STUDY OF THE PLASTIC DEFORMATION OF FIBER-POLYMERS VIA WEIBULL MODEL

by

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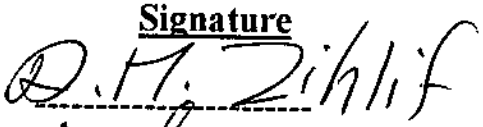
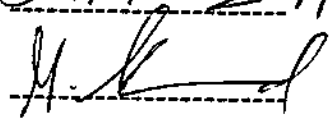
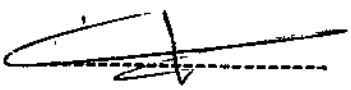
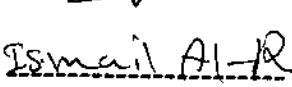
Prof. Dr. Awwad M. Zihlif, Chairman	
Prof. Dr. Mohammad Suleimen, Member	
Dr. Yahya Al-Ramadin, Member	
Dr. Ismail El-Rihail, Member	

DEDICATION

To
My Father
and
My Mother

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Abstract

Study of the Plastic Deformation of Fiber-Polymers Via Weibull Model

By

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This thesis work aims to study the plastic deformation for oriented Polypropylene films and Pan-based carbon fibers, at different gauge lengths, via Weibull statistical model. The Weibull analysis for both strength and breaking strain was carried out using the experimental data of tested above polymeric fibers. Mechanical parameters such as strength, breaking strain and Young's modulus have been studied as a function of specimen gauge length. It is found that both of strength and breaking strain of the tested fibers increase as length decrease, while the Young's modulus decreases and depend on both the strength and the breaking strain. The result analysis shows that the Weibull model is a quite good model to specify the statistical distribution for both strength and breaking strain; and adequate to characterize the plastic deformation of polymeric fibers.

CHAPTER ONE

INTRODUCTION

CHAPTER ONE

INTRODUCTION

1.1 Polymers

A polymer is constructed of smaller molecules called monomers. The atoms are held together by covalent bonds. Polymers can be built of one, two or even three different monomers and are termed homopolymer, copolymers and tripolymers, respectively (Nielson, 1962; Ward, 1971). The polymers' geometrical form can be linear, brunched, cross-link or three-dimensional net work (Fig. 1.1). The linear polymers are composed of monomers linked end to end in single chains. The monomers in cross-linked polymers are joined one to another at various positions, while the brunched polymers have side chains of molecules attached to main linear polymers (Roff and Scolt, 1971). Length to diameter ratio of polymer molecules is very high compared to that of simple molecules, this characteristic makes all the difference in their physical properties (Billmeyer, 1970).

1.2 Classification of Polymers

The polymers can be classified in different ways as:

(i) Amorphous and crystalline polymers:

When the polymer's molecules or chains arrange themselves in an order manner the polymer is said to be crystalline, and the physical proprieties such as density will be influenced by the degree of crystallinity. While the amorphous polymers are showing no crystalline order. Sometimes amorphous polymers called non-crystalline (or glassy) (Fig. 1.2). Crystalline polymers are usually stronger and more resistant to dissolution and softening by heat.

(ii) Isotropic and anisotropic polymers

Isotropy and anisotropy of polymers result from the nature of symmetry of the crystal structure. The polymers, which have the same properties in all directions and orientations are called isotropic polymers, while those polymers with observed physical properties depending on the direction (molecules orientation), are called anisotropic polymers (Hajra, 1986).

1.3 Orientation of Polymers

The orientated anisotropic polymers have chain orientation in one direction than in other direction, while unorientated isotropic polymers have the same units or chain sections pointing in all directions. There are several methods to produce oriented polymers like drawing, rolling and extrusion. The mechanical properties of the oriented polymers are not the

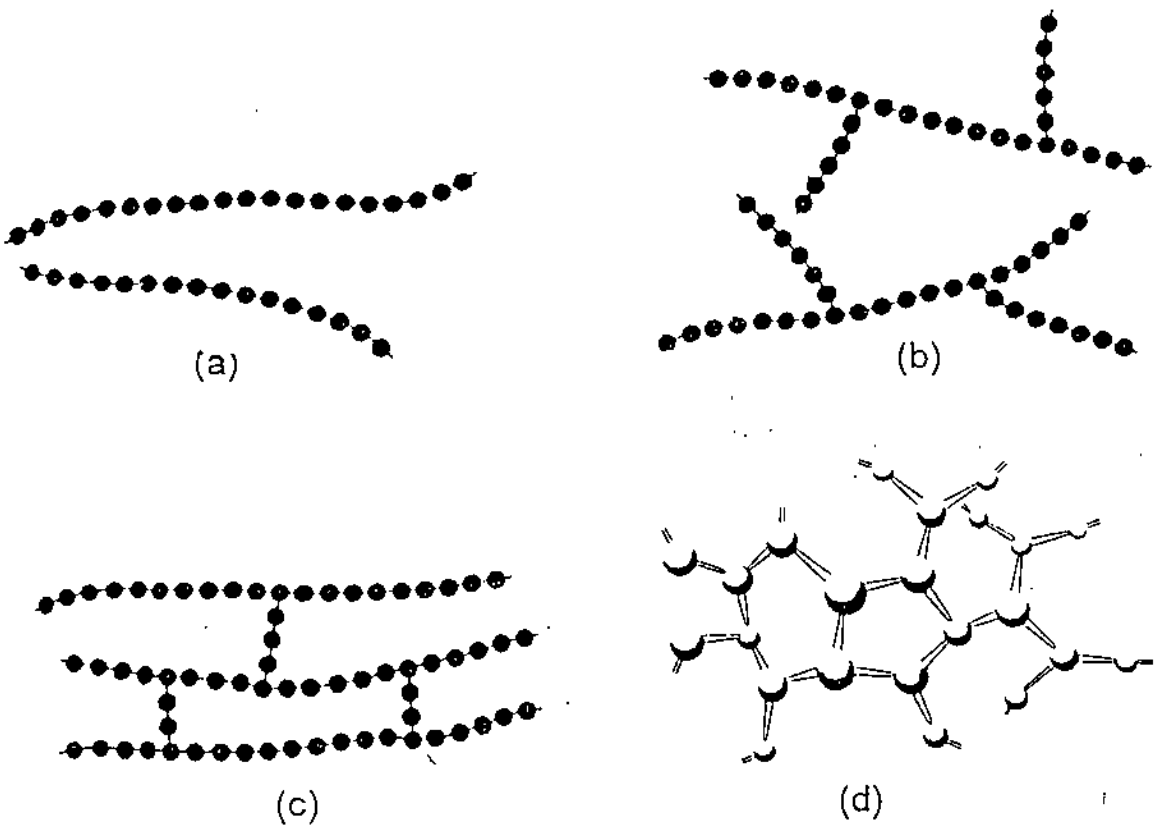


Figure 1.1 Schematic representations of (a) linear, (b) branched, (c) crosslinked, (d) network molecular structures. Circles designate individual mer units.

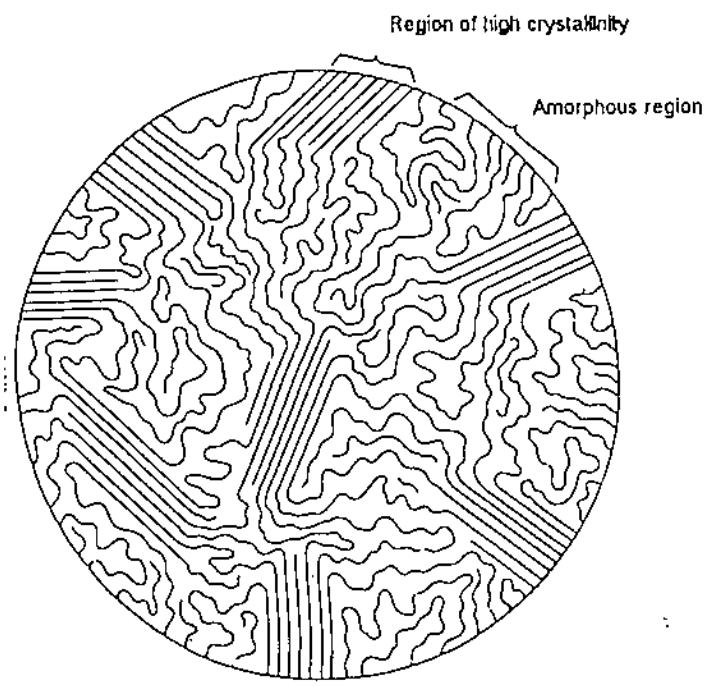


Figure 1.2 Model of a semicrystalline polymer, showing both crystalline and amorphous regions.

same, and depend strongly on the degree of orientation of the structure and on the noncrystalline regions (Zhchmann, 1979).

The oriented polymers have different properties for different directions. The mechanical behavior of oriented polymers is normally affected by the presence of microfibrils and by alternating ordered and disordered regions (Ward, 1975).

1.4 The Mechanical Properties

The physical characteristics of materials under stress are known as mechanical properties. The mechanical properties are often the most important properties, because virtually all service conditions and the majority of end-use applications involve some degree of mechanical loading.

The mechanical properties are the most immediate concern for the design engineer. The mechanical properties such as stiffness, ductility and brittleness describe the behavior of the material when it subjects to applied loads. These properties affect the performance characteristic of members of structure system (Robert, 1982).

The mechanical properties are most widely used for defining both the quality of production of the polymeric materials and their design, i.e., engineering behavior (Andrews, 1968). The mechanical properties are specified with many parameters such as modulus of elasticity, yield stress and yield strain, ductility toughness and others. The mechanical behavior of

a material reflects the relationship between its deformation to an applied load and is ascertained by performing carefully designed laboratory experiments that replicate as nearly as possible the service conditions.

1.5 The Polypropylene Polymer

The monomer of propylene is CH_2=$CHCH_3$. It is obtained from natural gas or light oil fraction in petroleum. The gas is cracked at temperature about 800°C and polypropylene films are separated from C_3 fractions by successive distillation. There are many different grades of polypropylene fall into three basic groups: homopolymers, copolymers and reinforced. The properties of homopolymers vary with the molecular weight distribution and the degree of crystallinity.

Polypropylene films are reinforced with glass or asbestos fibers to improve their mechanical properties and to increase resistance to deformation of high temperature (Hanat, 1978; Roff and Scolt, 1971).

Polypropylene is available as clear biaxially oriented sheets and as fibers. The uniaxially and biaxially oriented films have been provided by drawing and extruding processes (Kawamoto *et al.*, 1998). Polypropylene films are milky white semi-translucent, light in weight, resistance to heat and inert chemicals. It also has negligible water absorption, excellent electrical resistivity and ability in thin sections. The good combination of properties makes Polypropylene films suitable

for many appliance and automotive products such as cable covers insulators and used as general propose molding materials for storing goods and crops (Clasuer, 1975).

1.6 Carbon Fibers

The fiber is defined as a unit of matter having a length at least 100 times greater than its width or diameter. There are two kinds of fibers:

1- Natural fibers as silk, cotton and wool.

2- The artificial fibers as nylon, carbon, glass and steel fibers.

Carbon fibers are categorized as having high tensile strength. So that, carbon fibers is used in the table and hardening material and it is an excellent lubricant. It is used for reinforcement of composites (Jastrzebski, 1977). Carbon fibers are manufactured from three different precursors Pan, Rayon and Pitch. Pan was the original precursor for high strength carbon fibers. Pan is typically copolymers; and the Pan precursor has the light degree of molecular orientation and purity (Weisfeld, 1971).

The manufacture process of carbon fibers from Pan polymer includes three stages (Jones, 1994):

1- Heat treatment:

A ladder polymers consisting of six-membered rings is formed owing to the linkage between the groups. The ladder polymer is more stable towards heat than the original chain and therefore dos not melt easily.

2- Carbonization

Carbonization is carried out at the temperature about 1000°C. During this stage, the fibers must be kept out of contact with air, and it is usual to use an inert gas. In this stage the organic fiber is changed to carbon, and about 50% loss in weight may occur.

3- Graphitization

It occurs at high temperatures about 2800°C. This stage is necessary for the production of high-modulus carbon fibers with better electrical and thermal conductivity. Carbon fibers sometimes called graphite fibers to distinguish them from lower forms of carbon-based materials. The percentage of graphite in the fibers depends on the final processing temperature and their ability to graphitize.

1.7 Previous Work

There are many studies carried out on the deformation of polymeric fibers. Those studies have concentrated on the effect of strain rate, temperature, gauge length (Ugbolne and Uzomoh, 1995; Poudeyhimi *et al.*, 1996).

The previous investigations indicated that the testing rate affects the fibers breaking strain and Young's modulus. Duckett and Zihlif (1974) carried out extensive studied on anisotropy of plastic deformation in the isotropic and oriented Polypropylene; they noticed differences

between the tensile and compressive yield behavior. They concluded that the mode of deformation is quite complicated and depends on the testing conditions, hydrostatic component of stress and the orientation anisotropy. El-Abadla and Zihlif (1992) studied the mechanical properties of Polypropylene drawn films, they found that the mechanical properties of these films have temperature and strain rate dependence. Also in this study they performed preliminary investigation on the deformation using the Weibull model.

Ahmad, El-Rihail, Yasin and Zihlif (1987) have studied the mechanical and the electrical properties of Pan-based carbon fibers filaments, they found that the values of the electrical resistivity and elastic modulii are strongly dependent on the annealing temperatures of the carbon fibers.

Nofel and Zihlif (1995) studied the mechanical properties of Pan-based carbon fibers / polycarbonate composite, they found that the values of both Young's modulus and yield stress increases with the filler content. It was explained by the occurrence of some structural changes, which take place during aging process.

