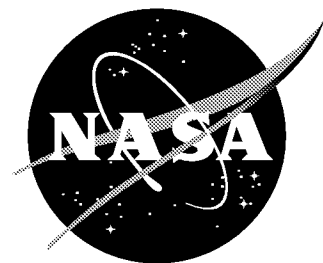


NASA/TM-1999-209359



# Design Requirements for Amorphous Piezoelectric Polymers

*Z. Ounaies*

*Institute for Computer Applications in Science and Engineering (ICASE)  
Langley Research Center, Hampton, Virginia*

*J. A. Young*

*University of Virginia, Charlottesville, Virginia*

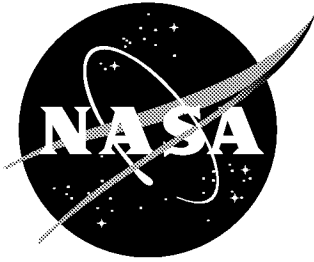
*J. S. Harrison*

*Langley Research Center, Hampton, Virginia*

---

August 1999

NASA/TM-1999-209359



# Design Requirements for Amorphous Piezoelectric Polymers

*Z. Ounaies*

*Institute for Computer Applications in Science and Engineering (ICASE)  
Langley Research Center, Hampton, Virginia*

*J. A. Young*

*University of Virginia, Charlottesville, Virginia*

*J. S. Harrison*

*Langley Research Center, Hampton, Virginia*

---

August 1999

## The NASA STI Program Office ... in Profile

Since its founding, NASA has been dedicated to the advancement of aeronautics and space science. The NASA Scientific and Technical Information (STI) Program Office plays a key part in helping NASA maintain this important role.

The NASA STI Program Office is operated by Langley Research Center, the lead center for NASA's scientific and technical information. The NASA STI Program Office provides access to the NASA STI Database, the largest collection of aeronautical and space science STI in the world. The Program Office is also NASA's institutional mechanism for disseminating the results of its research and development activities. These results are published by NASA in the NASA STI Report Series, which includes the following report types:

- **TECHNICAL PUBLICATION.** Reports of completed research or a major significant phase of research that present the results of NASA programs and include extensive data or theoretical analysis. Includes compilations of significant scientific and technical data and information deemed to be of continuing reference value. NASA counterpart of peer-reviewed formal professional papers, but having less stringent limitations on manuscript length and extent of graphic presentations.
- **TECHNICAL MEMORANDUM.** Scientific and technical findings that are preliminary or of specialized interest, e.g., quick release reports, working papers, and bibliographies that contain minimal annotation. Does not contain extensive analysis.
- **CONTRACTOR REPORT.** Scientific and technical findings by NASA-sponsored contractors and grantees.

- **CONFERENCE PUBLICATION.** Collected papers from scientific and technical conferences, symposia, seminars, or other meetings sponsored or co-sponsored by NASA.
- **SPECIAL PUBLICATION.** Scientific, technical, or historical information from NASA programs, projects, and missions, often concerned with subjects having substantial public interest.
- **TECHNICAL TRANSLATION.** English-language translations of foreign scientific and technical material pertinent to NASA's mission.

Specialized services that complement the STI Program Office's diverse offerings include creating custom thesauri, building customized databases, organizing and publishing research results ... even providing videos.

For more information about the NASA STI Program Office, see the following:

- Access the NASA STI Program Home Page at <http://www.sti.nasa.gov>
- E-mail your question via the Internet to [help@sti.nasa.gov](mailto:help@sti.nasa.gov)
- Fax your question to the NASA STI Help Desk at (301) 621-0134
- Phone the NASA STI Help Desk at (301) 621-0390
- Write to:  
NASA STI Help Desk  
NASA Center for AeroSpace Information  
7121 Standard Drive  
Hanover, MD 21076-1320

## The NASA STI Program Office ... in Profile

Since its founding, NASA has been dedicated to the advancement of aeronautics and space science. The NASA Scientific and Technical Information (STI) Program Office plays a key part in helping NASA maintain this important role.

The NASA STI Program Office is operated by Langley Research Center, the lead center for NASA's scientific and technical information. The NASA STI Program Office provides access to the NASA STI Database, the largest collection of aeronautical and space science STI in the world. The Program Office is also NASA's institutional mechanism for disseminating the results of its research and development activities. These results are published by NASA in the NASA STI Report Series, which includes the following report types:

- **TECHNICAL PUBLICATION.** Reports of completed research or a major significant phase of research that present the results of NASA programs and include extensive data or theoretical analysis. Includes compilations of significant scientific and technical data and information deemed to be of continuing reference value. NASA counterpart of peer-reviewed formal professional papers, but having less stringent limitations on manuscript length and extent of graphic presentations.
- **TECHNICAL MEMORANDUM.** Scientific and technical findings that are preliminary or of specialized interest, e.g., quick release reports, working papers, and bibliographies that contain minimal annotation. Does not contain extensive analysis.
- **CONTRACTOR REPORT.** Scientific and technical findings by NASA-sponsored contractors and grantees.

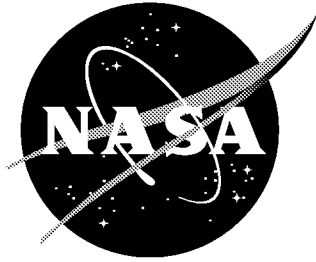
- **CONFERENCE PUBLICATION.** Collected papers from scientific and technical conferences, symposia, seminars, or other meetings sponsored or co-sponsored by NASA.
- **SPECIAL PUBLICATION.** Scientific, technical, or historical information from NASA programs, projects, and missions, often concerned with subjects having substantial public interest.
- **TECHNICAL TRANSLATION.** English-language translations of foreign scientific and technical material pertinent to NASA's mission.

Specialized services that complement the STI Program Office's diverse offerings include creating custom thesauri, building customized databases, organizing and publishing research results ... even providing videos.

For more information about the NASA STI Program Office, see the following:

- Access the NASA STI Program Home Page at <http://www.sti.nasa.gov>
- E-mail your question via the Internet to [help@sti.nasa.gov](mailto:help@sti.nasa.gov)
- Fax your question to the NASA STI Help Desk at (301) 621-0134
- Phone the NASA STI Help Desk at (301) 621-0390
- Write to:  
NASA STI Help Desk  
NASA Center for AeroSpace Information  
7121 Standard Drive  
Hanover, MD 21076-1320

NASA/TM-1999-209359



# Design Requirements for Amorphous Piezoelectric Polymers

*Z. Ounaies*

*Institute for Computer Applications in Science and Engineering (ICASE)  
Langley Research Center, Hampton, Virginia*

*J. A. Young*

*University of Virginia, Charlottesville, Virginia*

*J. S. Harrison*

*Langley Research Center, Hampton, Virginia*

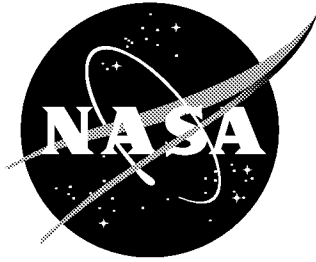
National Aeronautics and  
Space Administration

Langley Research Center  
Hampton, Virginia 23681-2199

---

August 1999

NASA/TM-1999-209359



# Design Requirements for Amorphous Piezoelectric Polymers

*Z. Ounaies*

*Institute for Computer Applications in Science and Engineering (ICASE)  
Langley Research Center, Hampton, Virginia*

*J. A. Young*

*University of Virginia, Charlottesville, Virginia*

*J. S. Harrison*

*Langley Research Center, Hampton, Virginia*

National Aeronautics and  
Space Administration

Langley Research Center  
Hampton, Virginia 23681-2199

---

August 1999

---

Available from:

NASA Center for AeroSpace Information (CASI)  
7121 Standard Drive  
Hanover, MD 21076-1320  
(301) 621-0390

National Technical Information Service (NTIS)  
5285 Port Royal Road  
Springfield, VA 22161-2171  
(703) 605-6000

---

Available from:

NASA Center for AeroSpace Information (CASI)  
7121 Standard Drive  
Hanover, MD 21076-1320  
(301) 621-0390

National Technical Information Service (NTIS)  
5285 Port Royal Road  
Springfield, VA 22161-2171  
(703) 605-6000



## Abstract

*An overview of the piezoelectric activity in amorphous piezoelectric polymers is presented. The criteria required to render a polymer piezoelectric are discussed. Although piezoelectricity is a coupling between mechanical and electrical properties, most research has concentrated on the electrical properties of potentially piezoelectric polymers. In this work, we present comparative mechanical data as a function of temperature and offer a summary of polarization and electromechanical properties for each of the polymers considered.*

## Introduction

Kawai's [1] pioneering work almost thirty years ago in the area of piezoelectric polymers has led to the development of strong piezoelectric activity in polyvinylidene fluoride (PVDF) and its copolymers with trifluoroethylene and tetrafluoroethylene. These semicrystalline fluoropolymers represent the state of the art in piezoelectric polymers. Research on the morphology [2-5], piezoelectric and pyroelectric properties [6-10], and applications of polyvinylidene fluoride [11-14] are widespread in the literature. More recently Scheinbeim et al. have demonstrated piezoelectric activity in a series of semicrystalline, odd numbered nylons [15-17]. When examined relative to their glass transition temperature, these nylons exhibit good piezoelectric properties ( $d_{31} = 17$  pC/N for Nylon 7) but have not been used commercially primarily due to the serious problem of moisture uptake. In order to render them piezoelectric, semicrystalline polymers must have a noncentrosymmetric crystalline phase. In the case of PVDF and nylon, these polar crystals cannot be grown from the melt. The polymer must be mechanically oriented to induce noncentrosymmetric crystals which are subsequently polarized by an electric field. In such systems the amorphous phase supports the crystalline orientation and polarization is stable up to the Curie temperature.

Nalwa et al. have also examined piezoelectricity in a series of polythioureas [18-19]. Though not highly crystalline, these thiourea polymers have a very high degree of hydrogen bonding which stabilizes the remanent polarization in such systems after poling.

The literature on amorphous piezoelectric polymers is much more limited than that for semicrystalline systems. This is in part due to the fact that no amorphous piezoelectric polymers have exhibited responses high enough to attract commercial interest. Much of the previous work resides in the area of nitrile substituted polymers including polyacrylonitrile (PAN) [20-22], poly(vinylidenecyanide vinylacetate) (PVDCN/VAc) [23-26], polyphenylethynitrile (PPEN) [27-28], and poly(1-bicyclobutanecarbonitrile) [29]. The most promising of these materials are the vinylidene cyanide copolymers which exhibit large dielectric relaxation strengths and strong piezoelectricity. The carbon-chlorine dipole in polyvinylchloride (PVC) has also been oriented to produce a low level of piezoelectricity [30,31]. Motivated by a need for high temperature piezoelectric sensor materials, NASA has recently begun research in the development of amorphous piezoelectric polymers. In this paper an amorphous, aromatic piezoelectric polyimide developed at NASA [32] is presented along with other amorphous and paracrystalline piezoelectric polymers shown in Table 1. The purpose of this overview is to explain the mechanism and key components required for developing piezoelectricity in amorphous polymers and to present a summary of polarization and electromechanical properties of currently

## Abstract

*An overview of the piezoelectric activity in amorphous piezoelectric polymers is presented. The criteria required to render a polymer piezoelectric are discussed. Although piezoelectricity is a coupling between mechanical and electrical properties, most research has concentrated on the electrical properties of potentially piezoelectric polymers. In this work, we present comparative mechanical data as a function of temperature and offer a summary of polarization and electromechanical properties for each of the polymers considered.*

## Introduction

Kawai's [1] pioneering work almost thirty years ago in the area of piezoelectric polymers has led to the development of strong piezoelectric activity in polyvinylidene fluoride (PVDF) and its copolymers with trifluoroethylene and tetrafluoroethylene. These semicrystalline fluoropolymers represent the state of the art in piezoelectric polymers. Research on the morphology [2-5], piezoelectric and pyroelectric properties [6-10], and applications of polyvinylidene fluoride [11-14] are widespread in the literature. More recently Scheinbeim et al. have demonstrated piezoelectric activity in a series of semicrystalline, odd numbered nylons [15-17]. When examined relative to their glass transition temperature, these nylons exhibit good piezoelectric properties ( $d_{31} = 17$  pC/N for Nylon 7) but have not been used commercially primarily due to the serious problem of moisture uptake. In order to render them piezoelectric, semicrystalline polymers must have a noncentrosymmetric crystalline phase. In the case of PVDF and nylon, these polar crystals cannot be grown from the melt. The polymer must be mechanically oriented to induce noncentrosymmetric crystals which are subsequently polarized by an electric field. In such systems the amorphous phase supports the crystalline orientation and polarization is stable up to the Curie temperature.

Nalwa et al. have also examined piezoelectricity in a series of polythioureas [18-19]. Though not highly crystalline, these thiourea polymers have a very high degree of hydrogen bonding which stabilizes the remanent polarization in such systems after poling.

