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Precision cutting of glassy polymers: influence of aging on the cutting process

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ABSTRACT

This paper presents the results of an experimental study on the influence of aging on the cutting mechanics of glassy polymers. Polystyrene (PS), a glassy polymer, typically behaves brittle when subjected to a stress, it can be made ductile by rejuvenation. It was expected that PS would show a different cutting behaviour when it would be aged or rejuvenated. To investigate this two different molecular weight PS grades were used. Both aged and (mechanically) rejuvenated samples were made from each grade and cut. Cutting forces, chip morphology and surface quality were investigated. Although the chips showed no differences in brittleness and ductility, the measured cutting forces indicated that there is a difference between aged and rejuvenated PS. Also an interesting difference in cutting forces between the two PS grades was found. Investigation of the surface quality of the PS samples showed that the aged samples have smoother surfaces than the rejuvenated samples. It can be concluded that aging does have effect on the cutting mechanics and the obtained surface roughness.

Keywords: precision cutting, glassy polymers, aging, rejuvenating, chip formation, surface quality

1. INTRODUCTION

The lens producing industry becomes more and more interested in the production of polymeric lenses using single-point diamond turning. The reason for this is the higher accuracy that can be achieved by this process in relation to the conventional processes of grinding and polishing. If a lens can be manufactured on one machine the accuracy of the product improves. Another aspect of turning a lens is the possibility of producing astigmatic (non-rotational symmetric) lenses with higher accuracies using a fast tool servo. An example of such a lens is a multi focal lens that has different radii of curvature over its whole surface. In order to be able to produce astigmatic polymeric lenses on one machine a research has been started to investigate the cutting mechanics of polymers and the diamond tool wear during the cutting of these polymers. Also, a long stroke 10 mm fast tool servo is being developed to be able to machine astigmatic surfaces. This article gives results on the experimental investigation of the cutting mechanics of glassy (amorphous) polymers.

2. MATERIAL BEHAVIOUR

This section gives a short introduction to the material behaviour of polymers, since the material behaviour of polymers cannot be completely compared to the material behaviour of metals.[1] First the general material behaviour of polymers is discussed and then several methods to influence the macroscopic material response of polymers. Section 3 will then describe the effects on the cutting process when changing the macroscopic material behaviour of a polymer.

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2.1. General material behaviour

In principle all polymers are intrinsically tough, provided that their molecular weight is sufficiently high to form a sufficiently strong entangled polymer network.[2] However, the macroscopic response of polymers can be brittle. In the left graph of figure 1 the mechanical response of both polystyrene (PS) and polycarbonate (PC) are given during uniaxial deformation. It can be seen that PS fractures shortly after a tensile stress has been
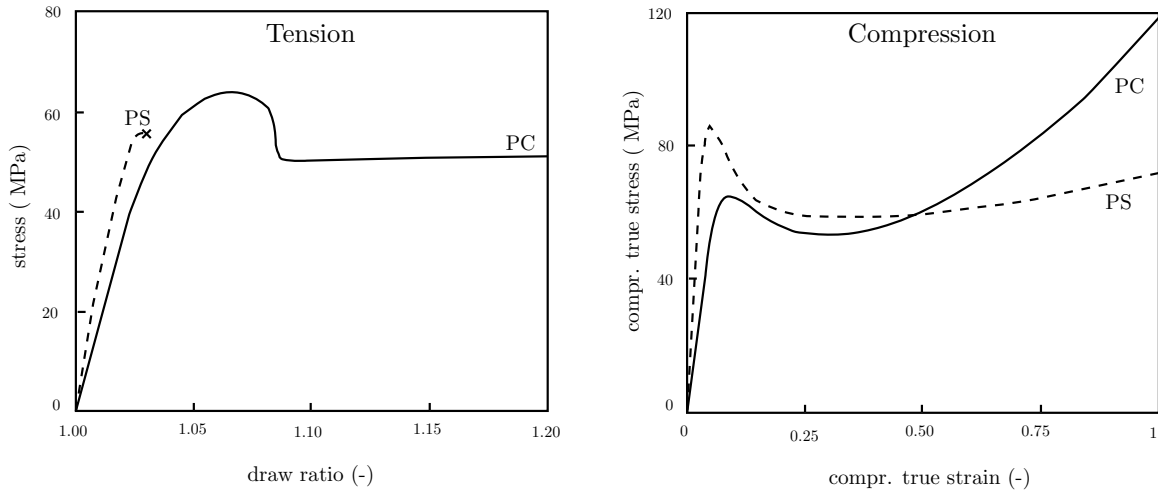
Figure 1: Tensile and compressive material behaviour of PS en PC (source v. Melick[2]).