There are few reports dealing with the fitness of the Weibull theory to deformation of polymeric fibers (Gao and Dongling, 1993; Beyerlein, 1996). The Weibull distribution function is proposed to characterize the length dependence of fiber strengths (Zhu *et al*, 1995). The strengths are evaluated in terms of Weibull statistics (Nakamura and Mikito, 1997). Pan, *et al* (1997) studies the fitness of Weibull model for

polymeric fibers as polyamide 66 and polyester. They showed that the Weibull model is quite technique good to specify the statistical distribution of both breaking strain and strength of polymeric fibers.

1.8 The Present Work

This thesis work covers a study on the mechanical properties of Pan-based carbon fibers and Polypropylene drawn films in tension under different conditions using Weibull distribution. The research covers the followings:

1) Studying the effect of gauge length on the tensile properties of Polypropylene films and Pan-based carbon fibers.

2) Studying the effect of constant strain rate on the mechanical properties of Polypropylene films and Pan-based carbon fibers.

3) Determining the Weibull parameters through fitting the Weibull model.

4) Studying the deformation for strength and the breaking strain via Weibull model.

5) Discussing the obtain results.

The present thesis is organized as:

- Chapter One: Introduction
- Chapter Two: Definition and theoretical aspects.
- Chapter Three: Experimental work.
- Chapter Four: Results and discussion.
- Chapter Five: Conclusion and future work.

CHAPTER TWO

DEFINITION & THEORITICAL ASPECTS

CHAPTER TWO

DEFINITION &THEORETICAL ASPECTS

2.1 Definitions of Some Mechanical Quantities

2.1.1 Stress:

Let a specimen with gauge length L_0 and cross-section area A_0 be subjected to tensile load. The engineering stress(σ) is defined as:

$$\sigma = \frac{F}{A_0} \qquad (2.1)$$

Where, F is the instantaneous load applied perpendicular to the specimen cross-section area, and A_0 is the cross section area before any load is applied. While the true stress σ_t is the load divided by the current cross-sectional area of the sample and can be written as:

$$\sigma t = \frac{F}{A} \qquad (2.2)$$

Where, A is the current area.

2.1.2 Strain:

Similarly, the engineering strain (ε) is defined as:

$$\varepsilon = \frac{\pm(L_i - L_0)}{L_0} = \frac{\pm\Delta L}{L_0} \qquad (2.3)$$

Where L_0 is the initial length of the sample before any loading is applied, and L_1 is the instantaneously length. The true strain (ε_t) is defined as the integral of the ratio of the incremental change in length to the current length of the sample.

$$\varepsilon_t = \int_{L0}^{L1} \frac{\pm \partial L}{L} = \pm \mathrm{Ln}\left(\frac{L_1}{L_0}\right) \text{ - - - - - - - - - - - - - - - } (2.4)$$

The true strain can be related to engineering strain through the relation

$$\varepsilon_t = \pm \mathrm{Ln}\ (1 \pm \varepsilon) \text{ - - - - - - - - - - - - - - - } (2.5)$$

The two quantities are identical at small strain and the divergence only becomes serious above 10%. If the volume of the sample is constant during loading, then the true stress can be related to engineering stress and strain as follows:

$$\sigma_t = \sigma(1 \pm \varepsilon) \text{ - - - - - - - - - - - - - - - - - - } (2.6)$$

2.1.3 Young's Modulus:

Young's elastic modulus (E) is the ratio of stress to the corresponding strain within the elastic region of the deformation and can be written as:

$$E = \frac{\sigma}{\varepsilon} \quad \text{-----------------------} (2.7)$$

Where (σ) and (ε) are the engineering stress and strain respectively. The units of the stress and Young's modulus are the same (Nm^{-2}) or Pascal (Pa).

2.1.4 Tensile Strength

The value of largest stress which is required to sample rupture. Tensile strength (σ_t) is written as:

$$\text{Tensile strength}(\sigma_t) = \frac{\text{maximum load}}{\text{cross-section area}} \quad \text{----------} (2.8)$$

2.1.5 Breaking Strain :

The breaking strain is the strain value when the material is broken or ruptured (Haward, 1973).

2.2 Stress- Strain Curves

Some of the mechanical characteristics of the material can be deduced from the stress-strain curves obtained through deformation tests. This is usually done by loading the test specimen at constant strain rate,

and then plotting the values of stress as ordinate and the value of the strain as abscissas from the obtained load-extension curves.

A stress- strain curve for ductile plastic is shown schematically in Fig. (2.1). The initial portion of the stress-strain curve between points A and B is the elastic region and follows Hook's law of elasticity. The point B is the "elastic limit". Deformation in the linear range is called elastic deformation, and the Young's modulus is determined from the slope of this linear part of the stress-strain curve. The point C, which is the maximum point on the stress-strain curve, is called the yield point. The plastic range includes the region where the deformation is unrecoverable (B to D). The stress corresponding to the "yield point" is termed as yield stress and denoted by (σ_y). and the strain at this point is termed the "yield strain" and denoted by (ε_y) (Callister, 1991; Billmeyer, 1970; Haward, 1977). Sometime the yield point is not well defined, it is determined from the intersection of the two tangents taken on stress strain-curve as shown in the Fig. (2.2).

2.3 The Weibull Distribution

The Weibull distribution is useful in a great variety of applications particularly as a model for the product life. It has also been used as the

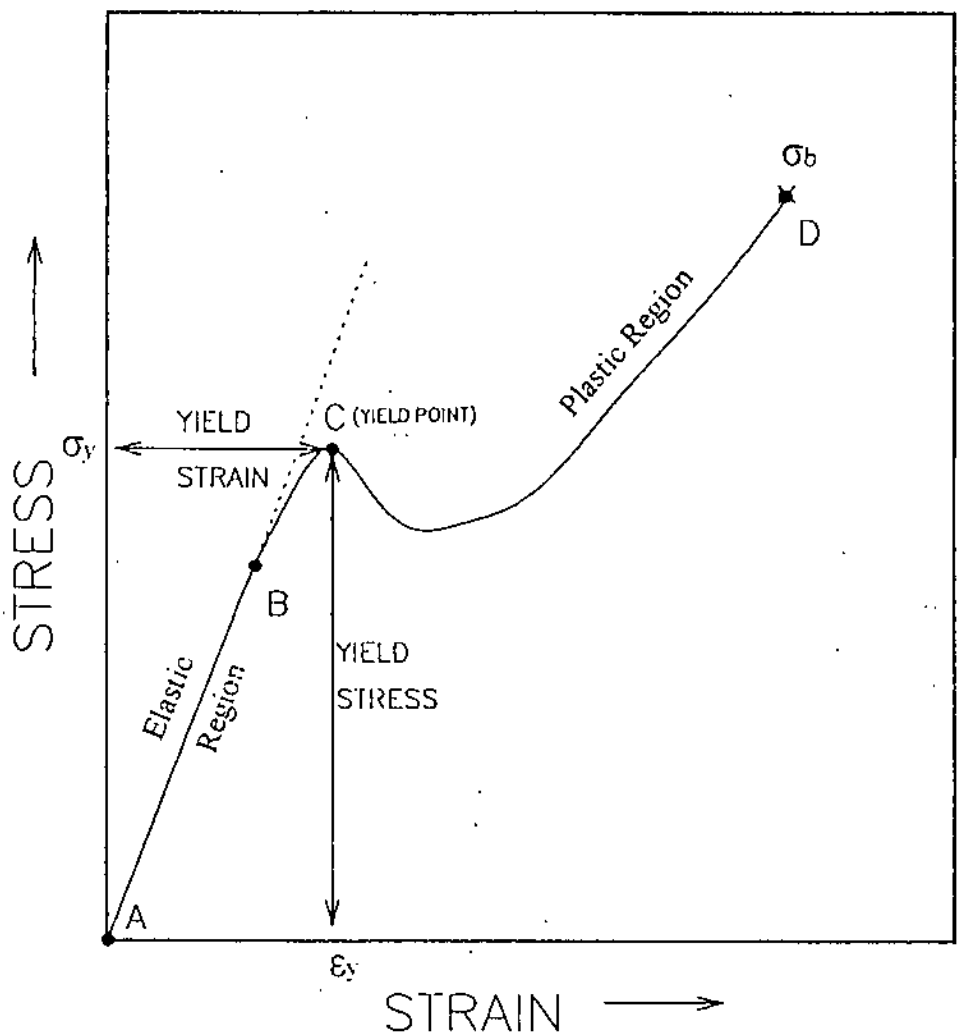


Figure 2.1 The generalized stress-strain curve for a deformed polymer.

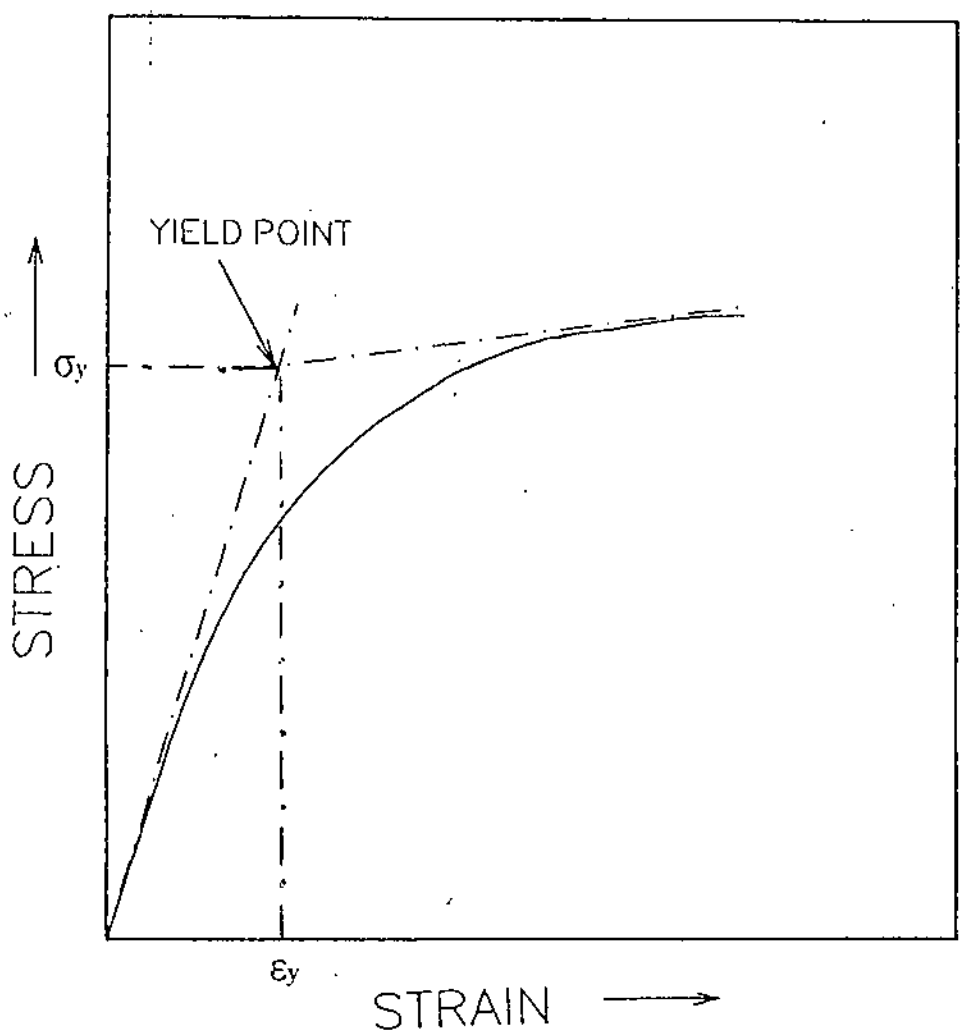


Figure 2.2 Determination of not well defined yield point from the intersection of two tangents.

distribution of strength of certain materials. It is suitable for weakest-link type product. One reason for its popularity is that it has a great variety of shapes. This makes it empirically fits many classes of data (Crowder *et al.*, 1991).

If the failure of component (specimen) is due to the failure of the weakest links, then it's strength and breaking strain can be written using Weibull statistic. The Weibull distribution is given by (Mann *et al.*, 1974):

$$P(\sigma_f) = 1 - \exp\left[-\left(\frac{\sigma_f}{\alpha}\right)^{\beta}\right] \quad \text{----------} \quad (2.9)$$

Where, α is known as scale parameter and its unit is the same as the strength unit or breaking strain and β is the shape parameter. The last equation can be written as:

$$1 - p = \exp\left[-\left(\frac{\sigma_f}{\alpha}\right)^{\beta}\right] \quad \text{----------} \quad (2.10)$$

Taking the natural logarithm twice gives :

$$\mathrm{Ln}\left[\mathrm{Ln}\left(\frac{1}{1-p}\right)\right] = [\beta \times \mathrm{Ln}(\sigma_f)] - \beta \,\mathrm{Ln}\,\alpha \quad \text{---------} \quad (2.11)$$

The parameters of the Weibull distribution (α & β) can be determined by using the maximum likelihood estimation technique, which gives two equations (Gong, 1997):

$$\frac{n}{\beta} + \sum_{i=1}^{n} \mathrm{Ln}(\sigma i) - \left[\frac{n \sum_{i=1}^{n} \left((\sigma i)^{\beta} \mathrm{Ln} \sigma i \right)}{\sum_{i=1}^{n} (\sigma i)^{\beta}} \right] = 0 \qquad \text{(2.12)}$$

$$\alpha = \left(\frac{\sum_{i=1}^{n} (\sigma i)^{\beta}}{n} \right)^{1/\beta} \qquad \text{(2.13)}$$

The last two equations can be solved by bisection method (Haddad, 1994). The program that has been used to solve these equations is written in Gwbasic and given as program I in the appendix.