The literature on amorphous piezoelectric polymers is much more limited than that for semicrystalline systems. This is in part due to the fact that no amorphous piezoelectric polymers have exhibited responses high enough to attract commercial interest. Much of the previous work resides in the area of nitrile substituted polymers including polyacrylonitrile (PAN) [20-22], poly(vinylidenecyanide vinylacetate) (PVDCN/VAc) [23-26], polyphenylethynitrile (PPEN) [27-28], and poly(1-bicyclobutanecarbonitrile) [29]. The most promising of these materials are the vinylidene cyanide copolymers which exhibit large dielectric relaxation strengths and strong piezoelectricity. The carbon-chlorine dipole in polyvinylchloride (PVC) has also been oriented to produce a low level of piezoelectricity [30,31]. Motivated by a need for high temperature piezoelectric sensor materials, NASA has recently begun research in the development of amorphous piezoelectric polymers. In this paper an amorphous, aromatic piezoelectric polyimide developed at NASA [32] is presented along with other amorphous and paracrystalline piezoelectric polymers shown in Table 1. The purpose of this overview is to explain the mechanism and key components required for developing piezoelectricity in amorphous polymers and to present a summary of polarization and electromechanical properties of currently





volume) is also important in determining the ultimate polarization,  $P_u$ , of a polymer,

$$P_u = N\mu \quad (1)$$

Equation (1) is for a rigid dipole model and gives a maximum value for the polarization which assumes all dipoles are perfectly aligned with the poling field. Table 2 lists some amorphous piezoelectric polymers along with ultimate polarizations, remanent polarizations, and calculated and measured dielectric relaxation strengths,  $\Delta\epsilon$ . The dielectric relaxation strength is defined as the change in dielectric constant as the polymer traverses the glass transition temperature. Semicrystalline PVDF is added for comparison.

Table 2. Polarization Data for some Amorphous Piezoelectric Polymers and PVDF.

Polymer	$\mu$ ( $10^{-30}$ Cm)	N ( $10^{28}$ m $^{-3}$ )	$P_u^a$ (mC/m $^2$ )	$\Delta\epsilon^b$ calculated	$\Delta\epsilon^c$ measured	$P_{\text{remnant}}^d$ (mC/m $^2$ ) ( $E_p = 32$ MV/m)	$P_r/P_u$ (%)
PVC	3.7	1.33	50	7.0	10.0	16.0	32
PAN	11.5	1.48	170	35.0	38.0	25.0	17
PVAc	6.0	0.83	50	6.6	6.5	5.0	10
PVDCN/VAc	19.0	0.44	84	30.0	125.0	50.0	60
PPEN	14.0	0.37	52	12.0	12.0	10.0	19
( $\beta$ -CN) APB/ODPA	29.5	0.14	40	23.0	17.6	14.0	35
PVDF	7.0	18.40	130	-	-	40.0-55.0	30-42

<sup>a</sup> Calculated using equation 1.

<sup>b</sup>  $\Delta\epsilon$  is calculated by equation 2.

<sup>c</sup>  $\Delta\epsilon$  is measured as  $(\epsilon_{\text{above } T_g} - \epsilon_{\text{below } T_g})$ .

<sup>d</sup>  $P_r$  is the actual polarization in the polymer, measured by the thermally stimulated current method, or hysteresis measurements in the case of PVDF.

The importance of dipole concentration on ultimate polarization is evident from a comparison of polyacrylonitrile (PAN) and the polyimide ( $\beta$ -CN) APB/ODPA. PAN has a single nitrile dipole per repeat unit ( $\mu = 3.5$  D) resulting in a dipole concentration of  $1.34 \times 10^{28}$  m $^{-3}$ . This translates into an ultimate polarization of 152 mC/m $^2$  [20]. The ( $\beta$ -CN) APB/ODPA polyimide, on the other hand, has a single nitrile dipole pendant to a phenyl ring ( $\mu = 4.2$  D), as well as two anhydride dipoles ( $\mu = 2.34$  D) resulting in a total dipole moment per monomer of 8.8 D [34]. However, the dipole concentration of ( $\beta$ -CN) APB/ODPA is only  $0.136 \times 10^{28}$  m $^{-3}$ , resulting in an ultimate polarization of 40 mC/m $^2$ , which is less than a fourth of that of PAN.

The second criterion for piezoelectricity is the ability to align the dipoles. Orientation polarization of

volume) is also important in determining the ultimate polarization,  $P_u$ , of a polymer,

$$P_u = N\mu \quad (1)$$

Equation (1) is for a rigid dipole model and gives a maximum value for the polarization which assumes all dipoles are perfectly aligned with the poling field. Table 2 lists some amorphous piezoelectric polymers along with ultimate polarizations, remanent polarizations, and calculated and measured dielectric relaxation strengths,  $\Delta\epsilon$ . The dielectric relaxation strength is defined as the change in dielectric constant as the polymer traverses the glass transition temperature. Semicrystalline PVDF is added for comparison.

Table 2. Polarization Data for some Amorphous Piezoelectric Polymers and PVDF.

Polymer	$\mu$ ( $10^{-30}$ Cm)	N ( $10^{28}$ m $^{-3}$ )	$P_u^a$ (mC/m $^2$ )	$\Delta\epsilon^b$ calculated	$\Delta\epsilon^c$ measured	$P_{\text{remnant}}^d$ (mC/m $^2$ ) ( $E_p = 32$ MV/m)	$P_r/P_u$ (%)
PVC	3.7	1.33	50	7.0	10.0	16.0	32
PAN	11.5	1.48	170	35.0	38.0	25.0	17
PVAc	6.0	0.83	50	6.6	6.5	5.0	10
PVDCN/VAc	19.0	0.44	84	30.0	125.0	50.0	60
PPEN	14.0	0.37	52	12.0	12.0	10.0	19
( $\beta$ -CN) APB/ODPA	29.5	0.14	40	23.0	17.6	14.0	35
PVDF	7.0	18.40	130	-	-	40.0-55.0	30-42

<sup>a</sup> Calculated using equation 1.

<sup>b</sup>  $\Delta\epsilon$  is calculated by equation 2.

<sup>c</sup>  $\Delta\epsilon$  is measured as  $(\epsilon_{\text{above } T_g} - \epsilon_{\text{below } T_g})$ .

<sup>d</sup>  $P_r$  is the actual polarization in the polymer, measured by the thermally stimulated current method, or hysteresis measurements in the case of PVDF.

The importance of dipole concentration on ultimate polarization is evident from a comparison of polyacrylonitrile (PAN) and the polyimide ( $\beta$ -CN) APB/ODPA. PAN has a single nitrile dipole per repeat unit ( $\mu = 3.5$  D) resulting in a dipole concentration of  $1.34 \times 10^{28}$  m $^{-3}$ . This translates into an ultimate polarization of 152 mC/m $^2$  [20]. The ( $\beta$ -CN) APB/ODPA polyimide, on the other hand, has a single nitrile dipole pendant to a phenyl ring ( $\mu = 4.2$  D), as well as two anhydride dipoles ( $\mu = 2.34$  D) resulting in a total dipole moment per monomer of 8.8 D [34]. However, the dipole concentration of ( $\beta$ -CN) APB/ODPA is only  $0.136 \times 10^{28}$  m $^{-3}$ , resulting in an ultimate polarization of 40 mC/m $^2$ , which is less than a fourth of that of PAN.

The second criterion for piezoelectricity is the ability to align the dipoles. Orientation polarization of

molecular dipoles is responsible for piezoelectricity in amorphous polymers. It is induced by applying an electric field ( $E_p$ ) at an elevated temperature ( $T_p \geq T_g$ ) where the molecular chains are sufficiently mobile to allow dipole alignment with the electric field. Partial retention of this orientation is achieved by lowering the temperature below  $T_g$  in the presence of  $E_p$  as shown in Figure 1. The resulting remanent polarization ( $P_r$ ) is directly proportional to  $E_p$  and the piezoelectric response. The procedure used to prepare a piezoelectric amorphous polymer clearly results in both oriented dipoles and space or real charge injection. The real charges are usually concentrated near the surface of the polymer as schematically shown in Figure 2, and they are introduced due to the presence of the electrodes. However, Broadhurst et.al. [35] have shown that the presence of space charges does not have a significant effect on the piezoelectric behavior. The reason for this is two fold. The magnitude of the space charges is usually not significant with respect to the polarization charges. Secondly, space charges are essentially symmetrical with respect to the thickness of the polymer therefore when the material is strained uniformly their contribution to the piezoelectric effect is negligible. A number of authors have demonstrated this by use of phenomenological models [33,35].

A study of the relationship between relaxation times, poling temperatures and poling fields is crucial to achieve optimal dipole alignment. Theoretically, the higher the electric field, the better the dipole alignment. However, the value of the electric field is limited by the dielectric breakdown of the polymeric material. In practice, 100 MV/m is the maximum field that can be applied to these materials. Poling times need to be of the order of the relaxation time of the polymer at the poling temperature. It is unlikely that a high degree of alignment is achievable in amorphous polymers as evidenced by  $P_r/P_u$  data in Table II. Using computational chemistry techniques the orientation polarization of the ( $\beta$ -CN) APB/ODPA polymer has been assessed by monitoring the angle,  $\theta$ , that the dipoles make with the applied electric field [34]. The unpoled state is found to exhibit random orientation of the dipoles,  $\theta = 90^\circ$ , as shown in Figure 3a. Upon poling, the nitrile and anhydride dipoles are perturbed by the electric field to form average angles of  $\theta = 50^\circ$  and  $\theta = 63^\circ$ , respectively, Figure 3b. As shown in Table 2, for most polymers the ratio of measured remanent polarization to the calculated ultimate polarization is 30%.

When local ordering or paracrystallinity is inherent in the polymer or is induced by mechanical stretching, an increase in the value of the remanent polarization is observed. For example, some researchers [23, 36, 37] assert that the large discrepancy between the measured and calculated  $\Delta\epsilon$  for PVDCN-VAc (Table 2) may be attributed to locally ordered regions in the polymer. A number of authors have suggested that PVDCN-VAc also exhibits ferroelectric-like behavior [38] due to switching of the nitrile dipoles under AC-field. Several investigators [20, 33, 39] have proposed that the difficulty of poling PAN in the unstretched state is related to the strong dipole-dipole interaction of nitrile groups of the same molecule which repel each other, thus preventing normal polarization. Upon stretching, the intermolecular dipole interactions facilitate the packing of the individual chains and give rise to ordered zones [39]. Comstock et al. [40] measured the remanent polarization of both unstretched and stretched PAN using the thermally stimulated current method (TSC) and observed a two-fold increase in the remanent polarization (TSC peak at  $90^\circ\text{C}$ ) for PAN that was stretched four times its original length.

The third criterion for making an amorphous piezoelectric polymer is the locking-in of dipole alignment and its subsequent stability. As explained earlier, the temperature is lowered to room temperature while the field is still on, to freeze in the dipole alignment. In a semi-crystalline material, however, the locking-in of the polarization is supported by the crystalline structure of the polymer, and is therefore stable above the glass transition temperature of the polymer. It is for that reason that PVDF ( $T_g = -35^\circ\text{C}$ ) can be used from room temperature to about  $100^\circ\text{C}$ . In semi-crystalline materials, piezoelectricity remains until the Curie temperature is reached. Although there is little data addressing the stability of piezoelectric activity in amorphous polymers, the general effect of time, temperature and pressure has been noted. Broadhurst and Davis [33] state that as temperature decreases the structural

molecular dipoles is responsible for piezoelectricity in amorphous polymers. It is induced by applying an electric field ( $E_p$ ) at an elevated temperature ( $T_p \geq T_g$ ) where the molecular chains are sufficiently mobile to allow dipole alignment with the electric field. Partial retention of this orientation is achieved by lowering the temperature below  $T_g$  in the presence of  $E_p$  as shown in Figure 1. The resulting remanent polarization ( $P_r$ ) is directly proportional to  $E_p$  and the piezoelectric response. The procedure used to prepare a piezoelectric amorphous polymer clearly results in both oriented dipoles and space or real charge injection. The real charges are usually concentrated near the surface of the polymer as schematically shown in Figure 2, and they are introduced due to the presence of the electrodes. However, Broadhurst et.al. [35] have shown that the presence of space charges does not have a significant effect on the piezoelectric behavior. The reason for this is two fold. The magnitude of the space charges is usually not significant with respect to the polarization charges. Secondly, space charges are essentially symmetrical with respect to the thickness of the polymer therefore when the material is strained uniformly their contribution to the piezoelectric effect is negligible. A number of authors have demonstrated this by use of phenomenological models [33,35].