applied. In fact, PS forms so called "crazes" during tension. During loading these crazes, which are bridged by fibrils, open up and macroscopic failure occurs by failure of these fibrils. That the PS behaved tough can be seen by the fact that these fibrils consist of highly stretched material, close to the maximum network draw ratio. Looking at PC in a tension test, one can see that PC (generally) behaves ductile. During tension, a stable neck is formed in the material and with increased loading the neck will grow further until ductile fracture occurs at a macroscopic strain of approximately 100%. The intrinsic material behaviour of polymers can be visualized by a test method where necking and crazing are absent. A method that can be applied is uniaxial compression. Doing this for PS and PC, one gets the results as depicted in the right graph of figure 1. From this graph it becomes clear that both PS and PC show the same intrinsic material behaviour. The intrinsic material behaviour can be described by a visco-elastic region, followed by yielding, strain softening and strain hardening. The main differences between PS and PC can be found by their post-yield behaviour: (1) by the drop in true stress after initial yielding (this is called "yield-drop") and by the slope of the strain hardening at larger strains. Differences in this post-yield behaviour give rise to differences in the macroscopic behaviour of a polymer, i.e. whether it behaves brittle or ductile. This difference in macroscopic response will be explained by the right graph of figure 1. It can be seen here that PS shows a relatively large yield-drop compared to PC. Also PS shows less strain hardening when strained further. Because PS has large strain softening and only little strain hardening, the result is that during tension the strain will be localized. Strain localisation is induced by intrinsic strain softening, but the evolution of the localized plastic zone depends on the stabilizing effect of the strain hardening. Since PS shows only little strain hardening, the localized plastic zone cannot be stabilized. This means that by further deformation in tension, the deformation will be localized more and more in this region, resulting in void nucleation and craze formation, resulting finally in a macroscopic brittle fracture. Polycarbonate, on the other hand, has relatively small strain softening and a stronger strain hardening than PS. This strain hardening results in a stabilizing effect after strain localisation and this results in a stable deformation process during tension.

It can be concluded that the amount of strain softening determines the macroscopic response of a polymer. The next paragraph describes that the amount of strain softening is dependent on the thermo-mechanical history of a polymer.

2.2. Influencing the intrinsic material behaviour

When a polymer is being cooled from the melt (above its glass transition temperature T_g) to just above T_g it is in a thermodynamic equilibrium. When it is cooled further, around T_g the polymer chains can no longer move fast enough to stay in a thermodynamic equilibrium.[3] The result is that the polymer changes from a liquid state to a solid glassy state. Since the polymer chains can no longer move fast enough, a free volume between the chains occurs when cooled further below T_g. The amount of free volume is susceptible to time. The reason for this is that the free volume gives part of chains the possibility to move. Since the polymer strives for a thermodynamic equilibrium the polymer chains will keep on moving and the free volume will decrease. This process is called "aging". The next paragraphs describe methods to influence the macroscopic material response as a function of thermo-mechanical treatments.

2.2.1. Aging

Aging does not only have influence on the free volume of a polymer, but it also influences the mechanical response of a polymer. Figure 2 shows the intrinsic material behaviour of PC while it ages. It can clearly be
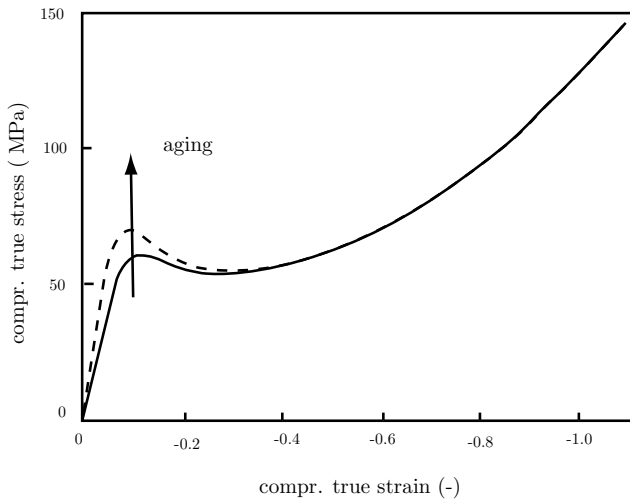
Figure 2: Influence of aging on the yield stress and yield-drop (source v. Melick[2]).

