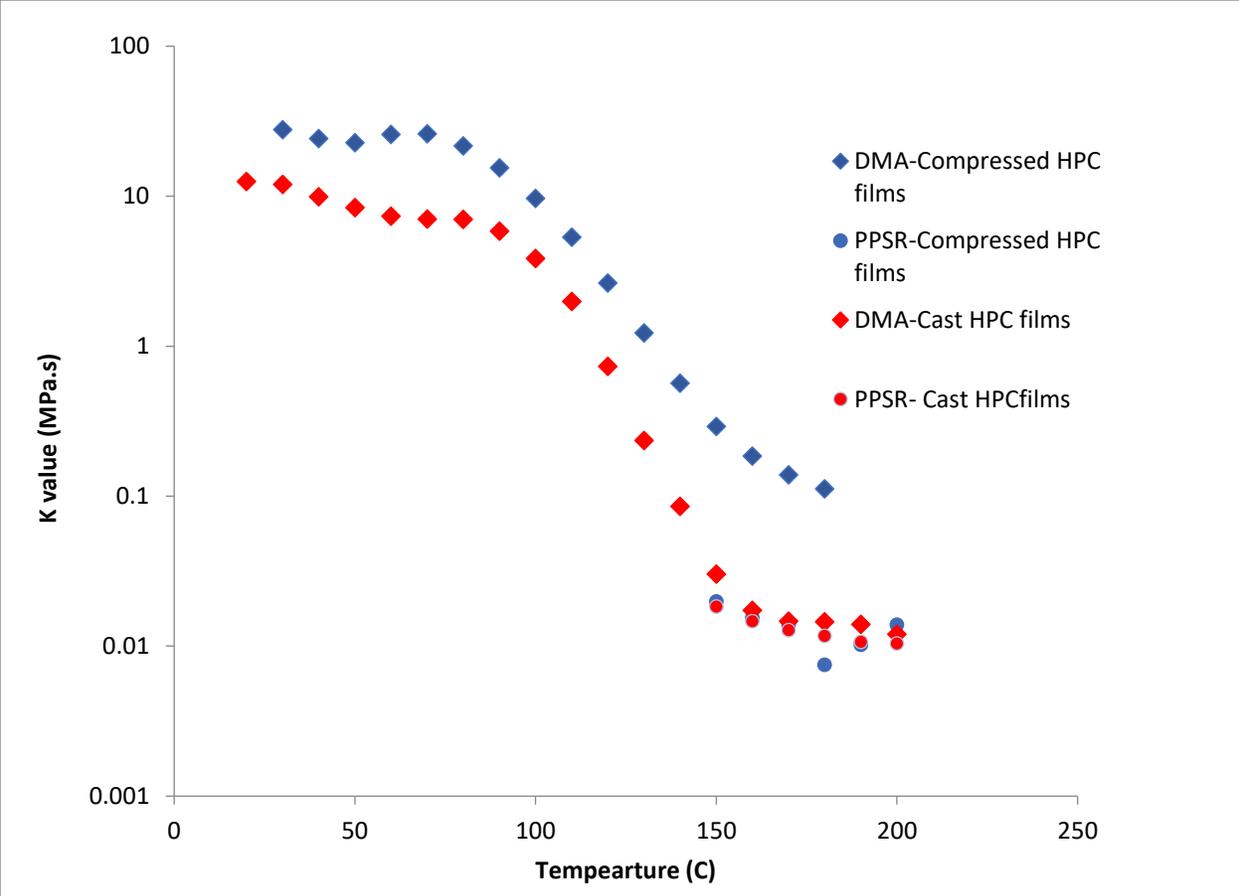
Figure 6. Comparison of K value of Power law model for compressed and cast HPC films
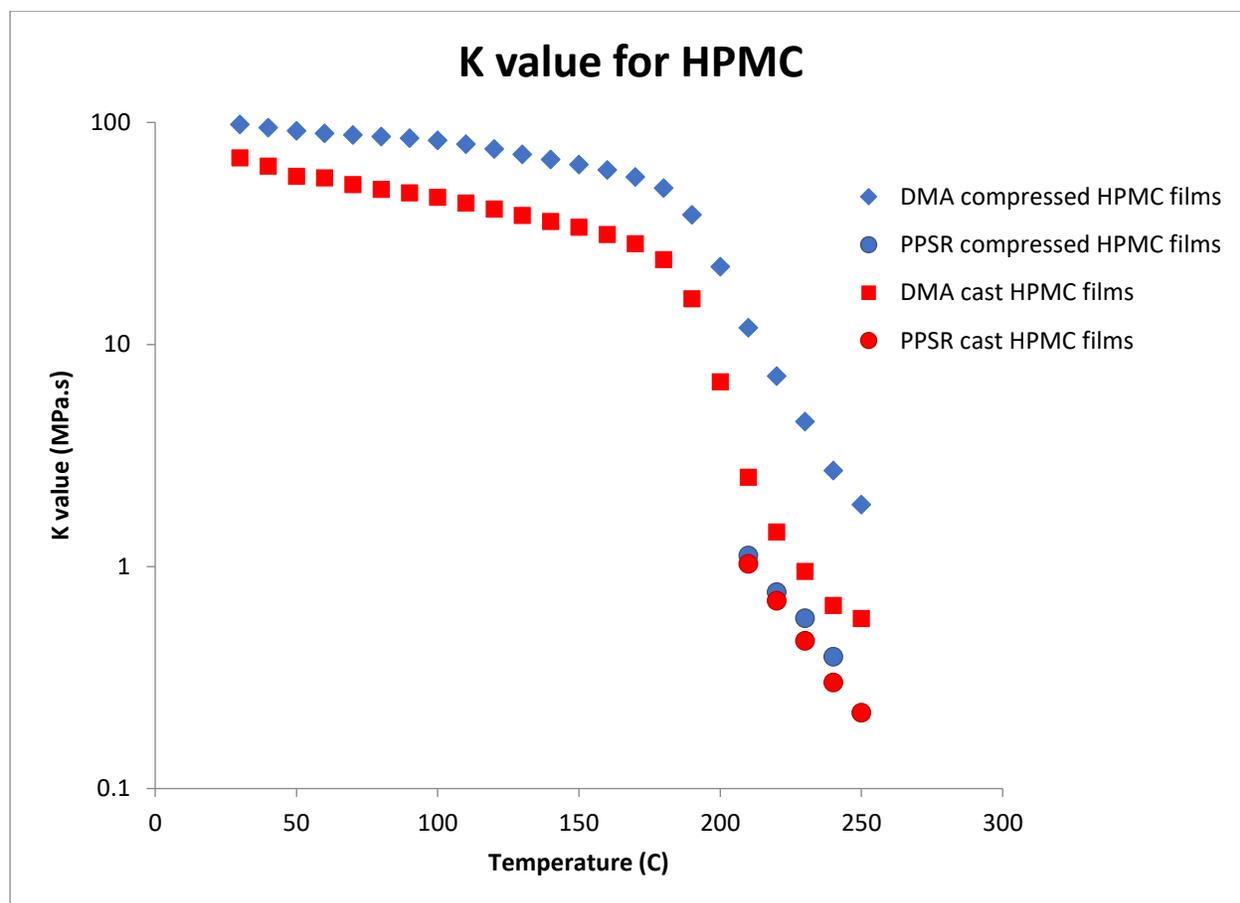
Figure 7. Comparison of K value of Power law model for compressed and cast HPMC films

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