A study of the relationship between relaxation times, poling temperatures and poling fields is crucial to achieve optimal dipole alignment. Theoretically, the higher the electric field, the better the dipole alignment. However, the value of the electric field is limited by the dielectric breakdown of the polymeric material. In practice, 100 MV/m is the maximum field that can be applied to these materials. Poling times need to be of the order of the relaxation time of the polymer at the poling temperature. It is unlikely that a high degree of alignment is achievable in amorphous polymers as evidenced by  $P_r/P_u$  data in Table II. Using computational chemistry techniques the orientation polarization of the ( $\beta$ -CN) APB/ODPA polymer has been assessed by monitoring the angle,  $\theta$ , that the dipoles make with the applied electric field [34]. The unpoled state is found to exhibit random orientation of the dipoles,  $\theta = 90^\circ$ , as shown in Figure 3a. Upon poling, the nitrile and anhydride dipoles are perturbed by the electric field to form average angles of  $\theta = 50^\circ$  and  $\theta = 63^\circ$ , respectively, Figure 3b. As shown in Table 2, for most polymers the ratio of measured remanent polarization to the calculated ultimate polarization is 30%.

When local ordering or paracrystallinity is inherent in the polymer or is induced by mechanical stretching, an increase in the value of the remanent polarization is observed. For example, some researchers [23, 36, 37] assert that the large discrepancy between the measured and calculated  $\Delta\epsilon$  for PVDCN-VAc (Table 2) may be attributed to locally ordered regions in the polymer. A number of authors have suggested that PVDCN-VAc also exhibits ferroelectric-like behavior [38] due to switching of the nitrile dipoles under AC-field. Several investigators [20, 33, 39] have proposed that the difficulty of poling PAN in the unstretched state is related to the strong dipole-dipole interaction of nitrile groups of the same molecule which repel each other, thus preventing normal polarization. Upon stretching, the intermolecular dipole interactions facilitate the packing of the individual chains and give rise to ordered zones [39]. Comstock et al. [40] measured the remanent polarization of both unstretched and stretched PAN using the thermally stimulated current method (TSC) and observed a two-fold increase in the remanent polarization (TSC peak at  $90^\circ\text{C}$ ) for PAN that was stretched four times its original length.

The third criterion for making an amorphous piezoelectric polymer is the locking-in of dipole alignment and its subsequent stability. As explained earlier, the temperature is lowered to room temperature while the field is still on, to freeze in the dipole alignment. In a semi-crystalline material, however, the locking-in of the polarization is supported by the crystalline structure of the polymer, and is therefore stable above the glass transition temperature of the polymer. It is for that reason that PVDF ( $T_g = -35^\circ\text{C}$ ) can be used from room temperature to about  $100^\circ\text{C}$ . In semi-crystalline materials, piezoelectricity remains until the Curie temperature is reached. Although there is little data addressing the stability of piezoelectric activity in amorphous polymers, the general effect of time, temperature and pressure has been noted. Broadhurst and Davis [33] state that as temperature decreases the structural



relaxation time of PVC increases rapidly to the order of years at room temperature. This is probably the case for most of the polymers mentioned in this discussion. It has been shown by TSC measurements that the remanent polarization of ( $\beta$ -CN) APB/ODPA is stable when heated at 1°C/min up to 200°C, where over 80% of the  $P_r$  is retained [41]. It is clear that time, pressure and temperature can all contribute to dipole relaxation in these polymers. For a given application and use temperature, the effect of these parameters on the stability of the frozen-in dipole alignment should be determined.

The final determining factor for a material's degree of piezoelectric response is the ability of the polymer to strain with applied stress. Since the remanent polarization in amorphous polymers is lost in the vicinity of  $T_g$ , the use of these piezoelectric polymers is limited to temperatures well below  $T_g$ . This means that the polymers are in their glassy state, and the further away from  $T_g$  the use temperature is, the stiffer the polymer. This also means that measurement of the bulk physical properties is crucial both for identifying practical applications and for comparing polymers. The electromechanical coupling coefficient,  $k_{31}$ , is a measure of the combination of piezoelectric and mechanical properties of a material (refer to Table 3). It can be calculated using the equation below:

$$k_{31} = d_{31} \sqrt{\frac{Y_{11}}{\epsilon \epsilon_0}} \quad (2)$$

where  $d_{31}$  is the piezoelectric strain coefficient,  $Y_{11}$  is Young's modulus,  $\epsilon$  is the dielectric constant of the polymer and  $\epsilon_0$  is the dielectric constant of free space. The piezoelectric amorphous polymer may be used at temperatures near its  $T_g$  to optimize the mechanical properties, but not too close so as not to lose the remanent polarization.

Table 3. Piezoelectric and mechanical properties (at 25°C).

	$E_{11}^a$ ( $10^9$ Dynes/cm <sup>2</sup> )	$d_{31}$ (pC/N)	$k_{31}^f$
PVC	0.34	0.7 <sup>b</sup>	0.001
PAN	1.70	1.7 <sup>c</sup>	0.010
PVDCN-VAc	-	7.0 <sup>d</sup>	0.050
( $\beta$ -CN)APB/ODPA	2.80	0.3 <sup>e</sup>	0.002
PVDF	2.60	27.0 <sup>e</sup>	0.120

<sup>a</sup> Measured in our laboratories using a Rheovibron.

<sup>b</sup> [from Ref. 30].

<sup>c</sup> [from Ref. 21].

<sup>d</sup> [from Ref.27].

<sup>e</sup> Measured in our laboratories.

<sup>f</sup> Calculated using equation 2.

relaxation time of PVC increases rapidly to the order of years at room temperature. This is probably the case for most of the polymers mentioned in this discussion. It has been shown by TSC measurements that the remanent polarization of ( $\beta$ -CN) APB/ODPA is stable when heated at 1°C/min up to 200°C, where over 80% of the  $P_r$  is retained [41]. It is clear that time, pressure and temperature can all contribute to dipole relaxation in these polymers. For a given application and use temperature, the effect of these parameters on the stability of the frozen-in dipole alignment should be determined.

The final determining factor for a material's degree of piezoelectric response is the ability of the polymer to strain with applied stress. Since the remanent polarization in amorphous polymers is lost in the vicinity of  $T_g$ , the use of these piezoelectric polymers is limited to temperatures well below  $T_g$ . This means that the polymers are in their glassy state, and the further away from  $T_g$  the use temperature is, the stiffer the polymer. This also means that measurement of the bulk physical properties is crucial both for identifying practical applications and for comparing polymers. The electromechanical coupling coefficient,  $k_{31}$ , is a measure of the combination of piezoelectric and mechanical properties of a material (refer to Table 3). It can be calculated using the equation below:

$$k_{31} = d_{31} \sqrt{\frac{Y_{11}}{\epsilon \epsilon_0}} \quad (2)$$

where  $d_{31}$  is the piezoelectric strain coefficient,  $Y_{11}$  is Young's modulus,  $\epsilon$  is the dielectric constant of the polymer and  $\epsilon_0$  is the dielectric constant of free space. The piezoelectric amorphous polymer may be used at temperatures near its  $T_g$  to optimize the mechanical properties, but not too close so as not to lose the remanent polarization.

Table 3. Piezoelectric and mechanical properties (at 25°C).

	$E_{11}^a$ ( $10^9$ Dynes/cm <sup>2</sup> )	$d_{31}$ (pC/N)	$k_{31}^f$
PVC	0.34	0.7 <sup>b</sup>	0.001
PAN	1.70	1.7 <sup>c</sup>	0.010
PVDCN-VAc	-	7.0 <sup>d</sup>	0.050
( $\beta$ -CN)APB/ODPA	2.80	0.3 <sup>e</sup>	0.002
PVDF	2.60	27.0 <sup>e</sup>	0.120

<sup>a</sup> Measured in our laboratories using a Rheovibron.

<sup>b</sup> [from Ref. 30].

<sup>c</sup> [from Ref. 21].

<sup>d</sup> [from Ref.27].

<sup>e</sup> Measured in our laboratories.

<sup>f</sup> Calculated using equation 2.

## Dielectric properties as predictors of piezoelectric behavior

This section addresses the origins of the dielectric contribution to the piezoelectric response of amorphous polymers. The potential energy  $U$  of a dipole  $\mu$  at an angle  $\theta$  with the applied electric field is  $U = \mu E \cos \theta$ . Using statistical mechanics and assuming a Boltzmann's distribution of the dipole energies, the mean projection of the dipole moment,  $\langle \mu_E \rangle$ , in the direction of the applied electric field is obtained.

$$\frac{\langle \mu_E \rangle}{\mu} = \coth \frac{\mu E_p}{kT} - \frac{kT}{\mu E} \quad (3)$$

This is the Langevin equation which describes the degree of polarization in a sample when an electric field,  $E$ , is applied at temperature  $T$ . Experimentally, a poling temperature in the vicinity of  $T_g$  is used to maximize dipole motion. The maximum electric field which may be applied, typically 100 MV/m, is determined by the dielectric breakdown strength of the polymer. For amorphous polymers  $\mu E / kT \ll 1$ , which places these systems well within the linear region of the Langevin function. The following linear equation for the remanent polarization results when the Clausius Mossotti equation is used to relate the dielectric constant to the dipole moment [42].

$$P_r = \Delta \epsilon \epsilon_0 E_p \quad (4)$$

It can be concluded that remanent polarization and hence piezoelectric response of a material is determined by  $\Delta \epsilon$ , making it a practical criterion to use when designing piezoelectric amorphous polymers. The dielectric relaxation strength,  $\Delta \epsilon$  may be the result of either free or cooperative dipole motion. Dielectric theory yields a mathematical way of examining the dielectric relaxation due to free rotation of the dipoles,  $\Delta \epsilon$ . The equation incorporates Debye's work based on statistical mechanics, the Clausius Mossotti equation, and the Onsager local field and neglects short range interactions [43]

$$\Delta \epsilon_{\text{calculated}} = \frac{N \mu^2}{3kT \epsilon_0} \left( \frac{n^2 + 2}{3} \right)^2 \left( \frac{3\epsilon(0)}{2\epsilon(0) + n^2} \right)^2 \quad (5)$$

$N$  is the number of dipoles per unit volume,  $k$  is the Boltzmann constant,  $\epsilon(0)$  is the static dielectric constant and  $n$  is the refractive index. If the experimental value of  $\Delta \epsilon$  ( $\Delta \epsilon_{\text{measured}}$ ) agrees with the theoretical value of  $\Delta \epsilon$  ( $\Delta \epsilon_{\text{calculated}}$ ), then the material exhibits free dipolar motion. Table II shows that in polymers such as PAN, VAc, PVC, PPEN, and ( $\beta$ -CN) APB/ODPA the dielectric relaxation strength corresponds to free dipolar motion since  $\Delta \epsilon_{\text{calculated}}$  and  $\Delta \epsilon_{\text{measured}}$  are in agreement. This table also shows that for the copolymer PVDCN/VAc,  $\Delta \epsilon_{\text{calculated}} = 30$  while  $\Delta \epsilon_{\text{measured}} = 125$  [25,26]. This large discrepancy in the values of  $\Delta \epsilon$  is indicative of cooperative motion of several CN dipoles within the locally ordered regions of the polymer. Cooperativity means that instead of each dipole acting independently, multiple CN dipoles respond to the applied electric field in a unified manner. When  $x$  dipoles act cooperatively, the number density of dipoles decreases by  $1/x$  yet the effective dipole moment increases by  $x^2$  to yield a large dielectric relaxation strength. Intramolecular and/or intermolecular interactions between individual dipoles may be responsible for this particular phenomenon [25]. Such interactions are manifested in the existence of paracrystalline regions within the PVDCN/VAc polymer [27]. The large relaxation strength

## Dielectric properties as predictors of piezoelectric behavior

This section addresses the origins of the dielectric contribution to the piezoelectric response of amorphous polymers. The potential energy  $U$  of a dipole  $\mu$  at an angle  $\theta$  with the applied electric field is  $U = \mu E \cos \theta$ . Using statistical mechanics and assuming a Boltzmann's distribution of the dipole energies, the mean projection of the dipole moment,  $\langle \mu_E \rangle$ , in the direction of the applied electric field is obtained.

$$\frac{\langle \mu_E \rangle}{\mu} = \coth \frac{\mu E_p}{kT} - \frac{kT}{\mu E} \quad (3)$$

This is the Langevin equation which describes the degree of polarization in a sample when an electric field,  $E$ , is applied at temperature  $T$ . Experimentally, a poling temperature in the vicinity of  $T_g$  is used to maximize dipole motion. The maximum electric field which may be applied, typically 100 MV/m, is determined by the dielectric breakdown strength of the polymer. For amorphous polymers  $\mu E / kT \ll 1$ , which places these systems well within the linear region of the Langevin function. The following linear equation for the remanent polarization results when the Clausius Mossotti equation is used to relate the dielectric constant to the dipole moment [42].

$$P_r = \Delta \epsilon \epsilon_0 E_p \quad (4)$$

It can be concluded that remanent polarization and hence piezoelectric response of a material is determined by  $\Delta \epsilon$ , making it a practical criterion to use when designing piezoelectric amorphous polymers. The dielectric relaxation strength,  $\Delta \epsilon$  may be the result of either free or cooperative dipole motion. Dielectric theory yields a mathematical way of examining the dielectric relaxation due to free rotation of the dipoles,  $\Delta \epsilon$ . The equation incorporates Debye's work based on statistical mechanics, the Clausius Mossotti equation, and the Onsager local field and neglects short range interactions [43]

$$\Delta \epsilon_{\text{calculated}} = \frac{N \mu^2}{3kT \epsilon_0} \left( \frac{n^2 + 2}{3} \right)^2 \left( \frac{3\epsilon(0)}{2\epsilon(0) + n^2} \right)^2 \quad (5)$$