seen that the initial yield stress increases during aging and that the post-yield behaviour remains the same. A consequence of this is that the yield-drop increases during aging. It can be seen that strain hardening is not influenced by aging. Paragraph 2.1 described that the amount of yield drop in combination with the amount of strain hardening is responsible for a brittle or ductile behaviour. Therefore, it may be clear that aging can change the macroscopic response of a material from ductile to brittle (since the yield-drop increases).

Aging of a polymer can be done by heating a polymer to a temperature just below its T_g for a short time (days or weeks) and then slowly cool it again to room temperature. The result is that the chains are able to relax during this heat treatment (due to their strive for thermodynamic equilibrium). This behaviour is also described by the Arrhenius equation:

$$\tau = c \cdot \exp A/RT \tag{1}$$

with τ the relaxation time, i.e. the time for a polymer to age, and T the temperature. In creep research this is a common technique that gives the possibility of investigating the material behaviour in several short periods at several temperatures to find the real material behaviour over a longer period of time at a certain temperature.[3]

2.2.2. Rejuvenation

Polymers can also be rejuvenated. There are actually two methods for this: (1) a heat treatment called "quenching" and (2) a mechanical pre-deformation. Quenching means that a polymer is heated slightly above

T_g and is then rapidly cooled to room temperature. As can be seen in figure 2 (solid line) this will result in a lower initial yield stress and smaller yield-drop during deformation. The second option for rejuvenating a polymer is mechanical pre-deformation. In this case a polymer is subjected to a deformation prior to subjecting it to the real deformation. It is known that PS can be rejuvenated by rolling it and giving it a thickness reduction of approximately 30%, while a PC cilinder can be torqued to and fro over an angle of 720° to rejuvenate it. Figure 3 shows what happens to the mechanical response of PS when it is being rejuvenated by
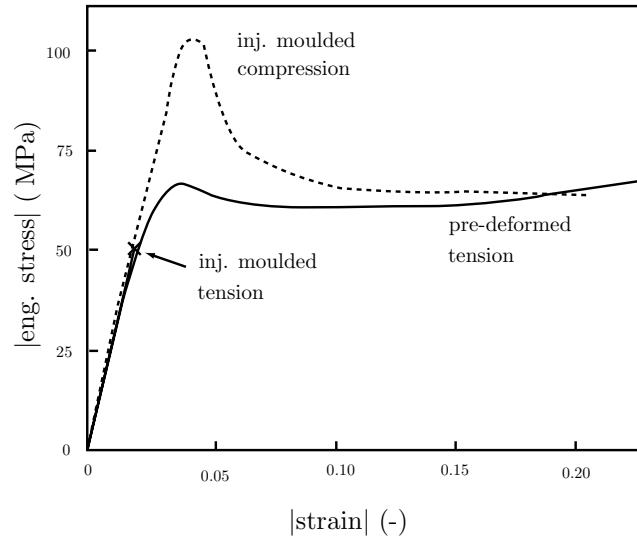

Figure 3: Deformation of PS in tension, compression and pre-deformed in tension (source v. Melick[2]).

pre-deformation. The figure shows that an injection moulded sample during tension fails at a small strain. It also shows the intrinsic material behaviour of this PS (high initial yield, pronounced strain softening and weak strain hardening) when subjected to a compression test. The figure shows that when the PS gets rejuvenated and subsequently deformed in an uniaxial tensile test it behaves in a ductile way and shows a long elongation. Thermo-mechanical rejuvenation leads to a reduction of the intrinsic softening of a polymer. This results in a ductile behaviour in tension. It can be said that rejuvenation erases the effects of physical aging. This is visualized by figure 4 where it can be seen that an injection moulded PC sample necks and deforms ductile in tension, figure (a). When such a sample is aged prior to tension it can be seen in figure (b) that it fractures in a brittle way. If the sample is first annealed and then thermally rejuvenated, it shows stable neck growth during tensile deformation again, figure (c).