CHAPTER THREE

EXPERIMENTAL WORK

CHAPTER THREE

EXPERIMINTAL WORK

3.1 Material

Two test materials were used in this research work, drawn Polypropylene films and Pan-based carbon fibers. The Polypropylene tapes filled slightly with calcium carbonate (7wt%) manufactured by Raffia Company in Amman-Jordan. Pan-based carbon fibers provided from Celanesse Company in USA in form of stands 6000 fibers and cross section is 0.24 mm^2 and Density is 1.77 g/cm^3(Nofel, 1995).

3.2 The Tensile Tests

The tensile tests were carried at different specimen lengths(20mm, 50mm, 70mm, and 100mm) using the Instron machine model 1026 Fig. (3.2) with fitted load cell of maximum load capacity of 5 kN. Tensile tests were performed by holding the specimens in special tensile crips bought from the Instron Company.

The recorder system of this machine is sensitive enough to record even the smallest or quickest change in the load with the good accuracy. The direct output obtained from the Instron chart recorder is the load-time curve, which represents the applied load on the sample as a

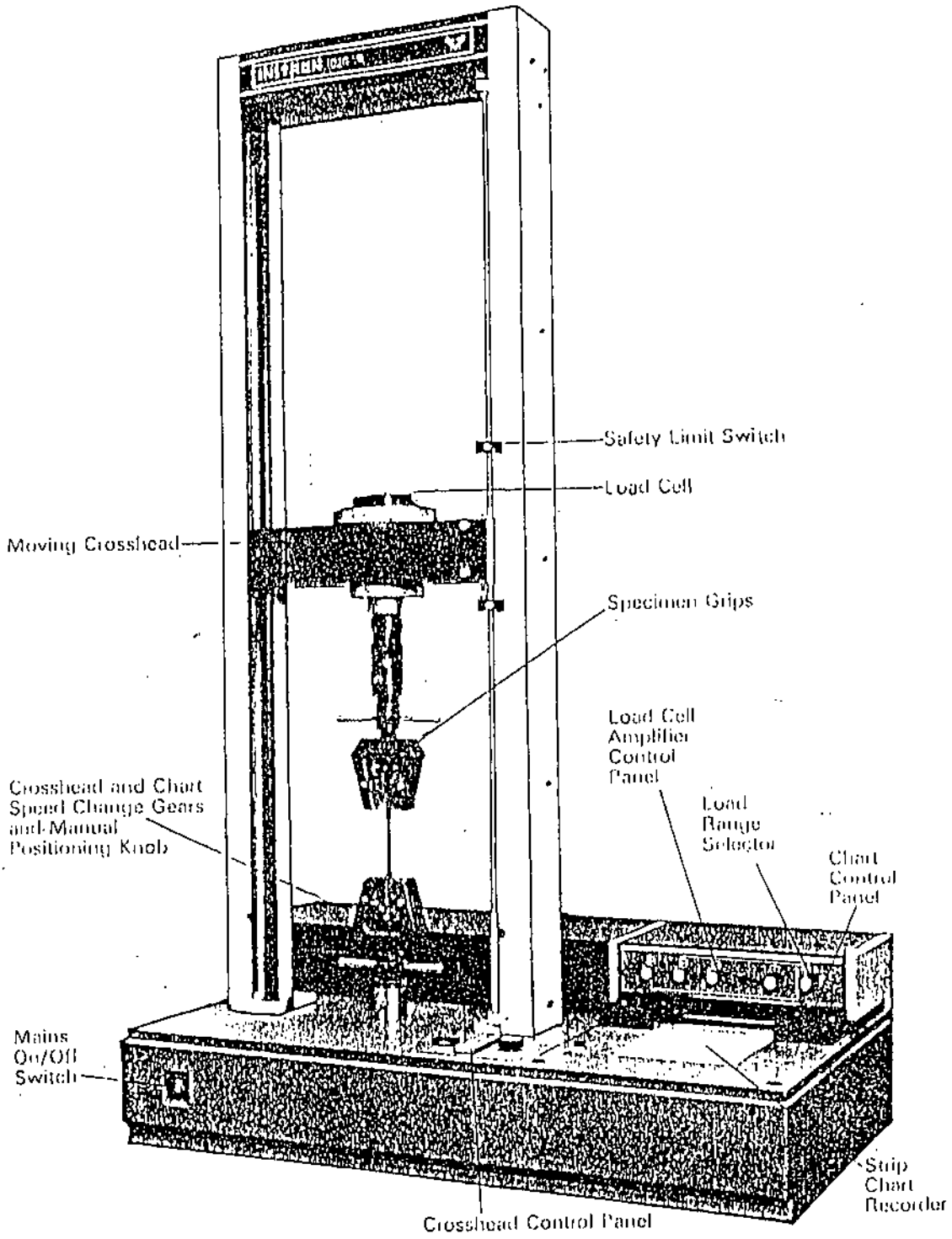


Figure 3.1 The Instron testing machine model 1026 in tension setup

function of time. The load-time curves were converted to stress-strain curves by using cross-sectional area, crosshead speed, and chart speed according to the equations:

$$\text{Stress}(\sigma_t) = \frac{\text{applied force}}{\text{cross - sectional area}} \quad \text{----------} \quad (3.1)$$

$$\text{Strain}(\varepsilon_t) = \frac{\text{elongation}}{\text{guage length}} \quad \text{----------} \quad (3.2)$$

but

$$\text{elongation} = \text{crosshead speed} \times \text{time} \quad \text{----------} \quad (3.3)$$

and

$$\text{time} = \frac{\text{distance travelled on chart}}{\text{chart speed}} \quad \text{----------} \quad (3.4)$$

then

$$\text{Strain}(\varepsilon_t) = \frac{\text{crosshead speed} \times \text{distance on chart}}{\text{guage length} \times \text{chart speed}} \quad \text{----------} \quad (3.5)$$

Thus the stress-strain curves can be plotted. The parameters detected from the stress-strain curves are: breaking stress and strain, yield stress, yield strain, and the Young's modulus. The Young's modulus was calculated

from the Hookian part of the stress-strain curves. The strain rate can be calculated by using the equation:

$$\text{Strain-Rate}(\dot{\varepsilon}) = \frac{\text{crosshead speed}}{\text{guage length}} \quad \text{---------- (3.6)}$$

Where, the crosshead speed represents the elongation rate, and the gauge length is the initial length of the specimen.

The estimated error in the collected stress-strain data is estimated about 5%.

CHAPTER FOUR

RESULTS & DISCUSSION

CHAPTER FOUR

RESULTS & DISCUSSION

The obtained experimental results deal with the mechanical properties of Pan-based carbon fibers and Polypropylene oriented films. Weibull analysis of the stress-strain curves of the deformed specimens in tensile tests is used. Also determination of some mechanical parameters will be presented and discussed in this section of the thesis.

4.1 Stress-Strain Behavior

The deformation behavior is investigated under different conditions for Polypropylene drawn films and Pan-based carbon fibers (El-Rhail, 1984; Zihlif *et al.*, 1987; Ugbolue and Uzomah, 1995). The obtained stress-strain curves are shown in Figs. (4.1), (4.2), (4.3) and (4.4) for Polypropylene films; and the stress-strain for Pan-based carbon fibers are shown in Figs. (4.5), (4.6), (4.7) and (4.8). It was noticed that the mechanical parameters, such as strength and breaking strain change with the gauge length. Fig. (4.9) shows the stress-strain behavior of Polypropylene films for three gauge lengths, and Fig. (4.10) shows that the stress-strain behavior of Pan-based carbon fibers for two gauge lengths. We can notice that increasing the gauge length decreases the breaking

strain. The stress-strain curves have nearly the same shape for Polypropylene films and for Pan-based carbon fibers. The deformation stage sample still exhibiting a ductile behavior, but the observed difference lies in the slope of the first Hookian part of each curve and in the location of strength and breaking strain points.

4.2 Gauge Length Effects on the Tensile Strength and Breaking Strain

The variations of the tensile strength with gauge length are shown in Figs. (4.11) and (4.12) for Polypropylene films and carbon fibers, respectively. The relationship between the gauge length and breaking-strain are shown in Figs. (4.13) and (4.14) for Polypropylene films and carbon fibers, respectively. Table (4.1) shows the values of the strength and breaking strain at different gauge lengths, and estimated standard deviation (SD) for Polypropylene drawn films and carbon fibers.

It was found that the average tensile strength and breaking strain decrease exponentially with the length of these samples. This decrease in the tensile strength and breaking strain with gauge length may be attributed to some the internal defects and surface flaws (El-Rihail, 1984; Jones, 1994). The standard deviation of the fiber strength and breaking strain increase when the fibers gauge length decreases.

Table (4-1) Fibers mechanical properties and (SD) values (in parenthesis) tested at constant rate of extension of 20-mm min^{-1}.

FIBERS	GAUGE LENGTH (mm)	20	50	70	100
PP	Breaking strain(%)	27.171(4.95)	23.24(3.49)	18.933(2.15)	15.62(1.925)
	Strength(Gpa)	0.371(0.054)	0.352(0.046)	0.315(0.004)	0.278(0.039)
	Initial modulus(Gpa)	3.249(0.417)	4.496(0.714)	5.386(0.738)	6.236(0.826)
CARBON FIBERS	Breaking strain(%)	2.764(0.279)	1.166(0.072)	0.834(0.016)	0.637(0.011)
	Strength(Gpa)	2.727(0.289)	2.093(0.257)	1.642(0.176)	1.289(0.162)
	Young's modulus(Gpa)	94.806(13.9)	178.75(15.17)	197.6(16.54)	203.9(23.5)

When the length of the tested fiber increases the possibility of existence of randomly distributed defects and surfaces flaws increases. In other words, a fiber with shorter length becomes stronger, more extensible; but lower stiffness in extension. Zhu (1995) reported similar results in case of deforming brittle ceramic fibers.

4.3 Strain Rate Effect

The comparison between the observed data taken at constant strain-rate CSR (same strain rate for all different gauge lengths) and taken at constant rate of extension CRE (strain-rate changes from gauge length to another) is illustrated in Figs. (4.15) and (4.16) for Polypropylene films and also in Figs. (4.17) and (4.18) for Pan-based carbon fibers for strength and breaking strain, respectively.

Table (4-2) Fibers mechanical properties and (SD) values (in parenthesis) tested at constant strain rate of 100% min^{-1}.

FIBERS	GAUGE LENGTH (mm)	20	50	100
PP	Breaking strain(%)	27.171(3.95)	25.56(3.699)	18.11(2.087)
	Strength(Gpa)	0.371(0.044)	0.372(0.041)	0.308(0.03)
	Initial modulus(Gpa)	3.249(0.317)	4.033(0.862)	5.296(0.819)
CARBON FIBERS	Breaking strain(%)	2.764(0.279)	1.252(0.2)	0.732(0.121)
	Strength(Gpa)	2.727(0.289)	2.192(0.216)	1.4 (0.179)
	Young's modulus(Gpa)	94.806(14.9)	176.89(16.17)	192.9 (16.81)

The results for Polypropylene drawn films and carbon fibers at both (CSR) and (CRE) are shown in tables (4.1) and (4.2). It is seen that the breaking strain collected at (CRE) of 20mm min^{-1} is lower than breaking strain collected at CSR of 100% min^{-1}. When the strain rate increases the tensile strength of the tested material increases. Therefore at longer gauge length, CRE corresponds to lower strain rate. The effect of strain rate appears to decrease at shorter gauge lengths. The above deformation behavior can be explained by considering the fragmentation process takes place inside the fiber structure, which initially appears for the data collected at CRE (Pan *et al.*, 1997). Therefore, it appeared more rational here to carry out the tensile tests at CRE by fixing the crosshead speed at 20mm min^{-1}.

4.4 Gauge Length Effect on Young's Modulus

The gauge length influence on the Young's modulus is shown in Figs. (4.19) and (4.20) for Polypropylene films and pan-based carbon fibers, respectively. The stress-strain curves at different gauge length are presented in Figs. (4.9) and (4.10) for Polypropylene films and carbon fibers, respectively. We can conclude that when the gauge length decreases the Young's modulus decreases; in other words the smaller length leads to lower modulus. The Young's modulus is dependent on both strength and breaking strain. It is seen from the results in table (4.3) that reduction of the gauge length of tested fibers increases both strength and breaking strain; but the increase in the values of breaking strain is greater than that of the strength, which leads to a decrease in the elastic modulus. The size effect on Young's modulus reveals that the increase in both strength and breaking strain of the specimen with smaller size are the results accumulated from every point of material. Also the size effect influences the strength and breaking strain of the material to different magnitude, causing a change in the observed modulus (Brostow and Coreliussen, 1986).