$N$  is the number of dipoles per unit volume,  $k$  is the Boltzmann constant,  $\epsilon(0)$  is the static dielectric constant and  $n$  is the refractive index. If the experimental value of  $\Delta \epsilon$  ( $\Delta \epsilon_{\text{measured}}$ ) agrees with the theoretical value of  $\Delta \epsilon$  ( $\Delta \epsilon_{\text{calculated}}$ ), then the material exhibits free dipolar motion. Table II shows that in polymers such as PAN, VAc, PVC, PPEN, and ( $\beta$ -CN) APB/ODPA the dielectric relaxation strength corresponds to free dipolar motion since  $\Delta \epsilon_{\text{calculated}}$  and  $\Delta \epsilon_{\text{measured}}$  are in agreement. This table also shows that for the copolymer PVDCN/VAc,  $\Delta \epsilon_{\text{calculated}} = 30$  while  $\Delta \epsilon_{\text{measured}} = 125$  [25,26]. This large discrepancy in the values of  $\Delta \epsilon$  is indicative of cooperative motion of several CN dipoles within the locally ordered regions of the polymer. Cooperativity means that instead of each dipole acting independently, multiple CN dipoles respond to the applied electric field in a unified manner. When  $x$  dipoles act cooperatively, the number density of dipoles decreases by  $1/x$  yet the effective dipole moment increases by  $x^2$  to yield a large dielectric relaxation strength. Intramolecular and/or intermolecular interactions between individual dipoles may be responsible for this particular phenomenon [25]. Such interactions are manifested in the existence of paracrystalline regions within the PVDCN/VAc polymer [27]. The large relaxation strength

exhibited by PVDCN/VAc gives it the largest value of  $P_r$  (Table II) and hence  $d_{31}$  (Table III) of all the amorphous polymers. Although the existence of cooperative dipole motion clearly increases the piezoelectric response of amorphous polymers, the mechanisms by which cooperativity can be systematically incorporated into the polymer structure remain unclear at this time. Finally, Table II demonstrates the efficiency with which polarization may be imparted into the sample. The polymers which exhibit free dipolar motion show relatively low ratios of  $P_r / P_u$  (10-30%) which indicates low dipole orientation, while the ratio of  $P_r / P_u$  for PVDCN/VAc polymer is as high as 60%. It is noted that in Table II,  $P_r$  is measured using the previously mentioned TSC method. As a result,  $P_r$  could include space charge effects as well as dipolar reorientation. However, for all the amorphous polymers of Table II except PVDCN-VAc,  $P_r$  measured was seen to vary linearly with  $E_p$ , which is an indication that space charge effect is negligible since space charge polarization varies nonlinearly with the poling field. Also for these polymers,  $P_r$  measured is of the order of  $P_r$  given by equation (4), again indicating a linear relationship.

Designing an amorphous polymer with a large dielectric relaxation strength and hence piezoelectric response would require the ability to incorporate highly polar groups at high concentrations and cooperative dipole motion.

## Mechanical and electromechanical properties

Mechanical properties are often overlooked when investigating piezoelectric polymers. It is important to note that the piezoelectric response is a result of the coupling between the mechanical and dielectric properties in an amorphous polymer. The piezoelectric coefficient,  $d$ , is defined as

$$d = \frac{1}{A} \left. \frac{\partial Q}{\partial T} \right|_{E=0, T} \quad (6)$$

where  $Q$  is the charge per unit area displaced through a closed circuit between two electrodes,  $T$  is the applied stress,  $A$  is the area of the electrodes, and  $E$  is the applied field. This equation is used for the direct measurement of the  $d_{31}$  piezoelectric coefficient by stressing the polymer in the plane of the film, and measuring the charge that forms on the electrodes under zero field. Figure 4 shows the Young's modulus,  $Y'_{11}$  as a function of temperature. A decrease in the modulus of the four amorphous polymers, PVDCN-VAc, unstretched PAN, PVC, and ( $\beta$ -CN) APB/ODPA, occurs in the region of the glass transition. Figure 5 presents  $d_{31}$  as a function of temperature for several piezoelectric polymers. As the polymers approach their respective glass transition temperatures,  $d_{31}$  increases due to the decrease in the modulus. The trend continues until the thermal energy randomizes the molecular dipoles to yield a decrease in  $d_{31}$  as shown for PAN which has a  $T_g = 90$  °C. The effect of the mechanical properties on the piezoelectric response is also evident by comparing two polymers with comparable remanent polarizations, PVC, and ( $\beta$ -CN) APB/ODPA. The lower modulus (higher compressibility) of PVC results in a larger piezoelectric response relative to ( $\beta$ -CN) APB/ODPA. It is important to note that data for identical processing conditions for the various polymers ( $E_p$ ,  $t_p$  and  $T_p$ ) is not readily available in the literature. This type of data would be very useful for a comparative analysis of the relative effects of mechanical and dielectric properties on the piezoelectric response.

Stretching can also have an effect on the piezoelectric coefficient of a polymer as shown for PAN in Figure 5. The increase in  $d_{31}$  with stretching has both mechanical and polarization contributions. Stretching in the 1-direction aligns the chains in the plane of the film, which results in an increase in the

exhibited by PVDCN/VAc gives it the largest value of  $P_r$  (Table II) and hence  $d_{31}$  (Table III) of all the amorphous polymers. Although the existence of cooperative dipole motion clearly increases the piezoelectric response of amorphous polymers, the mechanisms by which cooperativity can be systematically incorporated into the polymer structure remain unclear at this time. Finally, Table II demonstrates the efficiency with which polarization may be imparted into the sample. The polymers which exhibit free dipolar motion show relatively low ratios of  $P_r / P_u$  (10-30%) which indicates low dipole orientation, while the ratio of  $P_r / P_u$  for PVDCN/VAc polymer is as high as 60%. It is noted that in Table II,  $P_r$  is measured using the previously mentioned TSC method. As a result,  $P_r$  could include space charge effects as well as dipolar reorientation. However, for all the amorphous polymers of Table II except PVDCN-VAc,  $P_r$  measured was seen to vary linearly with  $E_p$ , which is an indication that space charge effect is negligible since space charge polarization varies nonlinearly with the poling field. Also for these polymers,  $P_r$  measured is of the order of  $P_r$  given by equation (4), again indicating a linear relationship.

Designing an amorphous polymer with a large dielectric relaxation strength and hence piezoelectric response would require the ability to incorporate highly polar groups at high concentrations and cooperative dipole motion.

## Mechanical and electromechanical properties

Mechanical properties are often overlooked when investigating piezoelectric polymers. It is important to note that the piezoelectric response is a result of the coupling between the mechanical and dielectric properties in an amorphous polymer. The piezoelectric coefficient,  $d$ , is defined as

$$d = \frac{1}{A} \left. \frac{\partial Q}{\partial T} \right|_{E=0, T} \quad (6)$$

where  $Q$  is the charge per unit area displaced through a closed circuit between two electrodes,  $T$  is the applied stress,  $A$  is the area of the electrodes, and  $E$  is the applied field. This equation is used for the direct measurement of the  $d_{31}$  piezoelectric coefficient by stressing the polymer in the plane of the film, and measuring the charge that forms on the electrodes under zero field. Figure 4 shows the Young's modulus,  $Y'_{11}$  as a function of temperature. A decrease in the modulus of the four amorphous polymers, PVDCN-VAc, unstretched PAN, PVC, and ( $\beta$ -CN) APB/ODPA, occurs in the region of the glass transition. Figure 5 presents  $d_{31}$  as a function of temperature for several piezoelectric polymers. As the polymers approach their respective glass transition temperatures,  $d_{31}$  increases due to the decrease in the modulus. The trend continues until the thermal energy randomizes the molecular dipoles to yield a decrease in  $d_{31}$  as shown for PAN which has a  $T_g = 90$  °C. The effect of the mechanical properties on the piezoelectric response is also evident by comparing two polymers with comparable remanent polarizations, PVC, and ( $\beta$ -CN) APB/ODPA. The lower modulus (higher compressibility) of PVC results in a larger piezoelectric response relative to ( $\beta$ -CN) APB/ODPA. It is important to note that data for identical processing conditions for the various polymers ( $E_p$ ,  $t_p$  and  $T_p$ ) is not readily available in the literature. This type of data would be very useful for a comparative analysis of the relative effects of mechanical and dielectric properties on the piezoelectric response.

Stretching can also have an effect on the piezoelectric coefficient of a polymer as shown for PAN in Figure 5. The increase in  $d_{31}$  with stretching has both mechanical and polarization contributions. Stretching in the 1-direction aligns the chains in the plane of the film, which results in an increase in the

compressibility in the 3-direction. This chain alignment also facilitates dipole orientation in response to an applied electric field [39,40] which results in a higher  $P_r$  than is achievable in unstretched PAN.

Figure 6 stresses the importance of the relative effect of the temperature on the piezoelectric activity. Although from Figure 5, PAN (stretched) looked like it outperforms ( $\beta$ -CN) APB/ODPA, a closer look of Figure 6 shows that at 70°C below  $T_g$ ,  $d_{31}$  of PAN is 1.7 pC/N whereas that of ( $\beta$ -CN) APB/ODPA is 5 pC/N. It is clear that the amorphous piezoelectric polymers have to be used below their  $T_g$  as PAN quickly depolarizes as  $T_g$  is approached. This is in contrast to PVDF (and other semicrystalline polymers) which are used well above their  $T_g$ 's.

The coupling between mechanical and dielectric properties is also evident in the hydrostatic piezoelectric coefficient,  $d_h$  which is given by [33]:

$$d_h = -\beta \Delta\epsilon \epsilon_0 \epsilon_\infty E_p / 3 \quad (7)$$

As seen in equation 7, both the mechanical properties (through the compressibility,  $\beta$ ) and the dielectric properties (represented by  $\Delta\epsilon$ ) affect the piezoelectric coefficient. Figure 7a presents the mechanical and electrical properties as a function of temperature for ( $\beta$ -CN) APB/ODPA. The compressibility of ( $\beta$ -CN) APB/ODPA increases slightly with temperature until  $T_g$  is reached. The remanent polarization is relatively stable until about 50 degrees below the glass transition at which point it decreases due to dipole randomization. Consequently as shown in Figure 7b,  $d_h$  increases slightly with temperature prior to the onset of depolarization.

## Summary

This review has brought together the dielectric theory and the mechanical properties which define the piezoelectric response in amorphous polymers. The basic requirements for designing an amorphous piezoelectric polymer are the presence and concentration of dipoles, the ability to orient these dipoles and to lock them in this alignment, and the ability to sufficiently strain the polymer. Calculating the ultimate polarization  $P_u$  is a good starting point when designing piezoelectric amorphous polymers. A primary weakness of amorphous polymers is poor dipole alignment during poling (low  $P_r/P_u$  value). This would be overcome by incorporating cooperativity such as the case of PVDCN-VAc. Until this phenomenon is understood where the polymer structure may be engineered to include cooperativity, incremental steps are being made to improve the response by incorporating large dipoles (primarily CN) at high concentrations.

## References

1. Kawai, H. *Jpn, J. Appl. Phys.* **1969**, 8, p. 975
2. Lovinger, A.J. In *Developments in Crystalline Polymers*; Basset, D.C., Ed.; Applied Science Publishers: London, UK, 1982, Vol. 1; p. 242
3. Carbeck, J.D.; Lacks, D.J.; Rutledge, G.C. *J. Chem. Phys.* **1995**, 103, p. 10347
4. Kepler, R.G.; Anderson, R.A. *J. Appl. Phys.* **1978**, 49, p. 1232
5. Davis, G.T.; McKinney, J.E.; Broadhurst, M.G.; Roth, S.C. *J. Appl. Phys.* **1978**, 49, p. 4998
6. Lovinger, A.J. *Science* **1983**, 220, p. 1115
7. Davis, G.T. In *Polymers for Electronic and Photonic Applications*, Wong, C. P., Ed., Academic Press, Inc.: Boston, MA, 1993; p. 435

compressibility in the 3-direction. This chain alignment also facilitates dipole orientation in response to an applied electric field [39,40] which results in a higher  $P_r$  than is achievable in unstretched PAN.

Figure 6 stresses the importance of the relative effect of the temperature on the piezoelectric activity. Although from Figure 5, PAN (stretched) looked like it outperforms ( $\beta$ -CN) APB/ODPA, a closer look of Figure 6 shows that at 70°C below  $T_g$ ,  $d_{31}$  of PAN is 1.7 pC/N whereas that of ( $\beta$ -CN) APB/ODPA is 5 pC/N. It is clear that the amorphous piezoelectric polymers have to be used below their  $T_g$  as PAN quickly depolarizes as  $T_g$  is approached. This is in contrast to PVDF (and other semicrystalline polymers) which are used well above their  $T_g$ 's.