2.3. Molecular weight

In the previous paragraphs (2.1 and 2.2) the intrinsic material behaviour and its manipulation was described. It was described how and when a polymer behaves brittle or ductile. In these paragraphs it was mentioned that the amount of yield-drop and strain hardening determine whether a polymer behaves brittle or ductile. The amount of yield-drop can be influenced by thermo-mechanical treatments, but the strain hardening of the polymer is not influenced by this. The amount of strain hardening is influenced by the molecular network of the polymer.[2] Some properties, such as impact strength and shear melt flow, are related to the molecular weight.[5] The following paragraph will describe how some material properties are related to these numbers and then a paragraph will describe the influence of molecular weight on the mechanical response of polymers.

Molecular weight and processing Every polymer can be made with different chain lengths and different chain length distributions, leading to different material behaviour. Polymers can be qualified by their molecular weight and their molecular weight distribution. The reason for this is that some properties, such as impact
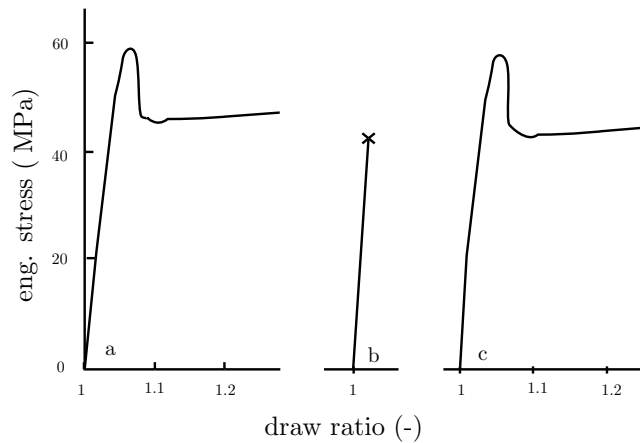
Figure 4. Effect of thermal treatments on the macroscopic behaviour during tensile deformation: (a) necking in injection moulded PC, (b) embrittlement after aging and (c) renewed necking after thermal rejuvenation (source v. Melick[2]).

strength and shear melt flow, are related to molecular weight.[5] Impact strength is determined by the number of chain ends (in essence the more short molecules), while the shear melt flow of a polymer is influenced by the midsection of a polymer. In this article two molecular weight numbers will be used: M_n, the number-average molecular weight and M_w, the weight-average molecular weight. Generally the impact strength of a polymer is better when it has shorter chains, in essence a higher M_n value. In general where larger deformations, as in extrusion processes (above T_g), are concerned M_w is important, i.e. the weight to be transfered.[5] Higher M_w results in a higher shear viscosity. The molecular weight distribution is the ratio of M_w and M_n ($PDI = M_w/M_n$) and it gives a representation of poly-dispersity of the polymer. A value of $PDI = 1$ means that the polymer is mono-disperse i.e. has only one chain length.

Mechanical response The network density of a polymer is determined by the amount of physical or chemical entanglements.[2] When a polymer grade has longer chains the contribution of these chains to the entanglement network is bigger and this will result in a different strain to break. Figure 5 shows this for the mechanical response of two different PS grades. Styron 648 has longer chains then Styron 660. So, when these two grades are subjected to a deformation (and when they have the same age) it can be seen that the higher molecular
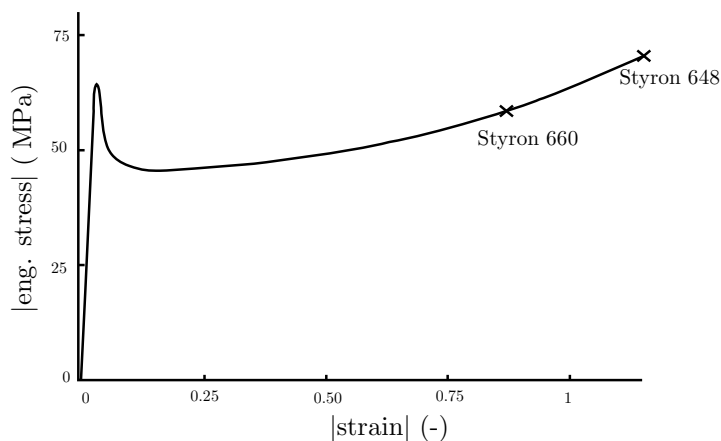
Figure 5: Schematic representation of the intrinsic tensile response of Styron 648 and Styron 660 (source v. Melick[2]).