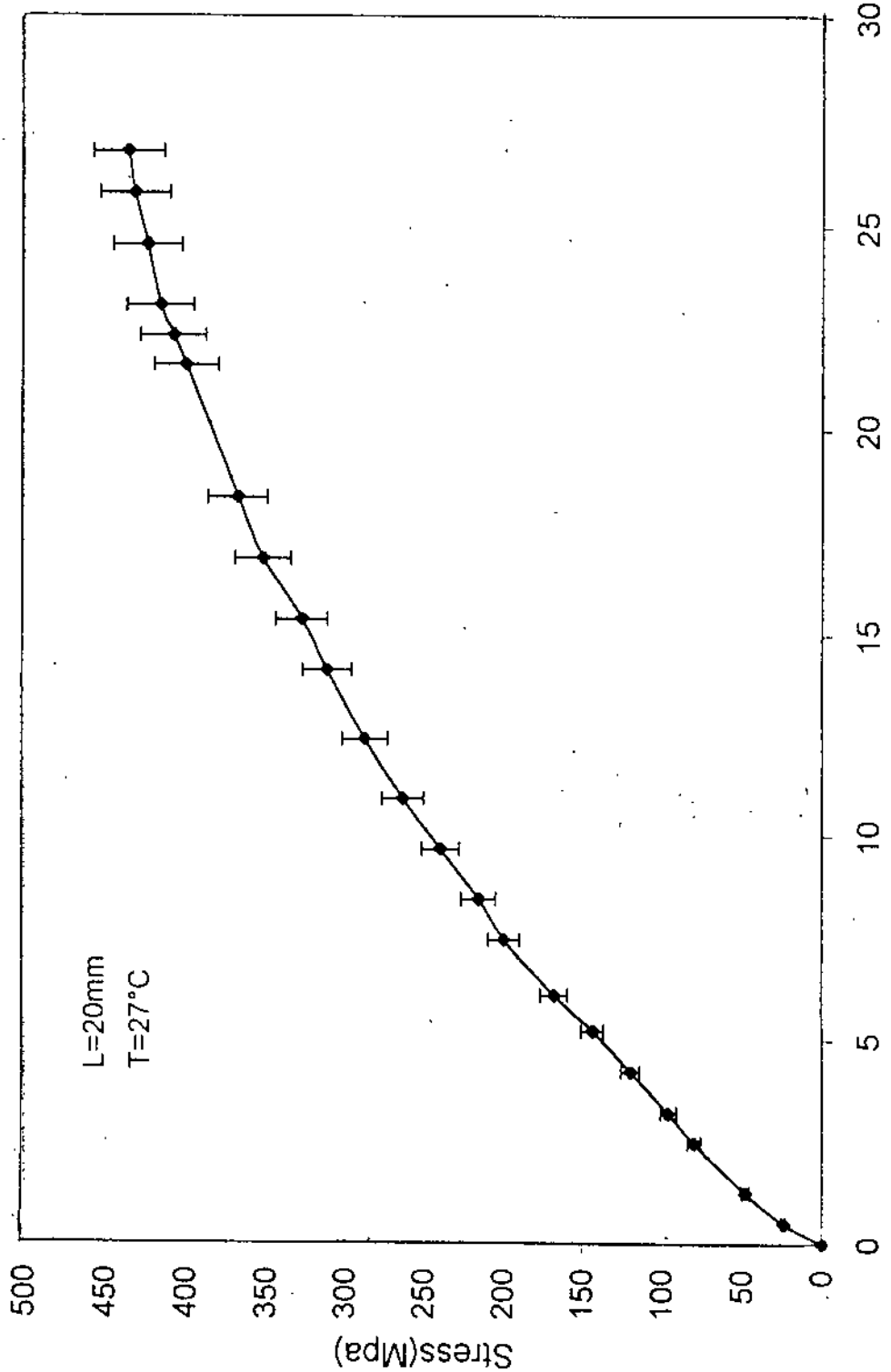
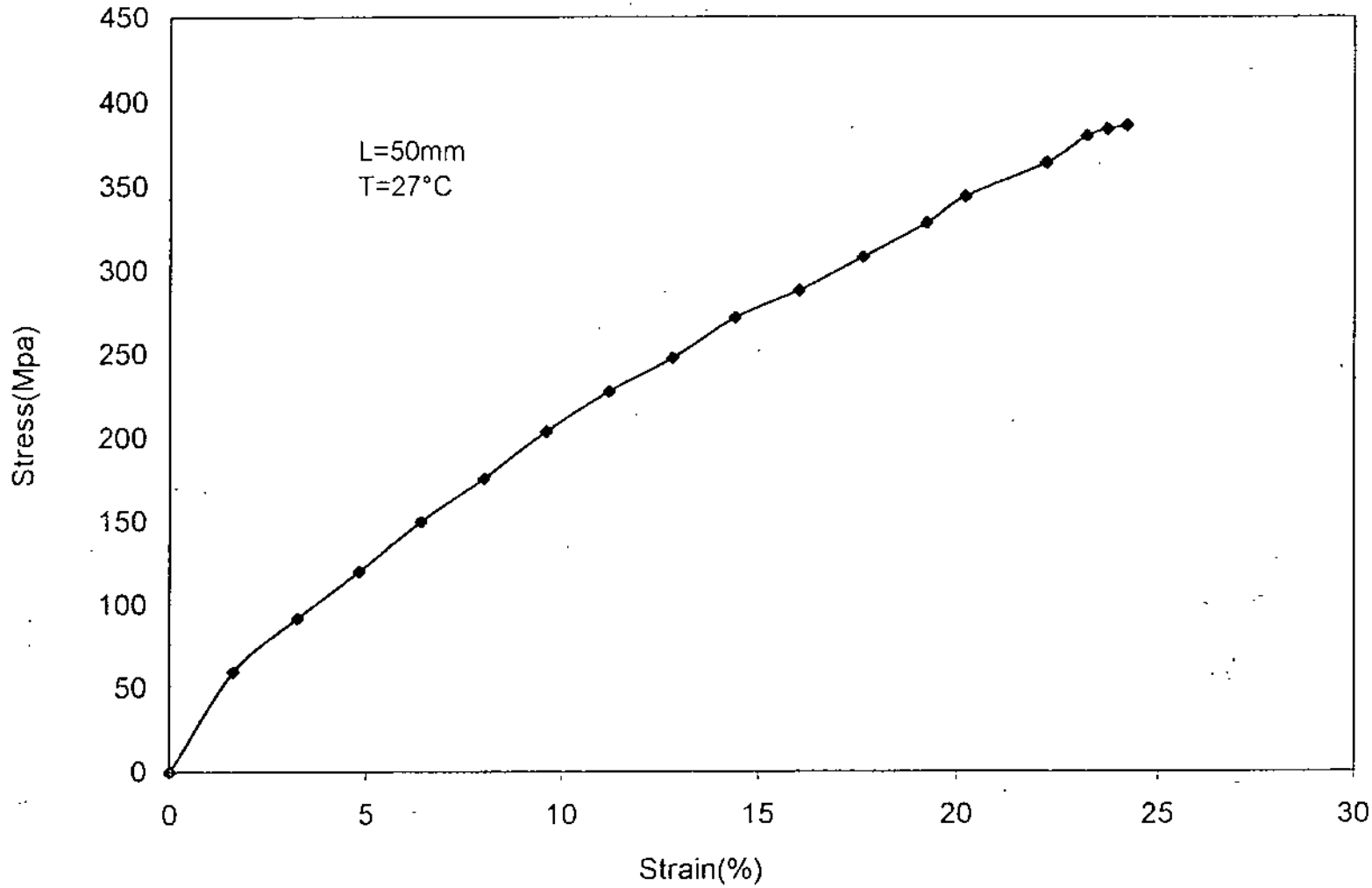