The coupling between mechanical and dielectric properties is also evident in the hydrostatic piezoelectric coefficient,  $d_h$  which is given by [33]:

$$d_h = -\beta \Delta\epsilon \epsilon_0 \epsilon_\infty E_p / 3 \quad (7)$$

As seen in equation 7, both the mechanical properties (through the compressibility,  $\beta$ ) and the dielectric properties (represented by  $\Delta\epsilon$ ) affect the piezoelectric coefficient. Figure 7a presents the mechanical and electrical properties as a function of temperature for ( $\beta$ -CN) APB/ODPA. The compressibility of ( $\beta$ -CN) APB/ODPA increases slightly with temperature until  $T_g$  is reached. The remanent polarization is relatively stable until about 50 degrees below the glass transition at which point it decreases due to dipole randomization. Consequently as shown in Figure 7b,  $d_h$  increases slightly with temperature prior to the onset of depolarization.

## Summary

This review has brought together the dielectric theory and the mechanical properties which define the piezoelectric response in amorphous polymers. The basic requirements for designing an amorphous piezoelectric polymer are the presence and concentration of dipoles, the ability to orient these dipoles and to lock them in this alignment, and the ability to sufficiently strain the polymer. Calculating the ultimate polarization  $P_u$  is a good starting point when designing piezoelectric amorphous polymers. A primary weakness of amorphous polymers is poor dipole alignment during poling (low  $P_r/P_u$  value). This would be overcome by incorporating cooperativity such as the case of PVDCN-VAc. Until this phenomenon is understood where the polymer structure may be engineered to include cooperativity, incremental steps are being made to improve the response by incorporating large dipoles (primarily CN) at high concentrations.

## References

1. Kawai, H. *Jpn, J. Appl. Phys.* **1969**, 8, p. 975
2. Lovinger, A.J. In *Developments in Crystalline Polymers*; Basset, D.C., Ed.; Applied Science Publishers: London, UK, 1982, Vol. 1; p. 242
3. Carbeck, J.D.; Lacks, D.J.; Rutledge, G.C. *J. Chem. Phys.* **1995**, 103, p. 10347
4. Kepler, R.G.; Anderson, R.A. *J. Appl. Phys.* **1978**, 49, p. 1232
5. Davis, G.T.; McKinney, J.E.; Broadhurst, M.G.; Roth, S.C. *J. Appl. Phys.* **1978**, 49, p. 4998
6. Lovinger, A.J. *Science* **1983**, 220, p. 1115
7. Davis, G.T. In *Polymers for Electronic and Photonic Applications*, Wong, C. P., Ed., Academic Press, Inc.: Boston, MA, 1993; p. 435



8. Sessler, G.M. *J. Acoust. Soc. Am.* **1981**, 70, p. 1596
9. Kepler, R. G.; Anderson, R. A. *Adv. Phys.* **1992**, 41, p. 1
10. Wada, Y.; Hayakawa, R. *Jpn. J. Appl. Phys.* **1976**, 15, p. 2041
11. Furukawa, T.; Wang, T.T. In *The Applications of Ferroelectric Polymers*, Wang, T.; Herbert, J.; Glass, A., Ed.; Blackie: London, U.K., 1988; p. 66
12. Garner, G. M. In *The Applications of Ferroelectric Polymers*, Wang, T.; Herbert, J.; Glass, A., Ed.; Blackie: London, U.K., 1988; p. 190
13. Meeker, T. R. In *The Applications of Ferroelectric Polymers*, Wang, T.; Herbert, J.; Glass, A., Ed.; Blackie: London, U.K., 1988; p. 305
14. Yamaka, E. In *The Applications of Ferroelectric Polymers*, Wang, T.; Herbert, J.; Glass, A., Ed.; Blackie: London, U.K., 1988; p. 329
15. Newman, B.A., Chen, P., Pae, K.D., and Scheinbeim, J.I. *J. Appl. Phys.* **1980**, 51, p. 5161
16. Scheinbeim, J.I. *J. Appl. Phys.* **1981**, 52, p. 5939
17. Mathur, S.C.; Scheinbeim, J.I.; Newman, B.A. *J. Appl. Phys.* **1984**, 56, p. 2419
18. Nalwa, H.S.; Taneja, K. L.; Tewari, U.; Vasudevan, P. *Proc. Nucl. Phys. Sol. St. Phys.Symp.* **1978**, 21C, p. 712
19. Fukada, E.; Tasaka, S.; Nalwa, H. S. In *Ferroelectric Polymers, Chemistry, Physics and Application*, Nalwa, H. S., Ed.; Plastics Engineering, 28; Marcel Dekker, Inc.:New York, NY, 1995, p. 353
20. Ueda, H.; Carr, S.H. *Polymer J.* **1984**, 16, p. 661
21. von Berlepsch, H.; Pinnow, M.; Stark, W. *J. Phys. D: Appl. Phys.* **1989**, 22, p. 1143
22. von Berlepsch., H.; Kunstler, W.; Wedel, A.; Danz, R.; Geiss, D. *IEEE Trans. Elec. Ins.* **1989**, 24, p. 357
23. Jo, Y. S.; Sakurai, M.; Inoue, Y.; Chujo, R.; Tasaka, S.; Miyata, S. *Polym.* **1987**, 28, p. 1583
24. Miyata, S.; Yoshikawa, M.; Tasaka, S.; Ko, M. *Polymer J.* **1980**, 12, p. 857
25. Furukawa, T.; Tada, M.; Nakajima, K.; Seo, I. *Jpn. J. Appl. Phys.* **1988**, 27, p. 200
26. Sakurai, M.; Ohta, Y.; Inoue, Y.; Chujo, R. *Polym. Comm.* **1991**, 32, p. 397
27. Tasaka, S.; Inagaki, N.; Okutani, T.; Miyata, S. *Polymer* **1989**, 30, p. 1639
28. Tasaka, S.; Toyama, T.; Inagaki, N. *Jpn. J. Appl. Phys.* **1994**, 33, p. 5838
29. Hall, H.K.; Chan, R.; Oku, J.; Huges, O.R.; Scheinbeim, J.; Newman, B. *Polym. Bulletin* **1987**, 17, p. 135
30. Broadhurst, M. G.; Malmberg, C. G.; F. I. Mopsik; Harris, W. P., In *Electrets, Charge Storage and Transport in Dielectrics*; Perlman, M. M., Ed.; The Electrochemical Society: New York, NY; p. 492
31. Mopsik, F. I.; Broadhurst, M. G. *J. Appl. Phys.* **1975**, 46, p. 4204
32. Ounaies, Z.; Young, J. A.; Simpson, J.O.; Farmer, B. L. In *Materials Research Society Proceedings: Materials for Smart Systems II*, George, E., P.; Gotthardt, R.; Otsuka, K.; Trolier-McKinstry, S.; Wun-Fogle, M., Ed.; Materials Research Society: Pittsburgh, PA, 1997, Vol. 459; p. 59
33. Broadhurst, M. G. and Davis, G. T., In *Electrets*; Sessler, G. M., Ed.; Springer-Verlag: New York, NY, 1980; Vol 33; p. 285
34. Young, J. A., Farmer, B. L. and Hinkley, J. A., *Polymer*, vol. 40, 10, 1999, p2787
35. Broadhurst, M. G.; Harris, W. P.; Mopsik, F. I.; Malmberg, C. G. *Polym. Prep.* **1973**, 14, p. 820
36. Furukawa, T. *IEEE Trans. Elect. Insul.* **1989**, 21, p. 375

8. Sessler, G.M. *J. Acoust. Soc. Am.* **1981**, 70, p. 1596
9. Kepler, R. G.; Anderson, R. A. *Adv. Phys.* **1992**, 41, p. 1
10. Wada, Y.; Hayakawa, R. *Jpn. J. Appl. Phys.* **1976**, 15, p. 2041
11. Furukawa, T.; Wang, T.T. In *The Applications of Ferroelectric Polymers*, Wang, T.; Herbert, J.; Glass, A., Ed.; Blackie: London, U.K., 1988; p. 66
12. Garner, G. M. In *The Applications of Ferroelectric Polymers*, Wang, T.; Herbert, J.; Glass, A., Ed.; Blackie: London, U.K., 1988; p. 190
13. Meeker, T. R. In *The Applications of Ferroelectric Polymers*, Wang, T.; Herbert, J.; Glass, A., Ed.; Blackie: London, U.K., 1988; p. 305
14. Yamaka, E. In *The Applications of Ferroelectric Polymers*, Wang, T.; Herbert, J.; Glass, A., Ed.; Blackie: London, U.K., 1988; p. 329
15. Newman, B.A., Chen, P., Pae, K.D., and Scheinbeim, J.I. *J. Appl. Phys.* **1980**, 51, p. 5161
16. Scheinbeim, J.I. *J. Appl. Phys.* **1981**, 52, p. 5939
17. Mathur, S.C.; Scheinbeim, J.I.; Newman, B.A. *J. Appl. Phys.* **1984**, 56, p. 2419
18. Nalwa, H.S.; Taneja, K. L.; Tewari, U.; Vasudevan, P. *Proc. Nucl. Phys. Sol. St. Phys.Symp.* **1978**, 21C, p. 712
19. Fukada, E.; Tasaka, S.; Nalwa, H. S. In *Ferroelectric Polymers, Chemistry, Physics and Application*, Nalwa, H. S., Ed.; Plastics Engineering, 28; Marcel Dekker, Inc.:New York, NY, 1995, p. 353
20. Ueda, H.; Carr, S.H. *Polymer J.* **1984**, 16, p. 661
21. von Berlepsch, H.; Pinnow, M.; Stark, W. *J. Phys. D: Appl. Phys.* **1989**, 22, p. 1143
22. von Berlepsch., H.; Kunstler, W.; Wedel, A.; Danz, R.; Geiss, D. *IEEE Trans. Elec. Ins.* **1989**, 24, p. 357
23. Jo, Y. S.; Sakurai, M.; Inoue, Y.; Chujo, R.; Tasaka, S.; Miyata, S. *Polym.* **1987**, 28, p. 1583
24. Miyata, S.; Yoshikawa, M.; Tasaka, S.; Ko, M. *Polymer J.* **1980**, 12, p. 857
25. Furukawa, T.; Tada, M.; Nakajima, K.; Seo, I. *Jpn. J. Appl. Phys.* **1988**, 27, p. 200
26. Sakurai, M.; Ohta, Y.; Inoue, Y.; Chujo, R. *Polym. Comm.* **1991**, 32, p. 397
27. Tasaka, S.; Inagaki, N.; Okutani, T.; Miyata, S. *Polymer* **1989**, 30, p. 1639
28. Tasaka, S.; Toyama, T.; Inagaki, N. *Jpn. J. Appl. Phys.* **1994**, 33, p. 5838
29. Hall, H.K.; Chan, R.; Oku, J.; Huges, O.R.; Scheinbeim, J.; Newman, B. *Polym. Bulletin* **1987**, 17, p. 135
30. Broadhurst, M. G.; Malmberg, C. G.; F. I. Mopsik; Harris, W. P., In *Electrets, Charge Storage and Transport in Dielectrics*; Perlman, M. M., Ed.; The Electrochemical Society: New York, NY; p. 492
31. Mopsik, F. I.; Broadhurst, M. G. *J. Appl. Phys.* **1975**, 46, p. 4204
32. Ounaies, Z.; Young, J. A.; Simpson, J.O.; Farmer, B. L. In *Materials Research Society Proceedings: Materials for Smart Systems II*, George, E., P.; Gotthardt, R.; Otsuka, K.; Trolier-McKinstry, S.; Wun-Fogle, M., Ed.; Materials Research Society: Pittsburgh, PA, 1997, Vol. 459; p. 59
33. Broadhurst, M. G. and Davis, G. T., In *Electrets*; Sessler, G. M., Ed.; Springer-Verlag: New York, NY, 1980; Vol 33; p. 285
34. Young, J. A., Farmer, B. L. and Hinkley, J. A., *Polymer*, vol. 40, 10, 1999, p2787
35. Broadhurst, M. G.; Harris, W. P.; Mopsik, F. I.; Malmberg, C. G. *Polym. Prep.* **1973**, 14, p. 820
36. Furukawa, T. *IEEE Trans. Elect. Insul.* **1989**, 21, p. 375