weight grade (Styron 648) has a longer strain to break. Thus, molecular weight does not influence the amount of yield-drop, but it does influence the amount of strain hardening. The only relation that molecular weight has in respect to brittle-ductile transitions is in the time when aging has increased the initial yield stress sufficiently for the stabilizing effect of strain hardening to disappear. This means that a low molecular weight grade ages faster then a high molecular weight grade. From van Melick[2] it follows that Styron 660 shows renewed brittle fracture (after mechanical rejuvenation) after 1 hour, while Styron 648 shows renewed brittle fracture after 1 week.

3. INFLUENCE OF AGING ON THE CUTTING PROCESS

From section 2 it became clear that the age of PS is of importance for its macroscopic material behaviour. Although in cutting a shear deformation exists, the intrinsic material properties described in the previous section are still valid for this kind of deformation . It is known that PS behaves ductile when it is rejuvenated and brittle when it is aged (paragraph 2.2.1). The question arises whether aging has effect on the chip formation during cutting of PS. This is interesting to know, since generally good surface qualities are achieved when cutting in a ductile regime.[6] Therefore it is interesting to know whether brittle or ductile chip formation occurs when cutting aged respectively rejuvenated PS.
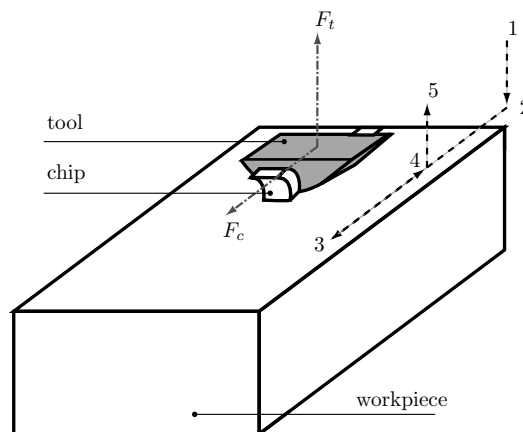
3.1. Experimental

The materials used in these experiments were general purpose polystyrene (PS) Styron 648 and Styron 660 from Dow Plastics (Dow Chemical Company). Table 1 shows the molecular weights of these two grades. From this table it can be concluded that the Styron 648 has longer chains than the Styron 660. The materials were

Table 1: Molecular weights of the used PS grades

Material	M_w(g/mol)	M_n(g/mol)	PDI
Styron 648	318.490	107.215	2,97
Styron 660	203.401	65.822	3,09

injection moulded to plates with a thickness of 3 mm and squareness of 60x60 mm^2 . Half of the samples were cut into workpieces of approximately 20x20 mm^2 and then placed into an oven for aging at 80 °C for three days and then slowly cooled to room temperature. The rest of the injection moulded plates were rolled with a thickness reduction of 30% in order to mechanically rejuvenate them. After the rolling the plates were cut into workpieces of approximately 20x20 mm^2 and immediatly cut, since aging of PS is on a minutes time-scale.[4] The

Figure 6: Tool movements for the experiments shown by the dotted arrows.

idea of the used experimental setup for cutting originated form Arcona and Dow's work.[7] The idea is to have the spindle fixed and make the cutting movement with the carriage. This is schematically depicted in figure 6. Disadvantage of this approach is that only relatively low cutting speeds can be achieved when compared to conventional cutting speeds (typically in the order of m/s). In these experiments a cutting speed (also top speed of the carriage) of 2 mm/s is used. The tool is moved to a certain depth of cut (from point 1 to 2) and then moved in one stroke to position 3 (stroke of 15 mm). A dwell time is used before the tool is moved backward for 7 mm (point 4) before retracting it out of the workpiece (point 5). Since it is not known what the effect of the cutting process is on the newly created surface (sub-surface damage may be present), the workpieces are not precision turned to a flat surface prior to the actual experiment (this in contrast to the work of Arcona and Dow). Another reason for not doing this is that the aging process would go on while the surface would be turned flat. This turning to a flat surface would take several minutes and the rejuvenated PS would have become aged by that time.[2] The main cutting force F_{cut} and the thrust force F_{thrust} are measured during the experiments with a 3D-component (Kistler 9251A) force transducer. Cuts were made on 10 workpiece samples for each material and material condition.