Figure (4-1) Stress-strain curve for PP films



Figure(4-2) Stress-strain curve for PP films

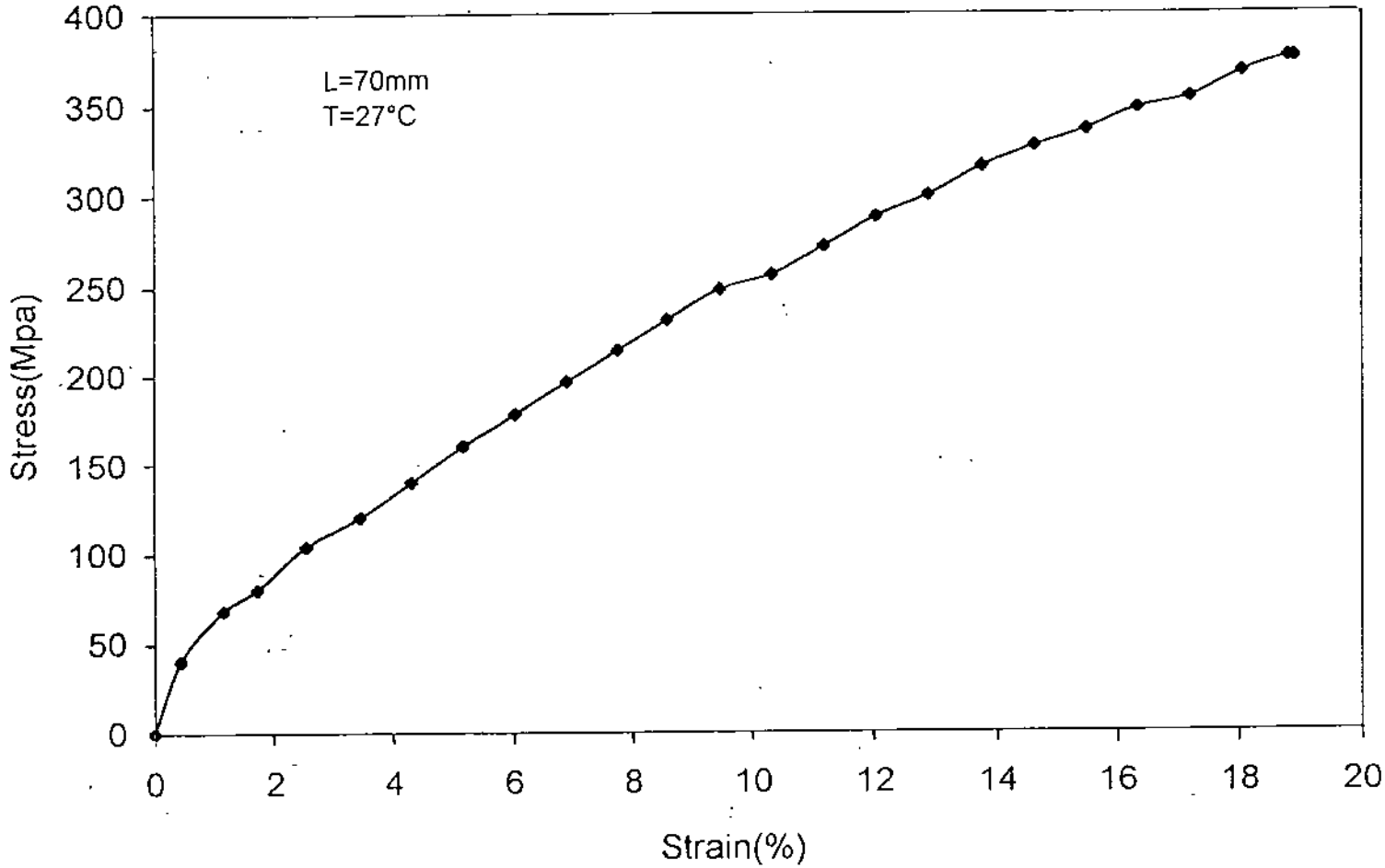


Figure (4-3) Stress-Strain curve for PP films

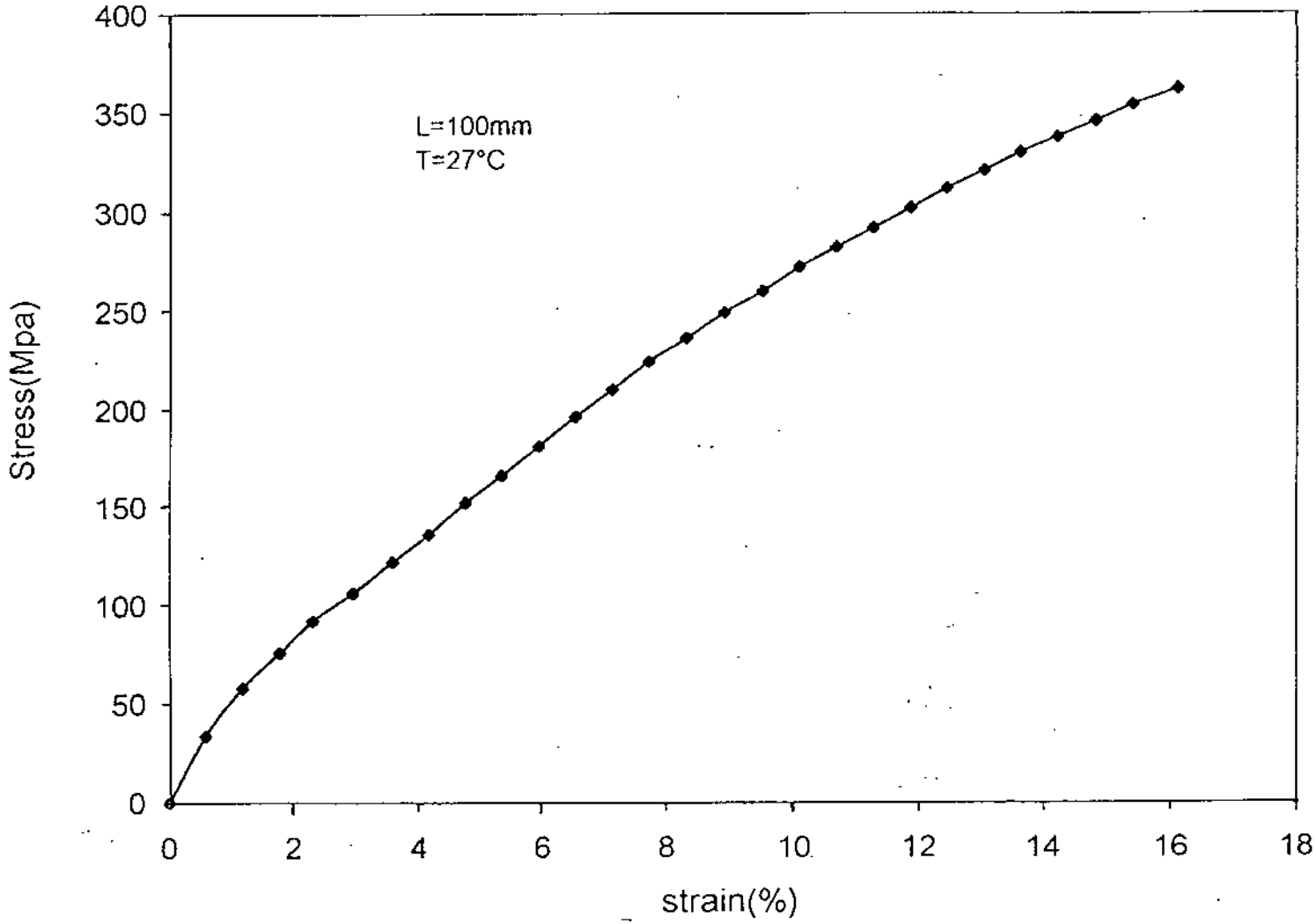
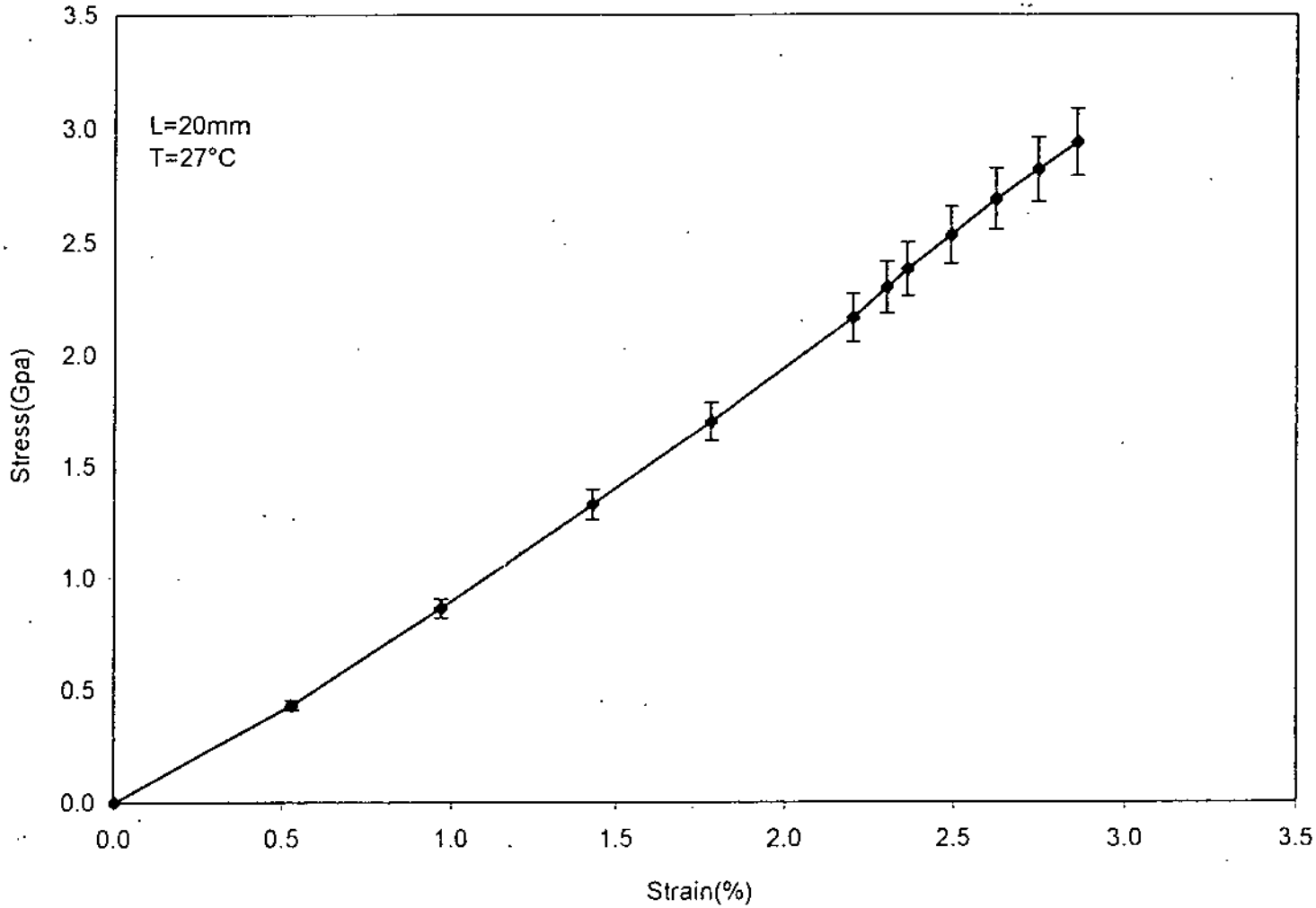