37. Furukawa, T.; Date, M.; Nakajima, K.; Kosaka, T.; Seo, I. *Jpn. J. Appl. Phys.* **1986**, *25*, p. 1178
38. Wang, T. T.; Takase, Y., *J. Appl. Phys.* **1987**, *62*, p. 3466
39. Tasaka, S. In *Ferroelectric Polymers*; Nalwa, H. S., Ed.; *Plastics Engineering 28*; Marcel Dekker Inc.: New York, NY, 1995; p. 325
40. Comstock, R. J.; Stupp, S. I.; Carr, S. H. *J. Macrom. Sci. Phys.* **1977**, *B13*, p. 101
41. Simpson, J.O., Ounaies, Z., and Fay, C., In *Materials Research Society Proceedings: Materials for Smart Systems II*, George, E., P., Gotthardt, R., Otsuka, K., Trolier-McKinstry, S., and Wun-Fogle, M., Ed., *Materials Research Society*: Pittsburgh, PA, 1997, Vol. 459; p. 53
42. Hilczer, B.; Malecki, J. In *Electrets*; *Studies in Electrical and Electronic Engineering 14*; Elsevier: New York, NY, 1986; p. 19
43. Frohlick, H. In *Theory of Dielectrics*; *Monographs on the Physics and Chemistry of Materials 42*; Oxford University Press: Oxford, U. K., 1958; p. 15

37. Furukawa, T.; Date, M.; Nakajima, K.; Kosaka, T.; Seo, I. *Jpn. J. Appl. Phys.* **1986**, *25*, p. 1178
38. Wang, T. T.; Takase, Y., *J. Appl. Phys.* **1987**, *62*, p. 3466
39. Tasaka, S. In *Ferroelectric Polymers*; Nalwa, H. S., Ed.; *Plastics Engineering 28*; Marcel Dekker Inc.: New York, NY, 1995; p. 325
40. Comstock, R. J.; Stupp, S. I.; Carr, S. H. *J. Macrom. Sci. Phys.* **1977**, *B13*, p. 101
41. Simpson, J.O., Ounaies, Z., and Fay, C., In *Materials Research Society Proceedings: Materials for Smart Systems II*, George, E., P., Gotthardt, R., Otsuka, K., Troler-McKinstry, S., and Wun-Fogle, M., Ed., *Materials Research Society*: Pittsburgh, PA, 1997, Vol. 459; p. 53
42. Hilczer, B.; Malecki, J. In *Electrets*; *Studies in Electrical and Electronic Engineering 14*; Elsevier: New York, NY, 1986; p. 19
43. Frohlick, H. In *Theory of Dielectrics*; *Monographs on the Physics and Chemistry of Materials 42*; Oxford University Press: Oxford, U. K., 1958; p. 15

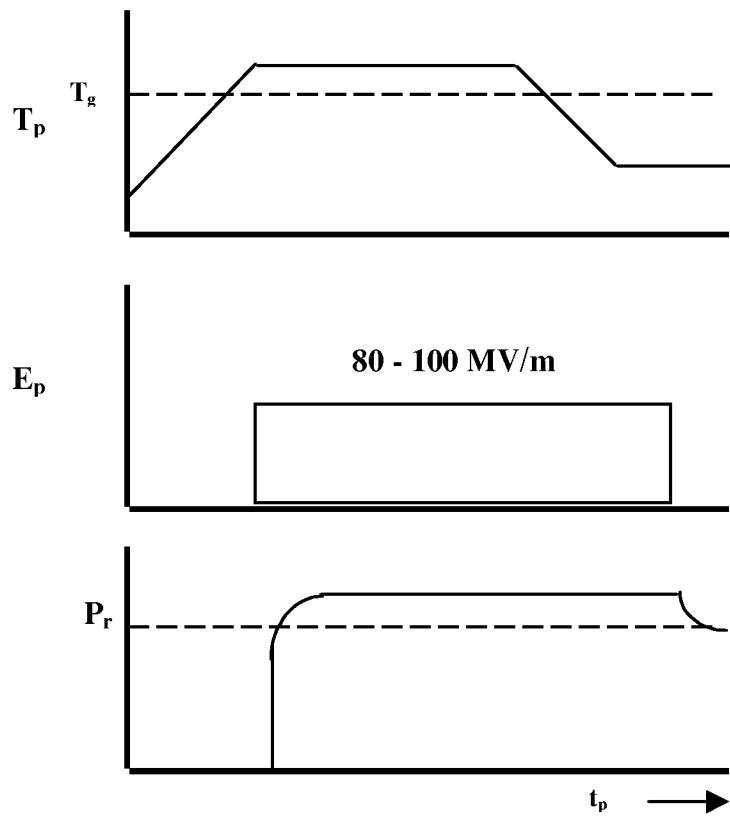


Figure 1. Poling of amorphous polymers.

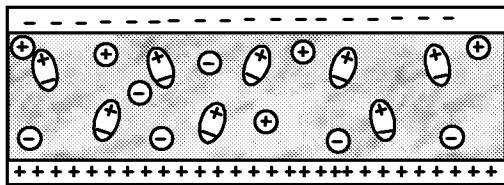


Figure 2. Real charges in an amorphous electret.

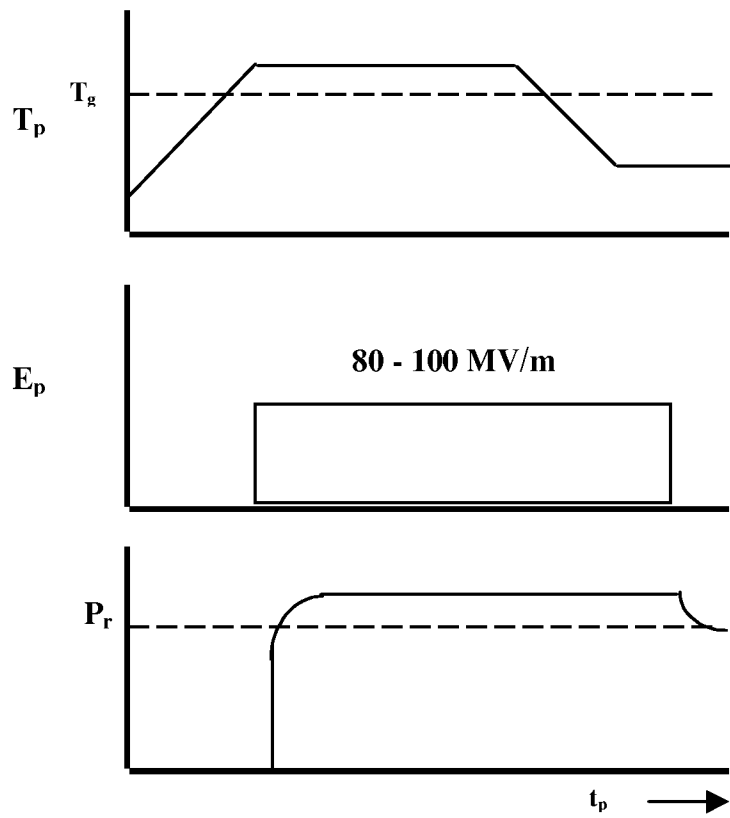


Figure 1. Poling of amorphous polymers.

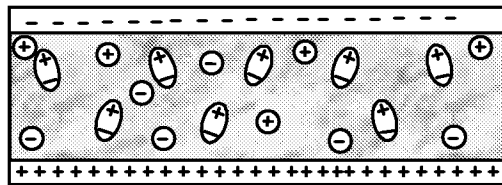


Figure 2. Real charges in an amorphous electret.

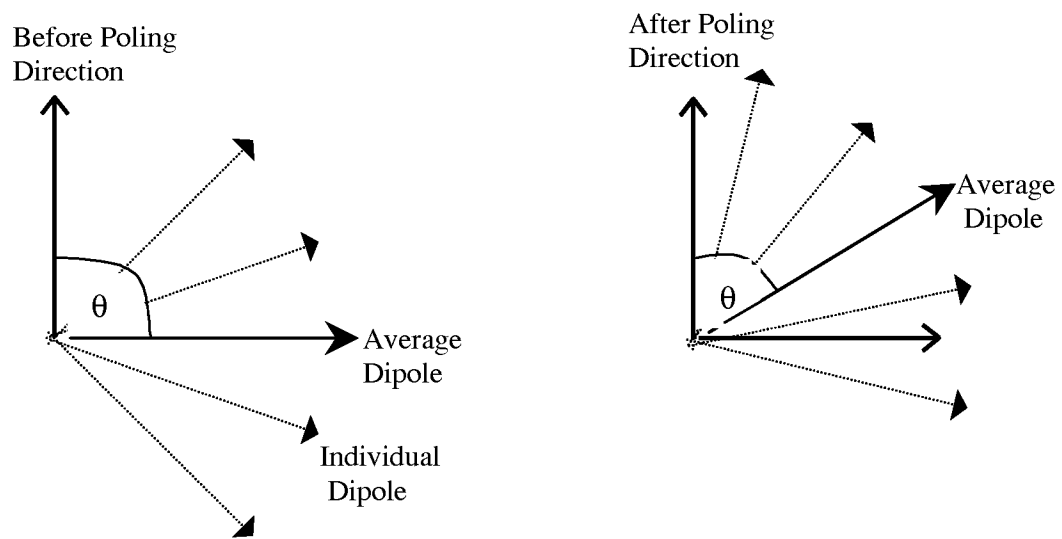


Figure 3. Orientation polarization of the ( $\beta$ -CN) APB/ODPA polymer.

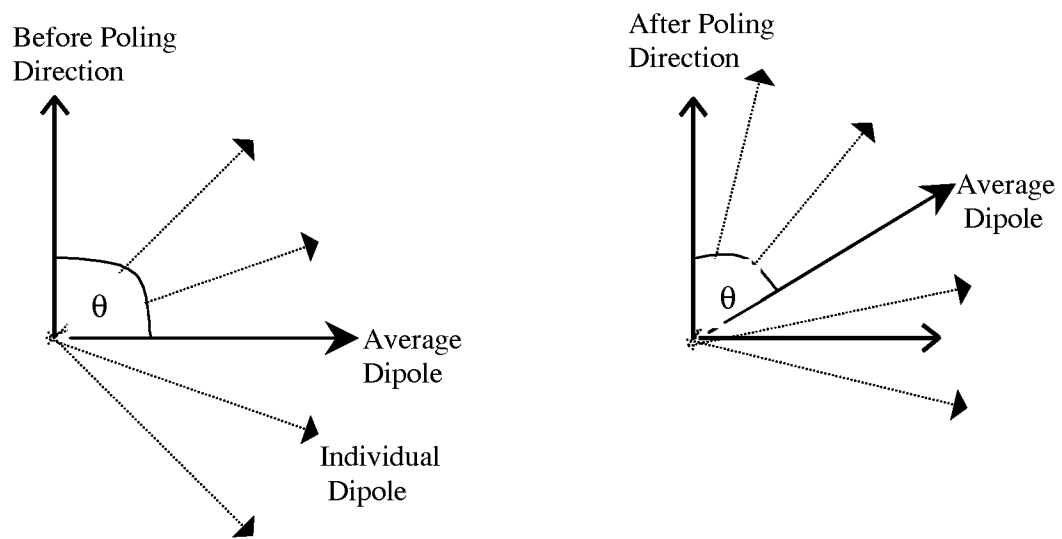


Figure 3. Orientation polarization of the ( $\beta$ -CN) APB/ODPA polymer.



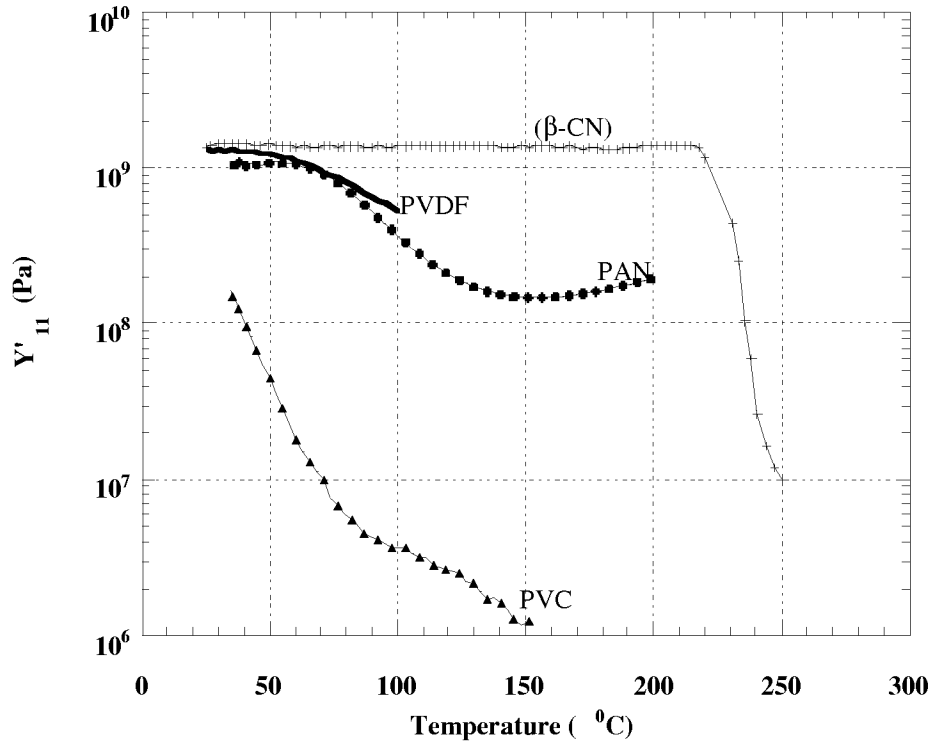


Figure 4. Modulus as a function of temperature.