3.2. Results

3.2.1. Force measurements

Since the surface is not flat during the experiment, the actual depth of cut changes during the cutting, but was mostly between 20 μm and 30 μm . The measured cutting forces are therefore divided by the depth of cut of the formed groove[†] (forces for depths of cut from $20 - 30$ μm have been evaluated). Figure 7 depicts the measured main cutting force F_{cut} and the trust force F_{thrust}. This figure shows the mean value plus/minus one standard deviation of the measurement data for all samples.
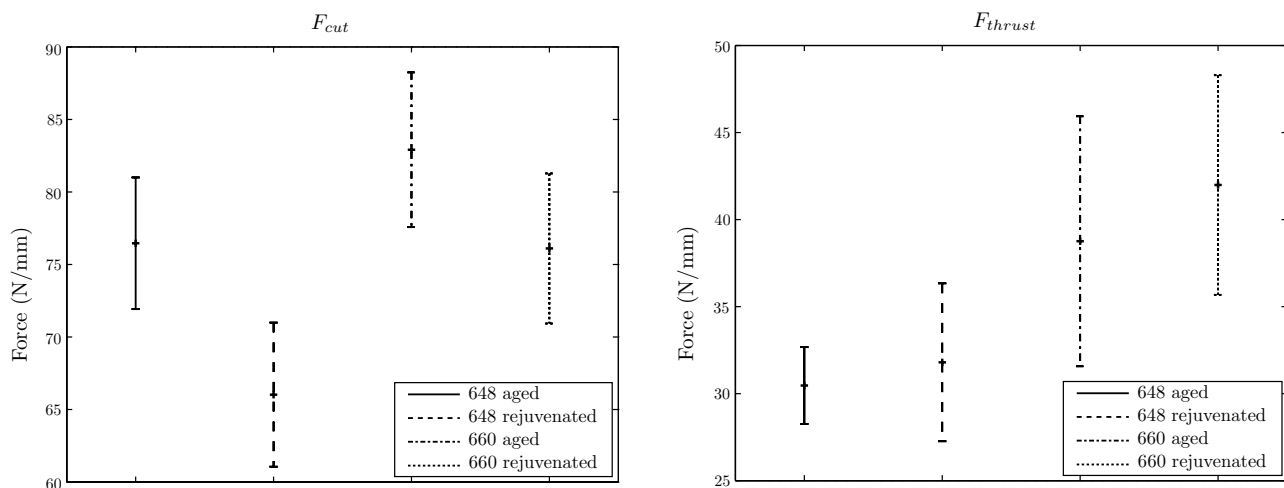

Figure 7: Measured cutting forces of aged and rejuvenated PS grades.

Main cutting force In the left graph it can be seen that the cutting force decreases as the material was rejuvenated. This is both true for Styron 648 and Styron 660. The fact that the measured force data of the rejuvenated Styron 660 grade is shifting into the force data of the aged one can be attributed to the fact that Styron 660 ages more quickly than Styron 648.[2] The difference in main cutting force between Styron 648 and Styron 660 will be discussed in section 4.

[†]The assumption here is that the amount of spring-back is independent of the actual depth of cut. Depth of cut was determined by measuring the groove width afterwards.