Figure (4-4) Stress-strain curve for PP flims



Figure(4-5) Stress-strain curve for carbon fibers

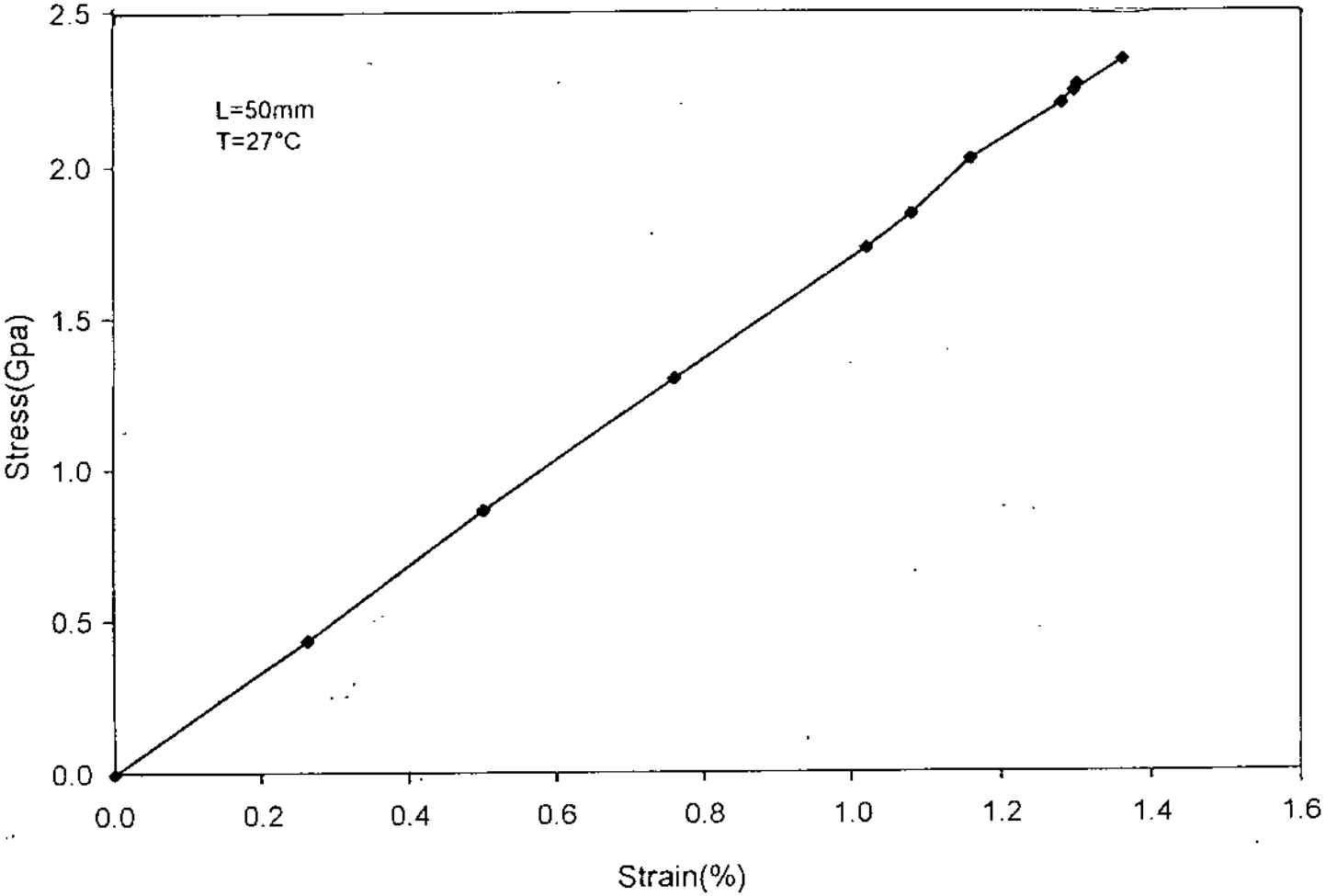


Figure (4-6) Stress-strain curve for carbon fibers

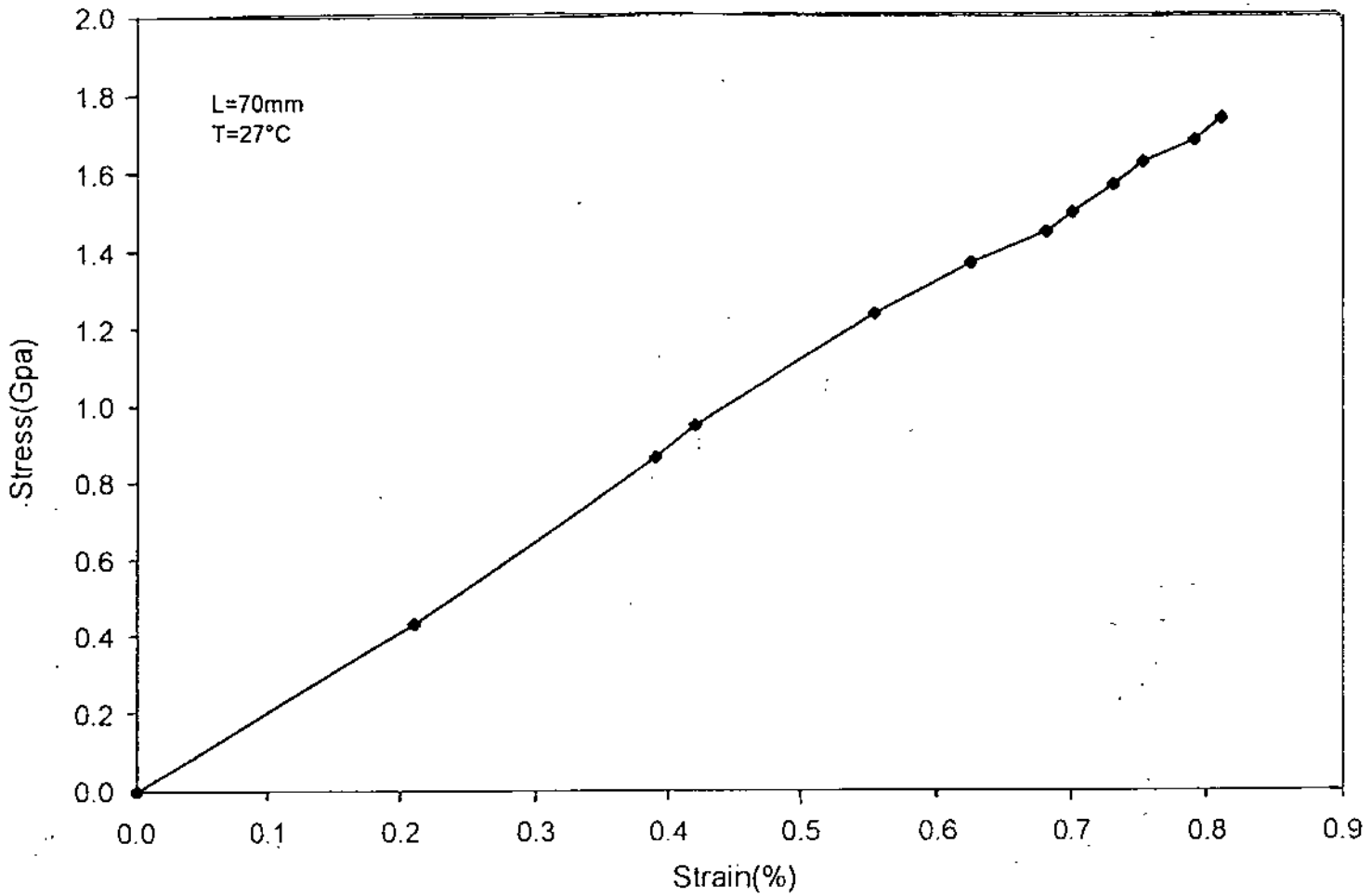


Figure (4-7) Stress-strain curve for carbon fibers

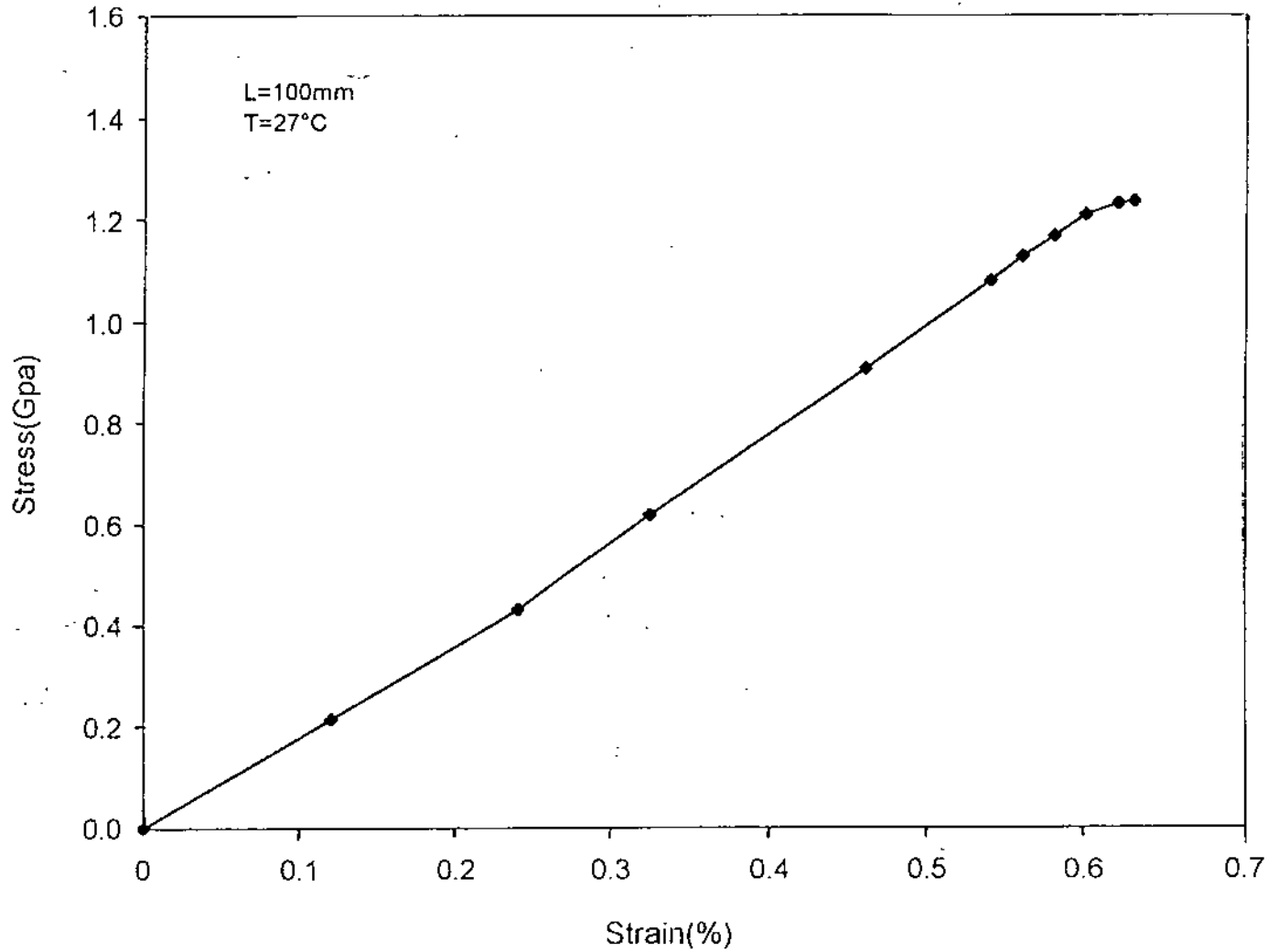
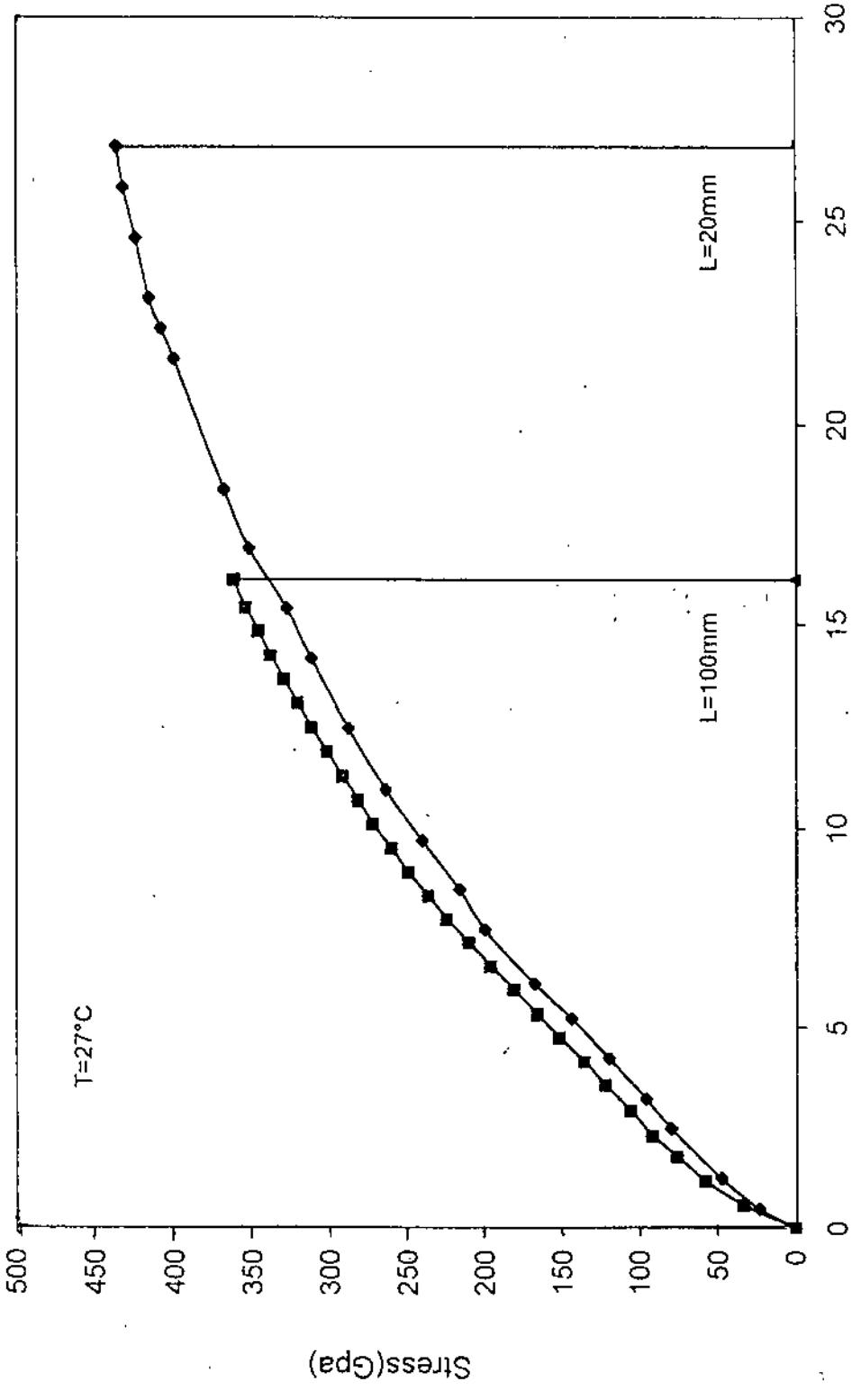


Figure (4-8) Stress-strain curve for carbon fibres



Figure(4-9) Stress-Strain curves at different guage length for PP films

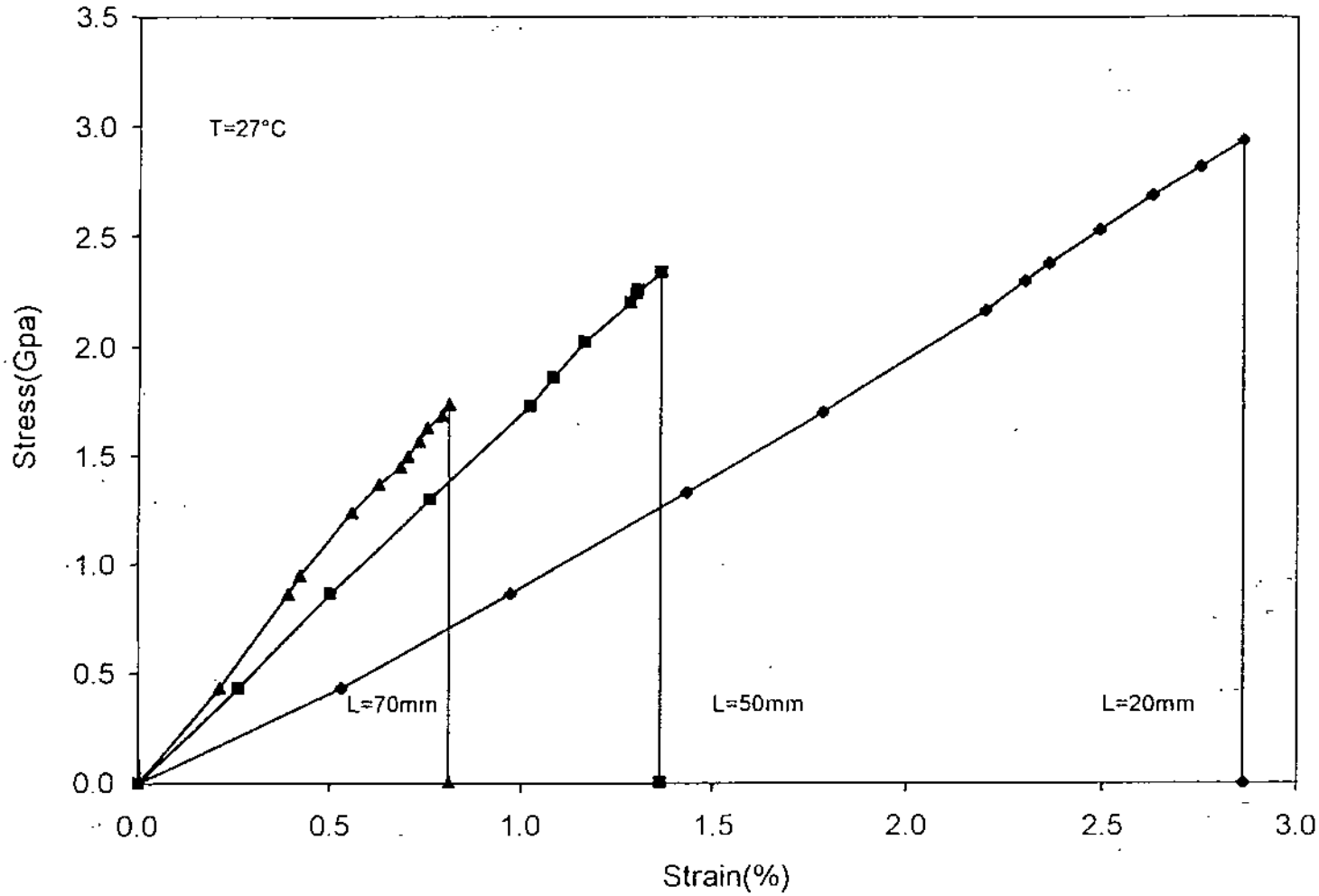
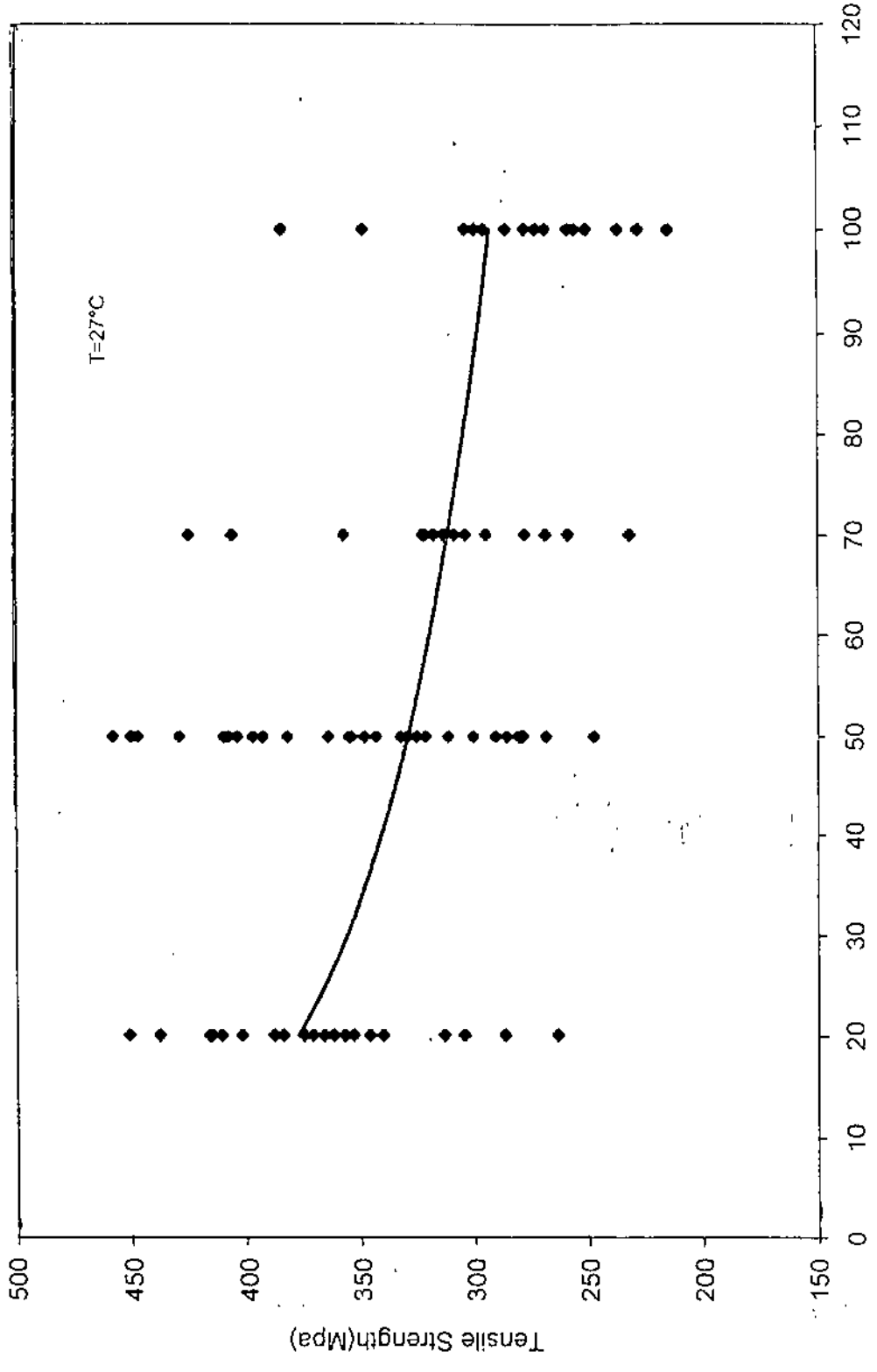