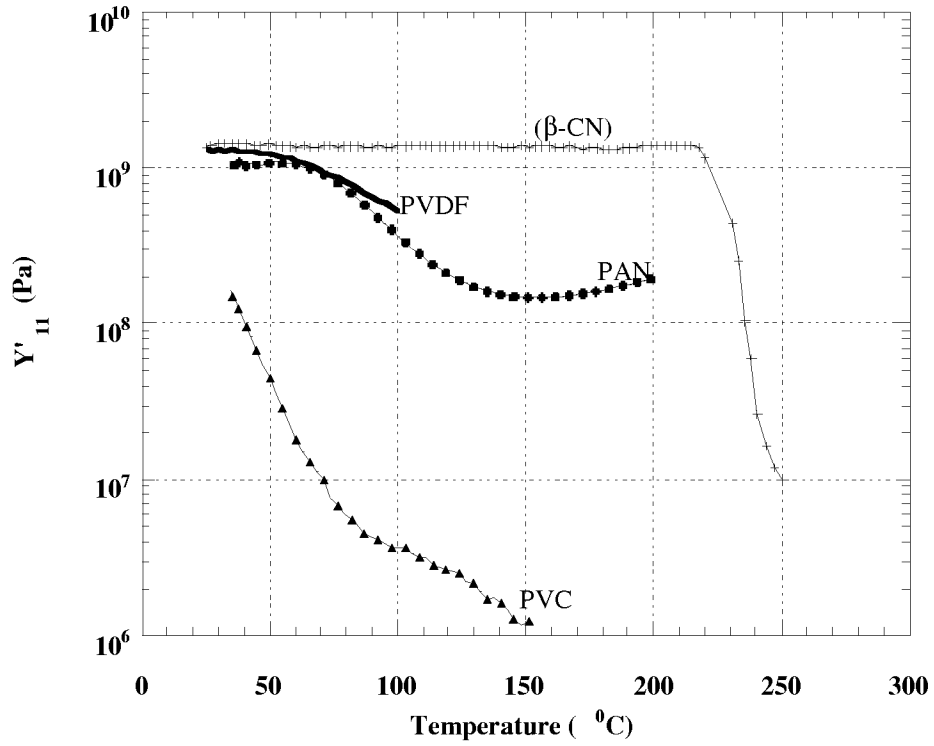


Figure 4. Modulus as a function of temperature.

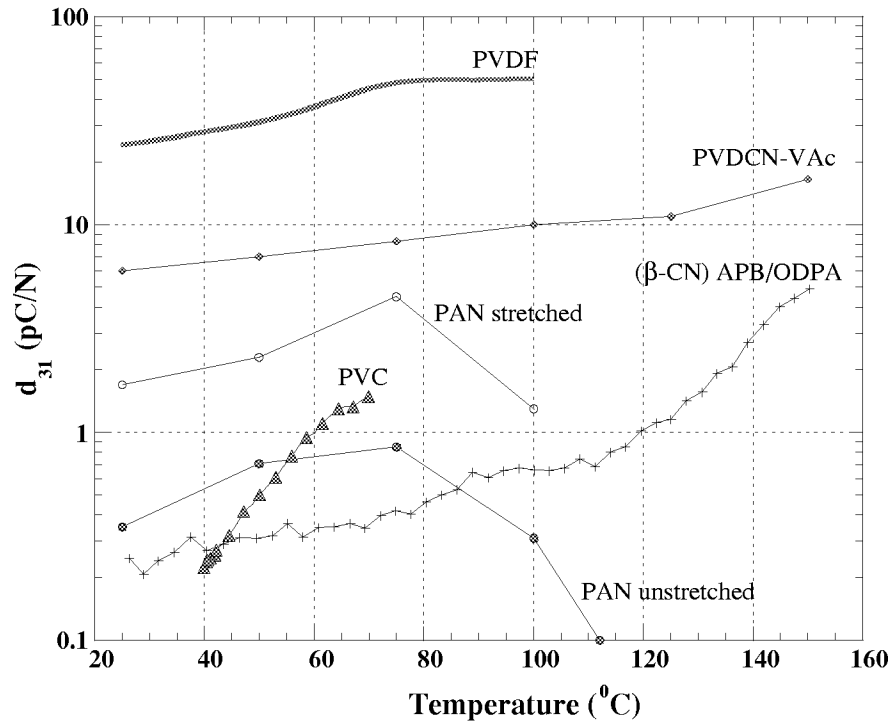


Figure 5. Comparison of the piezoelectric strain coefficient,  $d_{31}$ .

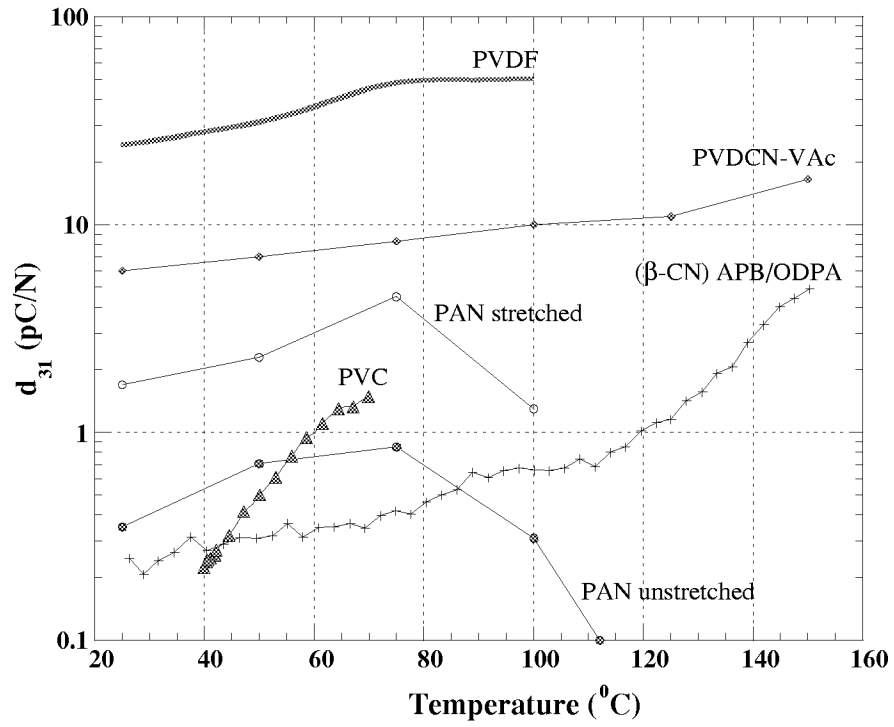


Figure 5. Comparison of the piezoelectric strain coefficient,  $d_{31}$ .

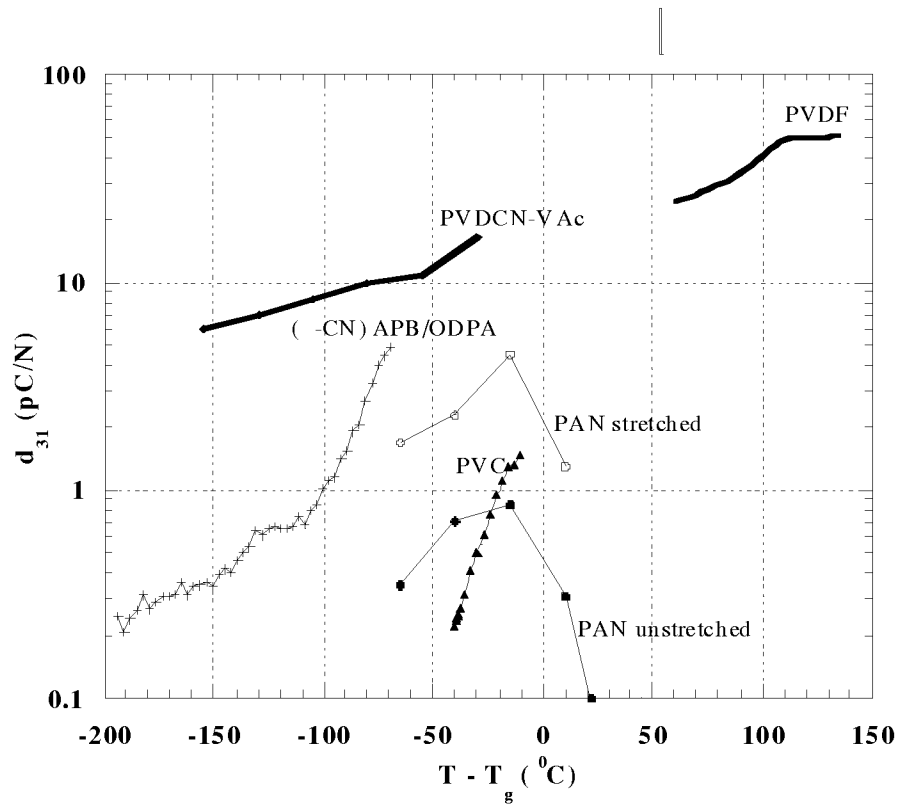


Figure 6. Piezoelectric strain coefficient  $d_{31}$  with respect to  $T_g$ .

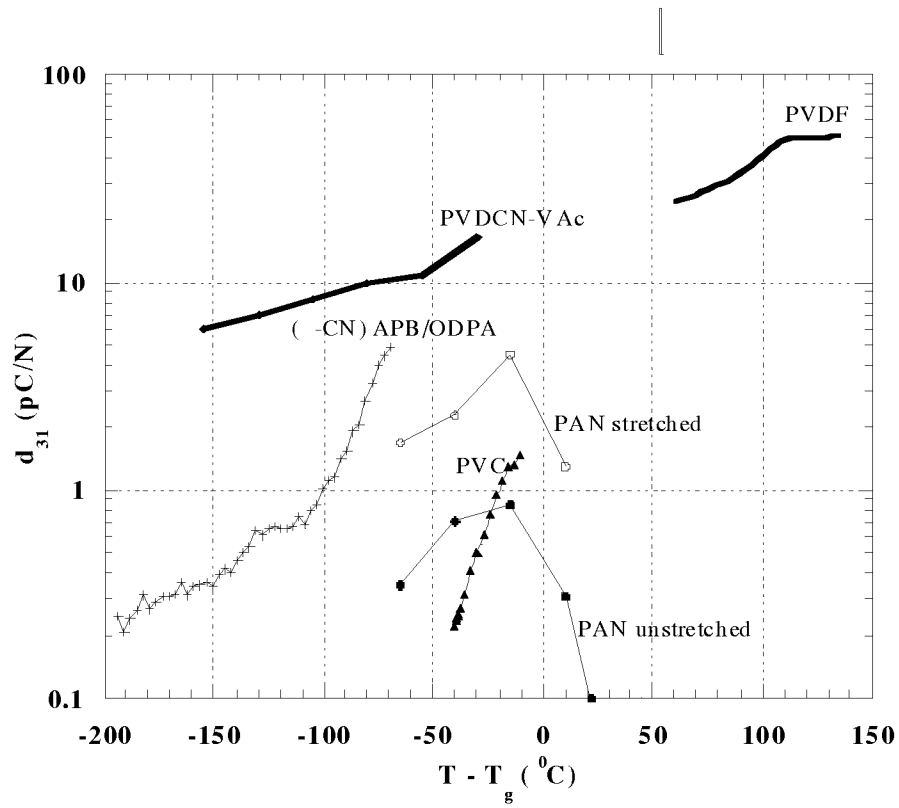


Figure 6. Piezoelectric strain coefficient  $d_{31}$  with respect to  $T_g$ .

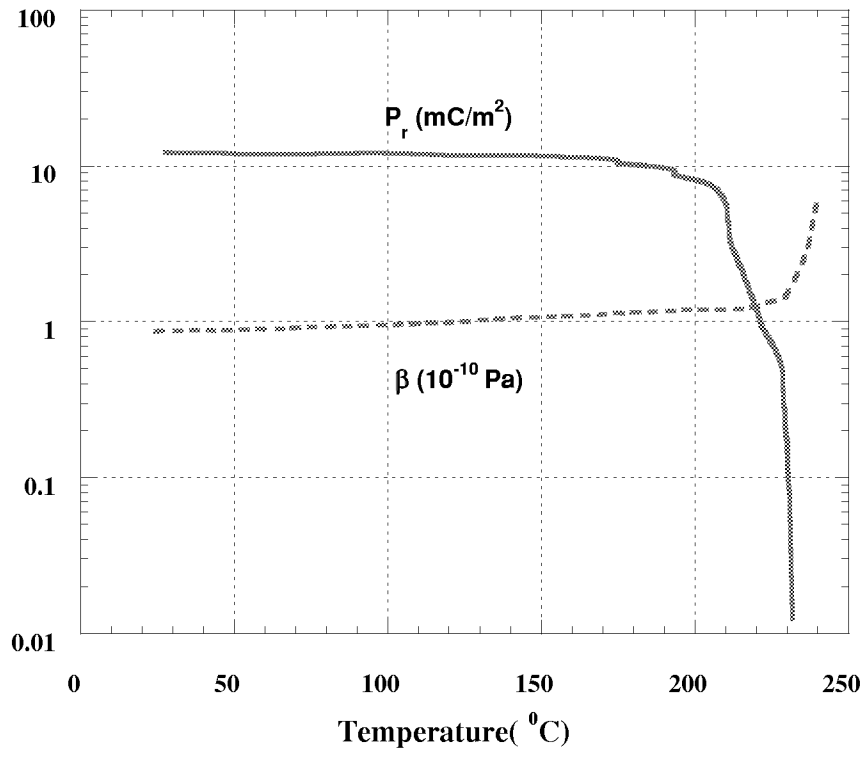


Figure 7a. Mechanical and electrical properties of ( $\beta$ -CN) APB/ODPA as a function of temperature.