Thrust force The right graph of figure 7 shows the thrust forces during cutting. It can be seen that the mean values of the measurement data for the aged samples and rejuvenated samples differ a bit, but the mean values are within the standard deviations of each grade. A striking difference between thrust forces can be noticed between the two molecular weight grades. Styron 660 has higher trust forces than Styron 648. A possible explanation of this may be that there is more material transport underneath the tool for Styron 660. This would cause an extra force in the thrust direction. If this would happen, this can be seen in the reverse stroke of the tool. More material will be transported underneath the tool in the reverse stroke and more material transport will result in a higher (thrust) force. Figure 8 shows two measurements of an aged Styron 648 and Styron 660 workpiece. The reverse stroke runs from approximately $10 - 14$ s and it can be seen that the force of the Styron 660 workpiece indeed has a higher value. This probably means that more material is transported underneath the tool when cutting Styron 660.[7]
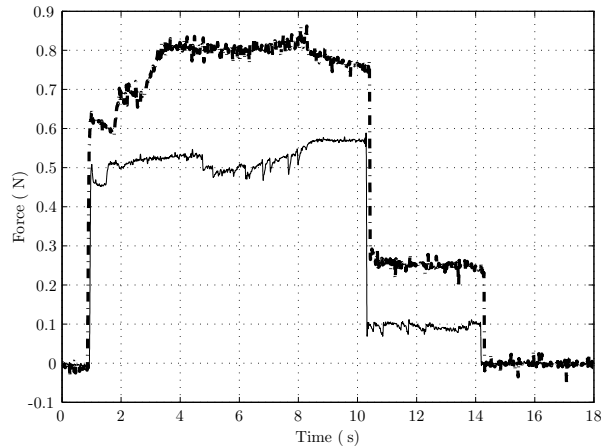

Figure 8: Measured thrust forces of aged Styron 648 (solid) and aged Styron 660 (dash-dot).

Chip formation As described before PS can show brittle and ductile behaviour depending on its age (paragraph 2.2.1). It was expected that this difference in behaviour could also be seen in the type of chip being formed, i.e. brittle or ductile chips. To investigate this, scanning electron microscopy (SEM) pictures were taken of the formed chips. However, these images showed no differences in brittleness between aged and rejuvenated samples. All chips that were formed were of the continuous ductile type.

3.2.2. Surface quality

Figure 9 shows the structures of the formed surface in the middle of the cut of an aged and a rejuvenated sample of Styron 648 measured by atomic force microscopy (AFM). It appears that the aged sample has a smoother surface than the rejuvenated sample. The aged sample shows marks in the cutting direction. Typical in this picture is the different surface at the bottom of the groove. It appears as if the workpiece material sticks more to the clearance face in this region of the tool. The reason for this is not known at this moment. The rejuvenated samples clearly shows a rough surface with a lot of flaws. From scanning electron microscopy (SEM) images these flaws seemed to have formed by sticking of the cut surface to the clearance plane of the tool. In comparison to the aged sample the absence of the tool marks is clear. Rather strange is the presence of the vertical strokes in the figure of the rejuvenated sample. The origin of this is not known at this moment, but it may be caused by machine vibration. Although the aged sample shows a smoother surface than the rejuvenated sample it is not known at this moment what the influence of aging is on the surface quality during normal cutting conditions (approximately 3 m/s instead of 2 mm/s, which was used in these experiments). Further experimental work will be performed to determine this.
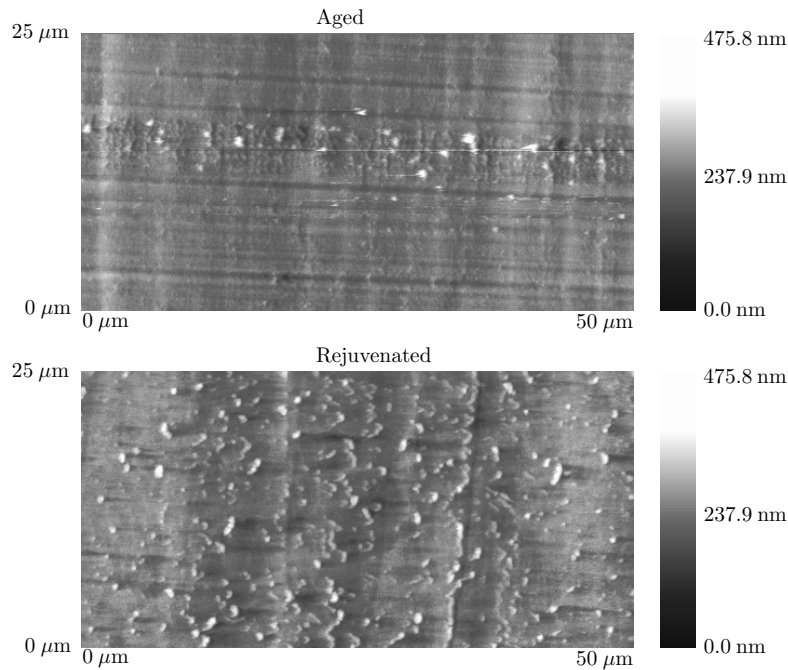
Figure 9. Surfaces of aged respectively rejuvenated Styron 648 taken in the middle of the groove. Cutting direction was from right to left.