Figure (4-10) Stress-strain curves at different guage length for carbon fiber



Figure(4-11) Strength versus Gauge Length for PP films

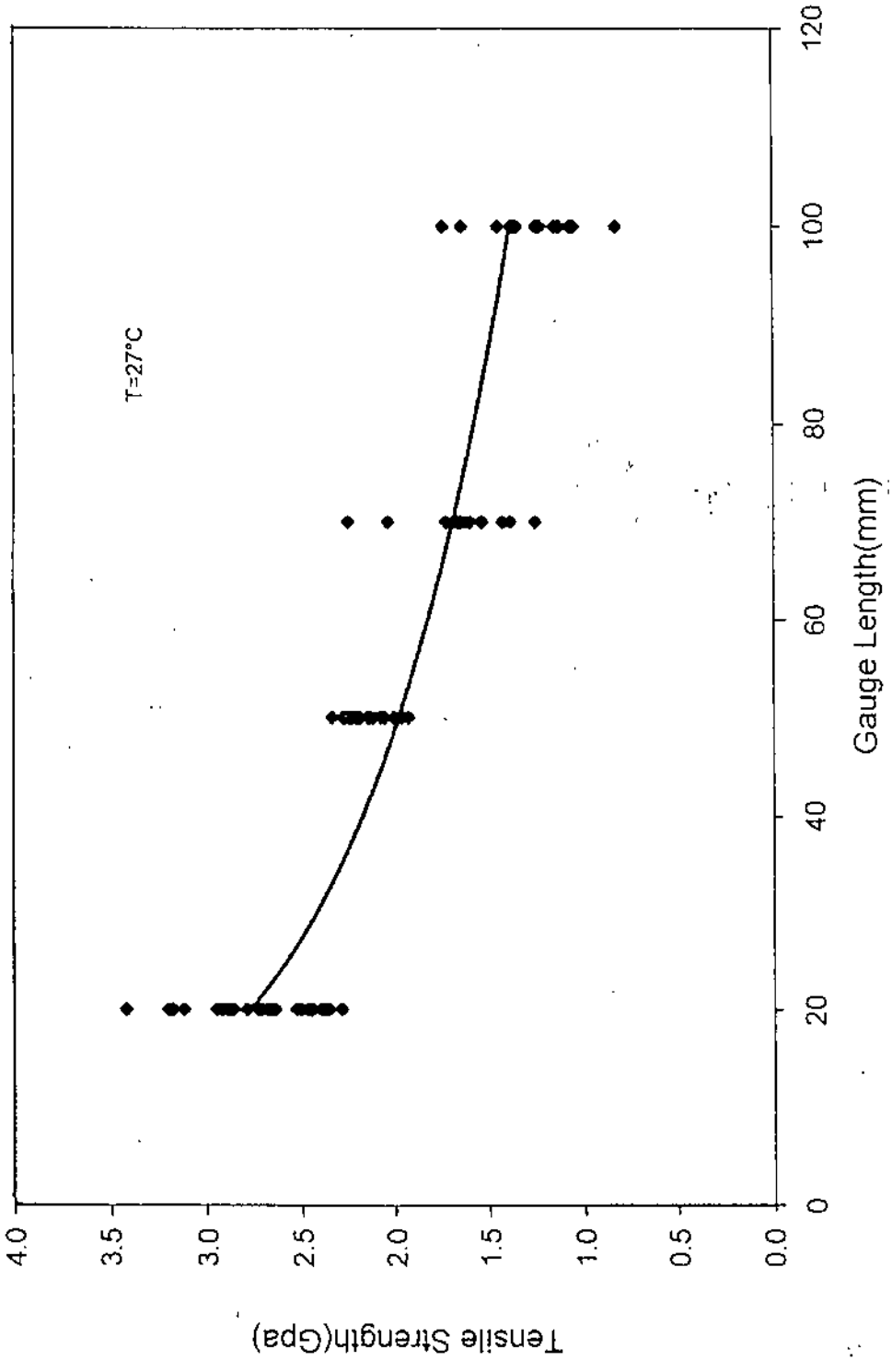
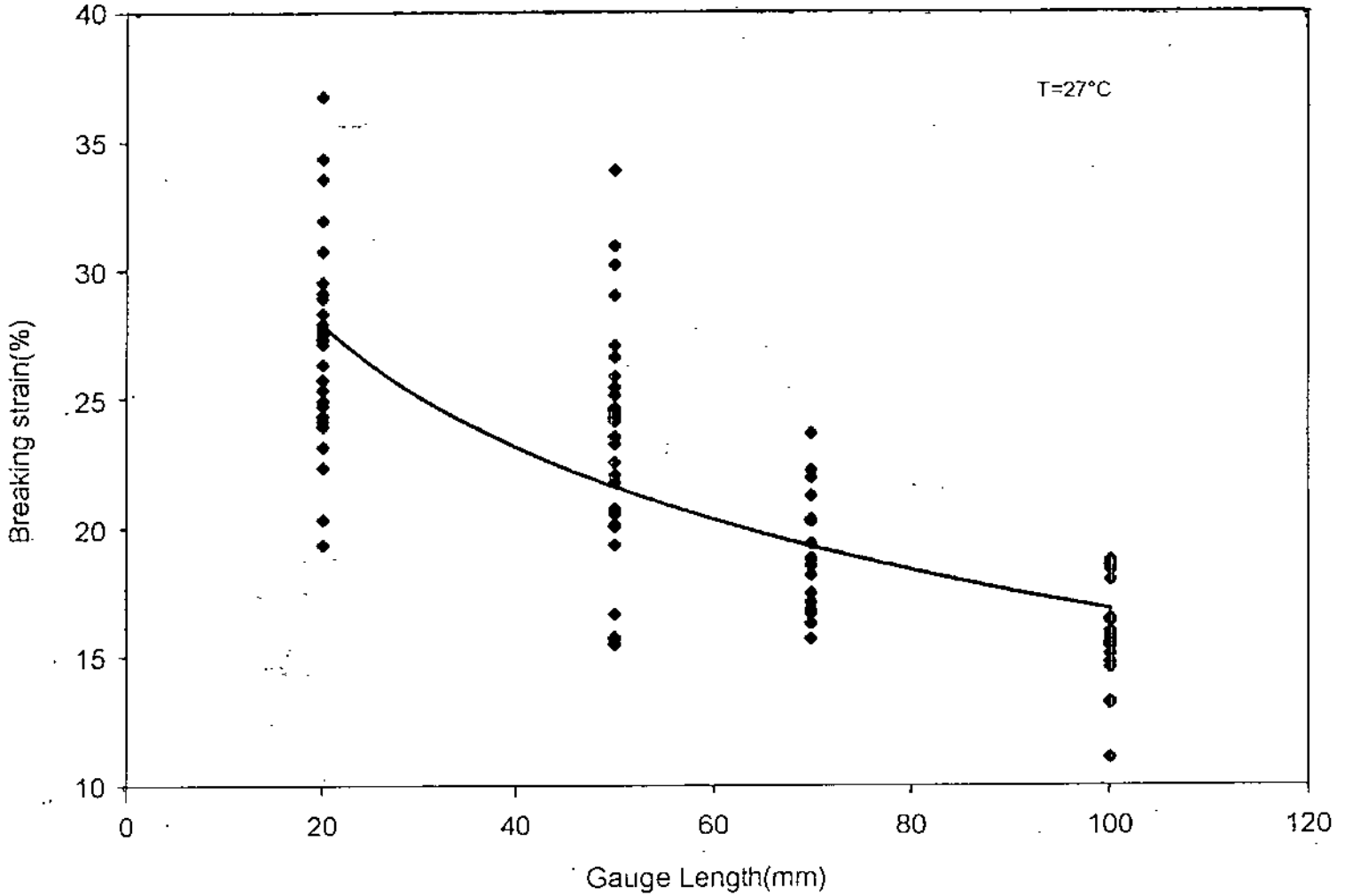


Figure (4-12) Strength versus Gauge Length for carbon fibers



Figure(4-13) Breaking-Strain versus Gauge Length for PP films

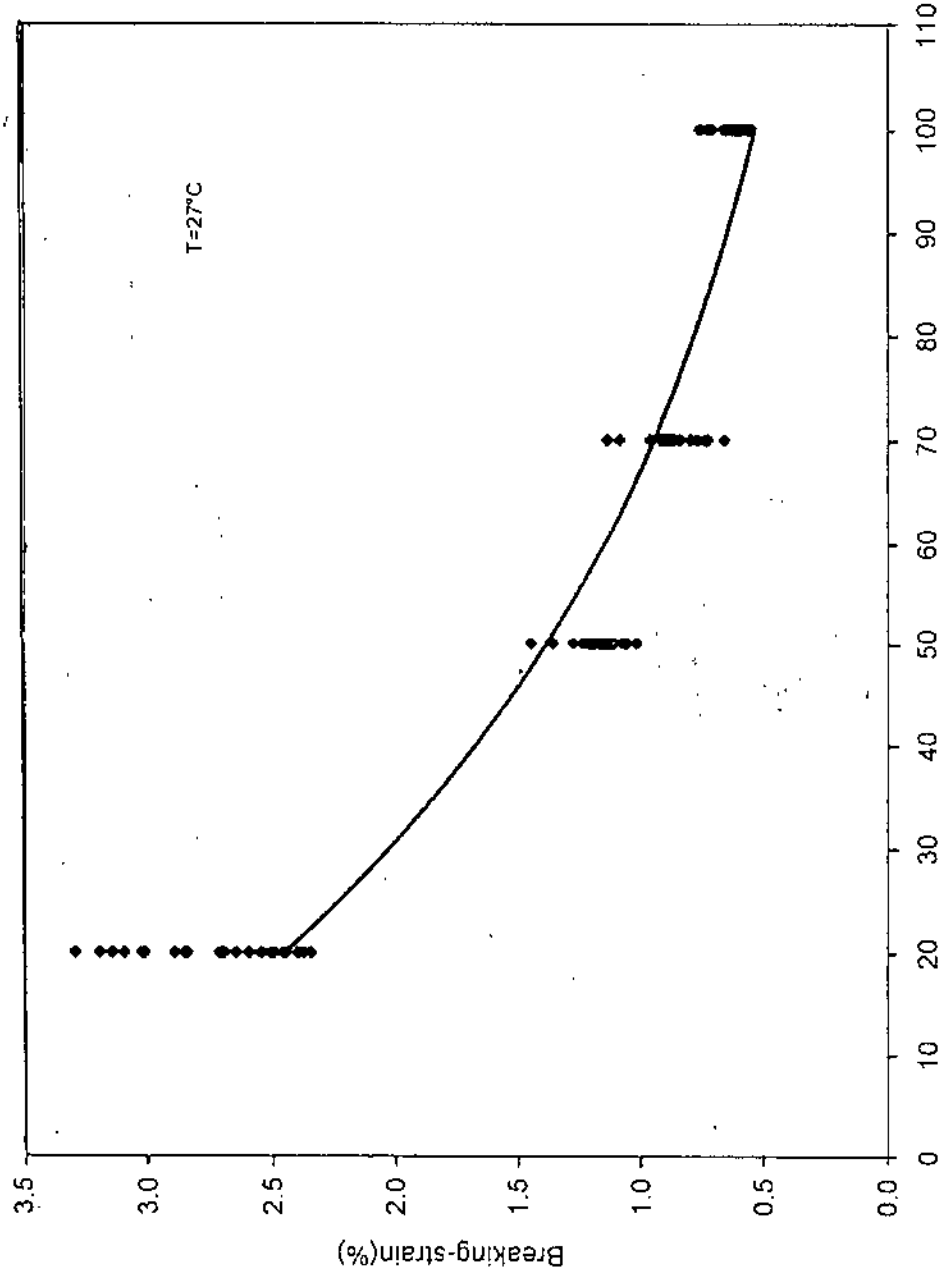
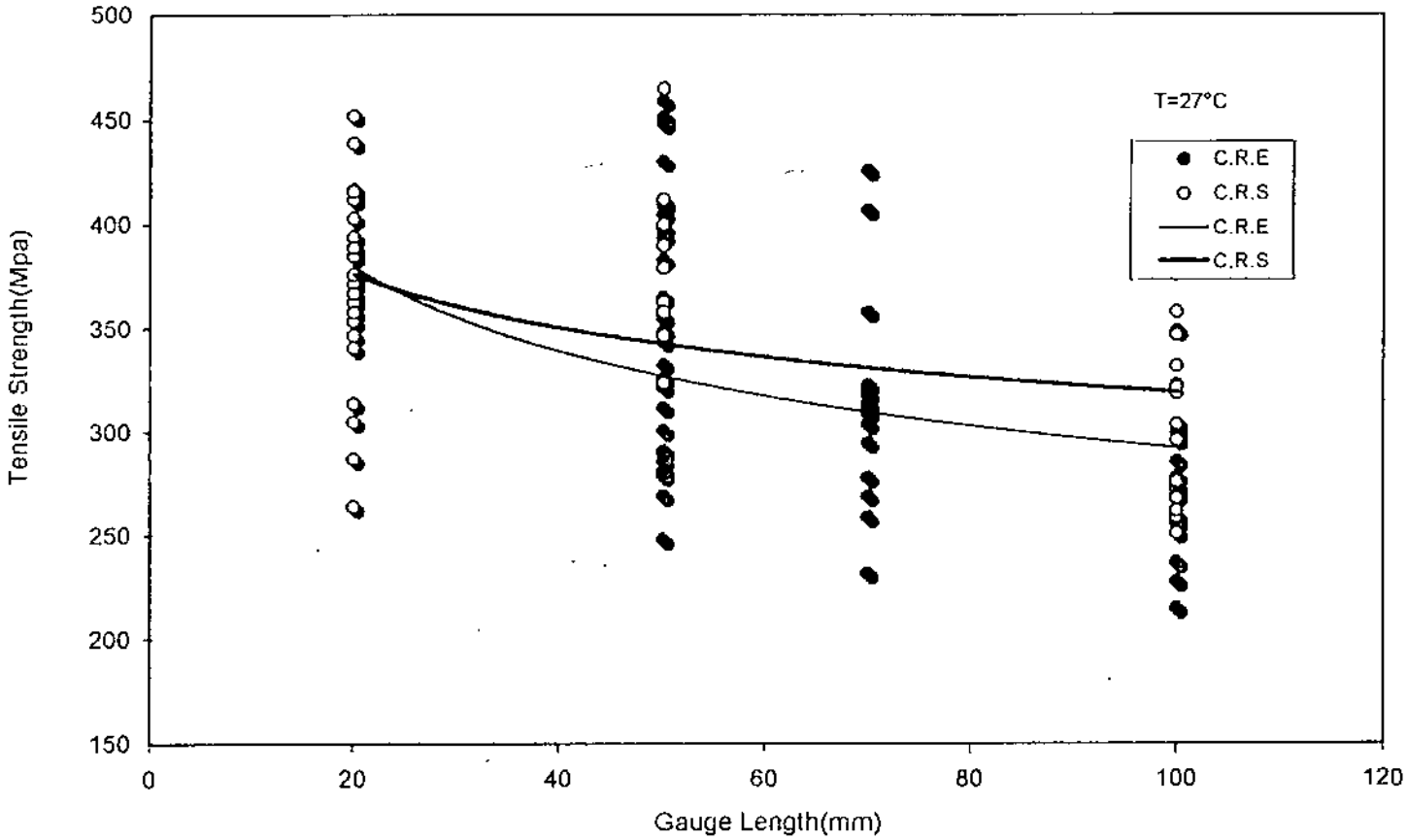