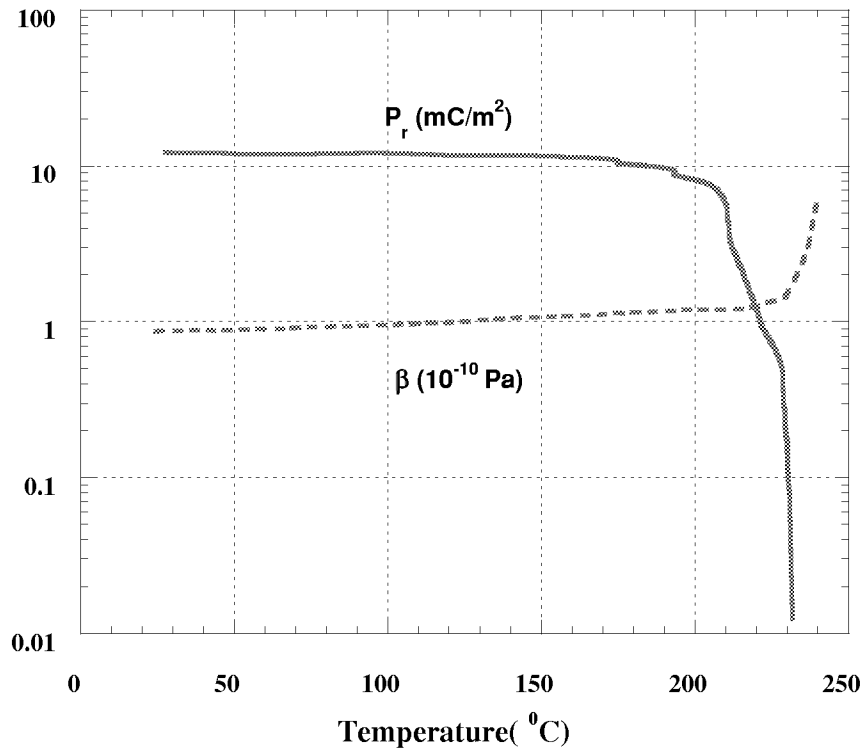


Figure 7a. Mechanical and electrical properties of ( $\beta$ -CN) APB/ODPA as a function of temperature.



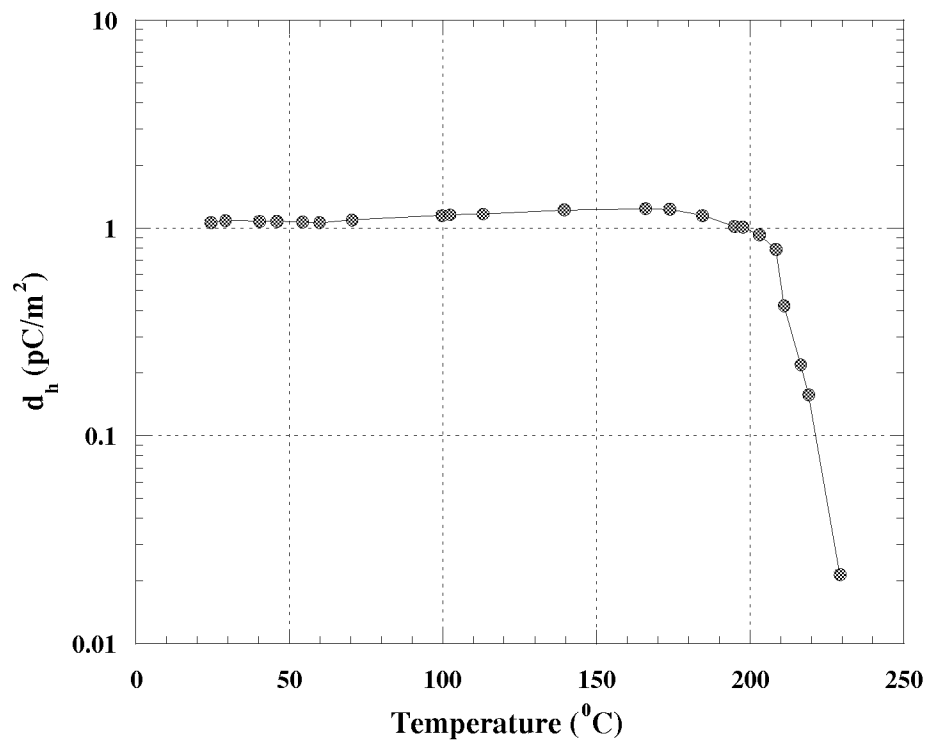


Figure 7b. Hydrostatic piezoelectric coefficient,  $d_h$ .

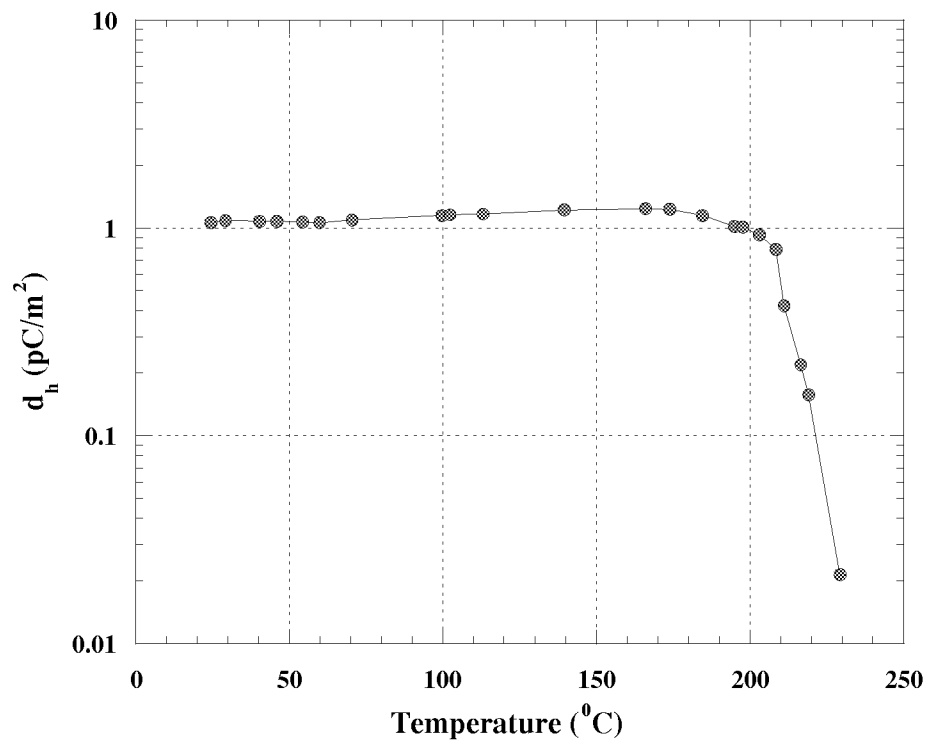


Figure 7b. Hydrostatic piezoelectric coefficient,  $d_h$ .

REPORT DOCUMENTATION PAGE			Form Approved OMB No. 0704-0188	
Public reporting burden for this collection of information is estimated to average 1 hour per response, including the time for reviewing instructions, searching existing data sources, gathering and maintaining the data needed, and completing and reviewing the collection of information. Send comments regarding this burden estimate or any other aspect of this collection of information, including suggestions for reducing this burden, to Washington Headquarters Services, Directorate for Information Operations and Reports, 1215 Jefferson Davis Highway, Suite 1204, Arlington, VA 22202-4302, and to the Office of Management and Budget, Paperwork Reduction Project (0704-0188), Washington, DC 20503.				
1. AGENCY USE ONLY (Leave blank)		2. REPORT DATE August 1999	3. REPORT TYPE AND DATES COVERED Technical Memorandum	
4. TITLE AND SUBTITLE Design Requirements for Amorphous Piezoelectric Polymers			5. FUNDING NUMBERS WU 522-32-00-02	
6. AUTHOR(S) Z. Ounaies, J. A. Young and J. S. Harrison				
7. PERFORMING ORGANIZATION NAME(S) AND ADDRESS(ES) NASA Langley Research Center Hampton, VA 23681-2199			8. PERFORMING ORGANIZATION REPORT NUMBER L-17883	
9. SPONSORING/MONITORING AGENCY NAME(S) AND ADDRESS(ES) National Aeronautics and Space Administration Washington, DC 20546-0001			10. SPONSORING/MONITORING AGENCY REPORT NUMBER NASA/TM-1999-209359	
11. SUPPLEMENTARY NOTES				
12a. DISTRIBUTION/AVAILABILITY STATEMENT Unclassified-Unlimited Subject Category 27 Distribution: Standard Availability: NASA CASI (301) 621-0390			12b. DISTRIBUTION CODE	
13. ABSTRACT (Maximum 200 words)  An overview of the piezoelectric activity in amorphous piezoelectric polymers is presented. The criteria required to render a polymer piezoelectric are discussed. Although piezoelectricity is a coupling between mechanical and electrical properties, most research has concentrated on the electrical properties of potentially piezoelectric polymers. In this work, we present comparative mechanical data as a function of temperature and offer a summary of polarization and electromechanical properties for each of the polymers considered.				
14. SUBJECT TERMS Piezoelectric, Amorphous polymers, Polyimides			15. NUMBER OF PAGES 22	
			16. PRICE CODE A03	
17. SECURITY CLASSIFICATION OF REPORT Unclassified	18. SECURITY CLASSIFICATION OF THIS PAGE Unclassified	19. SECURITY CLASSIFICATION OF ABSTRACT Unclassified	20. LIMITATION OF ABSTRACT UL	

REPORT DOCUMENTATION PAGE			Form Approved OMB No. 0704-0188	
Public reporting burden for this collection of information is estimated to average 1 hour per response, including the time for reviewing instructions, searching existing data sources, gathering and maintaining the data needed, and completing and reviewing the collection of information. Send comments regarding this burden estimate or any other aspect of this collection of information, including suggestions for reducing this burden, to Washington Headquarters Services, Directorate for Information Operations and Reports, 1215 Jefferson Davis Highway, Suite 1204, Arlington, VA 22202-4302, and to the Office of Management and Budget, Paperwork Reduction Project (0704-0188), Washington, DC 20503.				
1. AGENCY USE ONLY (Leave blank)		2. REPORT DATE August 1999	3. REPORT TYPE AND DATES COVERED Technical Memorandum	
4. TITLE AND SUBTITLE Design Requirements for Amorphous Piezoelectric Polymers			5. FUNDING NUMBERS WU 522-32-00-02	
6. AUTHOR(S) Z. Ounaies, J. A. Young and J. S. Harrison				
7. PERFORMING ORGANIZATION NAME(S) AND ADDRESS(ES) NASA Langley Research Center Hampton, VA 23681-2199			8. PERFORMING ORGANIZATION REPORT NUMBER L-17883	
9. SPONSORING/MONITORING AGENCY NAME(S) AND ADDRESS(ES) National Aeronautics and Space Administration Washington, DC 20546-0001			10. SPONSORING/MONITORING AGENCY REPORT NUMBER NASA/TM-1999-209359	
11. SUPPLEMENTARY NOTES				
12a. DISTRIBUTION/AVAILABILITY STATEMENT Unclassified-Unlimited Subject Category 27 Distribution: Standard Availability: NASA CASI (301) 621-0390			12b. DISTRIBUTION CODE	
13. ABSTRACT (Maximum 200 words)  An overview of the piezoelectric activity in amorphous piezoelectric polymers is presented. The criteria required to render a polymer piezoelectric are discussed. Although piezoelectricity is a coupling between mechanical and electrical properties, most research has concentrated on the electrical properties of potentially piezoelectric polymers. In this work, we present comparative mechanical data as a function of temperature and offer a summary of polarization and electromechanical properties for each of the polymers considered.				
14. SUBJECT TERMS Piezoelectric, Amorphous polymers, Polyimides			15. NUMBER OF PAGES 22	
			16. PRICE CODE A03	
17. SECURITY CLASSIFICATION OF REPORT Unclassified	18. SECURITY CLASSIFICATION OF THIS PAGE Unclassified	19. SECURITY CLASSIFICATION OF ABSTRACT Unclassified	20. LIMITATION OF ABSTRACT UL	