4. DISCUSSION

In the previous paragraph the results have been presented for cutting of aged and rejuvenated PS. Although a difference exists between the measured cutting forces, a clear difference in brittle or ductile chip formation is absent. Also the forces when cutting Styron 648 and Styron 660 were presented in figure 7. From this figure it becomes clear that there is a big difference in cutting forces between the low and high molecular weight PS. The next paragraphs discuss some material parameters that may explain the difference in main cutting forces.

Strain to break   Looking at figure 5 one can see that the strain to break for low molecular weight grades is smaller than for high molecular weight grades. Styron 648 respectively Styron 660 has a failure stress of approximately 70 MPa respectively 60 MPa. Since Styron 660 has a lower failure stress one would expect a lower cutting force, which is not the case.

Melt viscosity   Several papers mention the possibility of cutting above the glass transition temperature (T_g) due to adiabatic heating of the material in the shear zone.[8, 9]   The dependence of cutting force on the age of the polymer indicates that cutting occurs in the solid state of the polymer. Another aspect for assuming that the cutting does not appear in a temperature region above T_g is the fact that, if this would happen, Styron 660 would have a lower shear flow viscosity based on its lower molecular weight M_w (paragraph 2.3 and product information of Styron[10]). A lower shear viscosity would result in lower cutting forces. Therefore cutting in the "melt" can be excluded.

Initial yield stress   Paragraph 2.2.1 showed that aging increases the initial yield stress of a polymer. Figure 7 shows that the main cutting force also increases by aging the workpiece prior to cutting. This would imply that initial yield stress could be a dominant factor in the cutting process, but if Styron 648 and Styron 660 are compared something different can be concluded. Paragraph 2.3 showed that the influence of molecular weight does not influence the initial yield behaviour (see figure 5), so if the initial yield behaviour would be a

dominant factor, no difference in main cutting forces should have appeared between annealed Styron 648 and annealed Styron 660. The next paragraph shows that initial yield stress may be a dominant factor in the cutting mechanics of polymers and that the difference in forces between Styron 648 and Styron 660 originates from a different aspect.

Contact area Because the two Styron grades have the same initial yield stress when deformed, one would expect the same cutting forces while cutting these grades. Unfortunately this is not the case and another explanation is needed to explain the difference and show that initial yield may be a dominant factor in the cutting process. This explanation can be found in the plastic deformation zone when cutting. From single asperity friction measurements on PS it is known that a low molecular weight PS has a higher friction force than a high molecular weight PS.[11] Literature shows that the low molecular weight PS has a reduced glass transition temperature (5 K). This means that the low molecular weight PS chains posses a higher thermal mobility resulting in an easier plastic deformation. Easier plastic deformation will lead to a larger contact area.[11] A larger contact area will result in higher cutting forces. Looking again at the question whether initial yield stress is a significant factor in the description of the cutting process of polymers, it can be stated, taking into account the influence of contact area, that initial yield stress is significant.

5. CONCLUSIONS

Although it is known from polymer mechanics that PS can behave in a macroscopically brittle or ductile way if it is aged or rejuvenated, such a difference has not been observed during cutting experiments. Cutting experiments were performed on two different molecular weight PS grades and although a clear difference between brittle and ductile was absent a difference in cutting forces was found. Aged samples have higher cutting forces than rejuvenated samples. This indicates that initial yield stress has an important contribution in the description of the cutting mechanics of polymers. Although a difference in cutting forces between Styron 648 and Styron 660 could contradict this, a good explanation for this difference is found in the size of the contact area and not in the possibility of cutting in the "melt" as suggested by some researchers.[8,9] Although the difference in size of contact between low and high molecular grade can be found in literature,[11] it still has to be verified for cutting mechanics. Besides cutting forces also the surfaces of aged and rejuvenated PS were investigated and it was shown that the aged surface is smoother than the rejuvenated surface.

ACKNOWLEDGMENTS

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