Figure (4-14) Breaking-Strain versus Guage Length for carbon fibers



Figure(4-15) Strength versus gauge length for PP films

C.R.S =constant strain rate
C.R.E =constant rate of extenshion

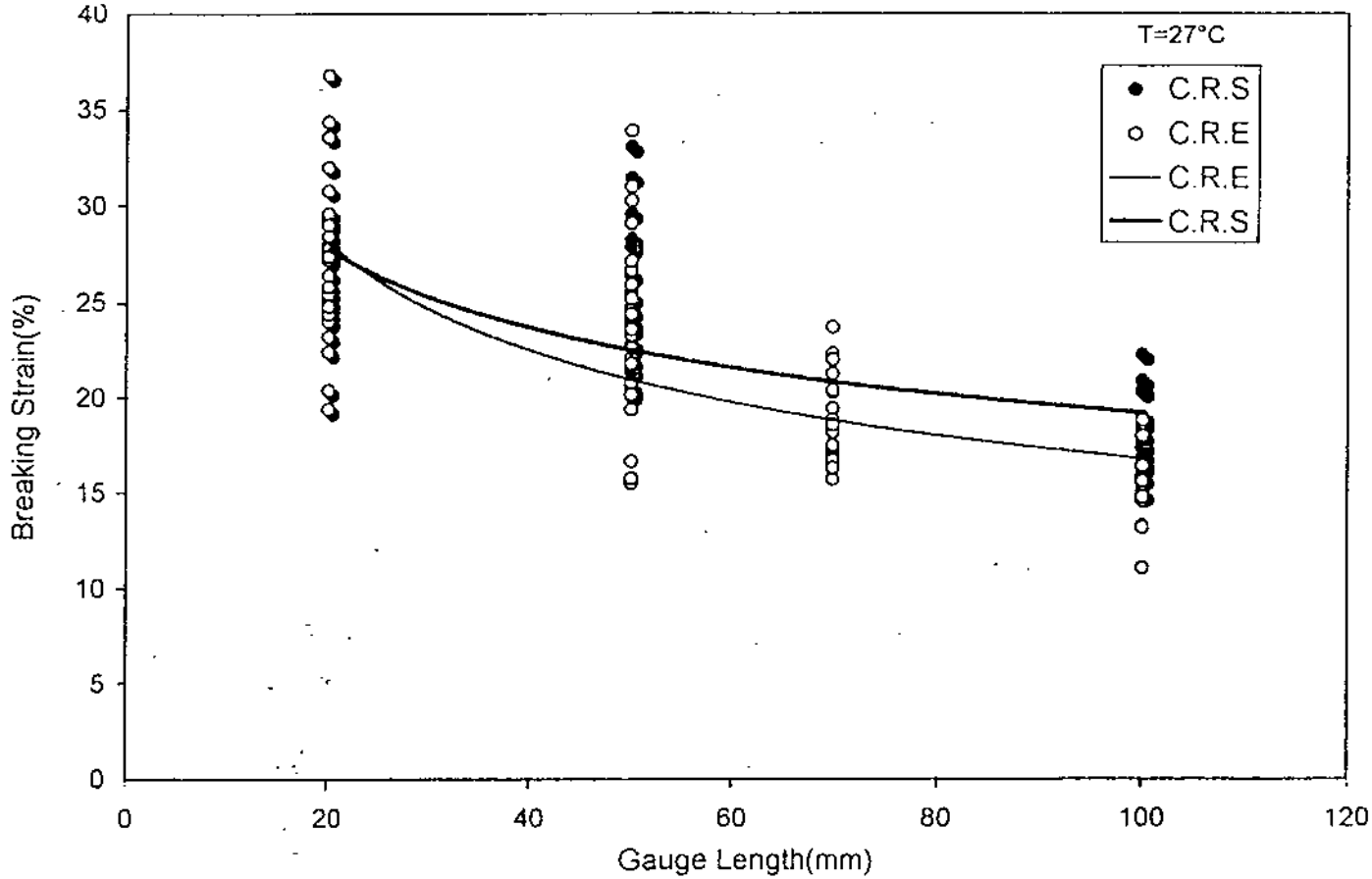
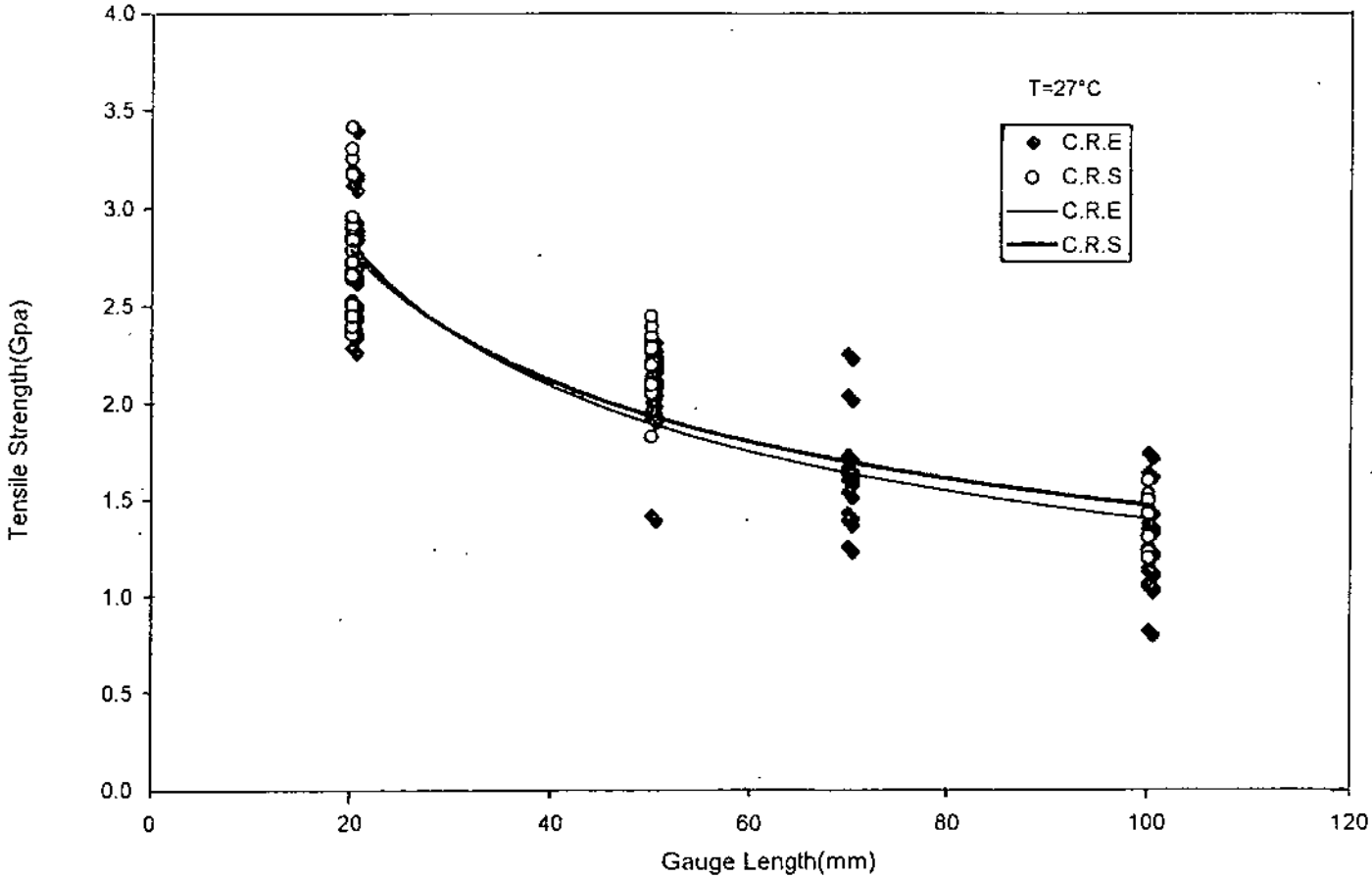


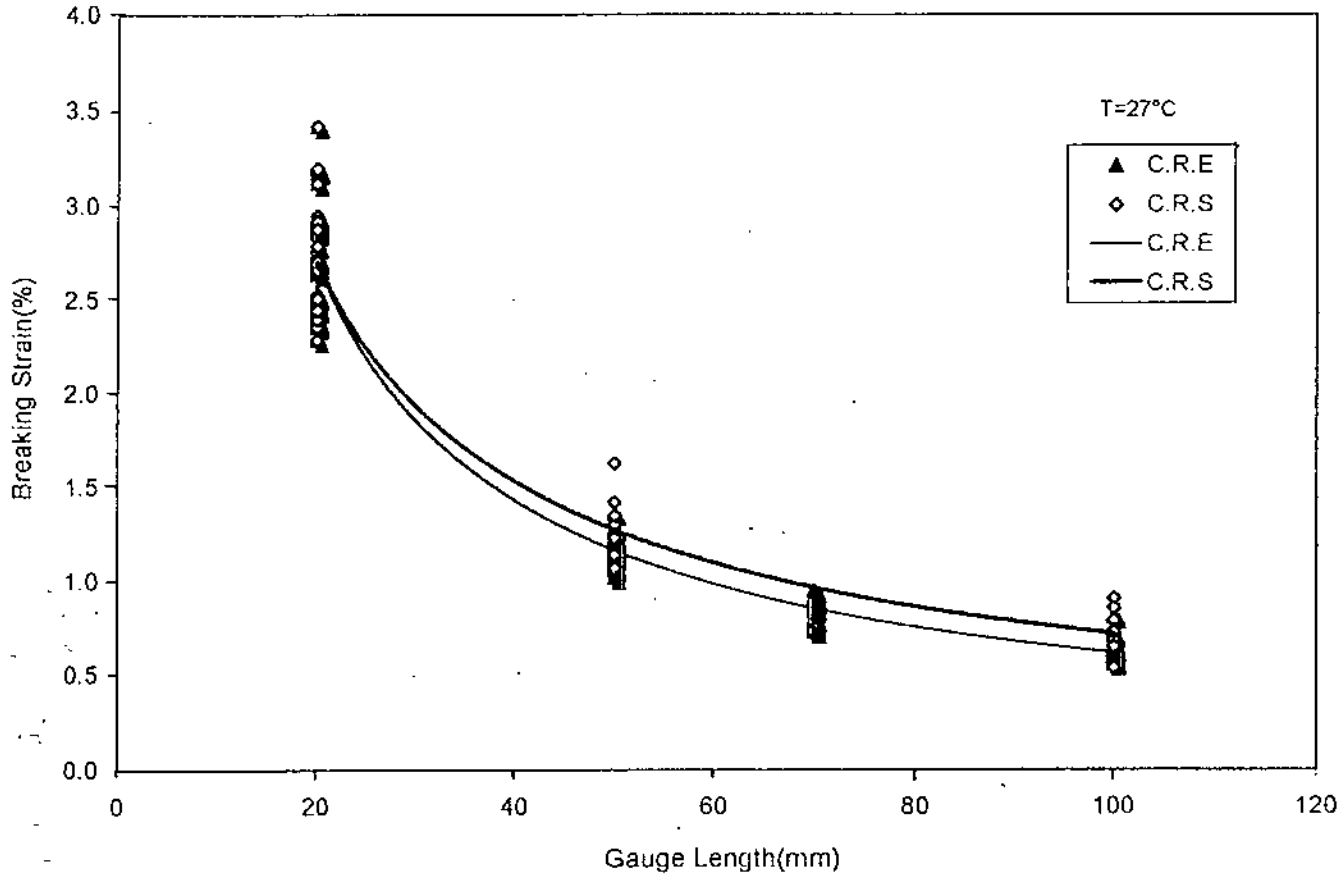
Figure (4-16) Breaking-Strain versus gauge length for PP films

C.R.S= constant strain rate
C.R.E= constant rate of extension



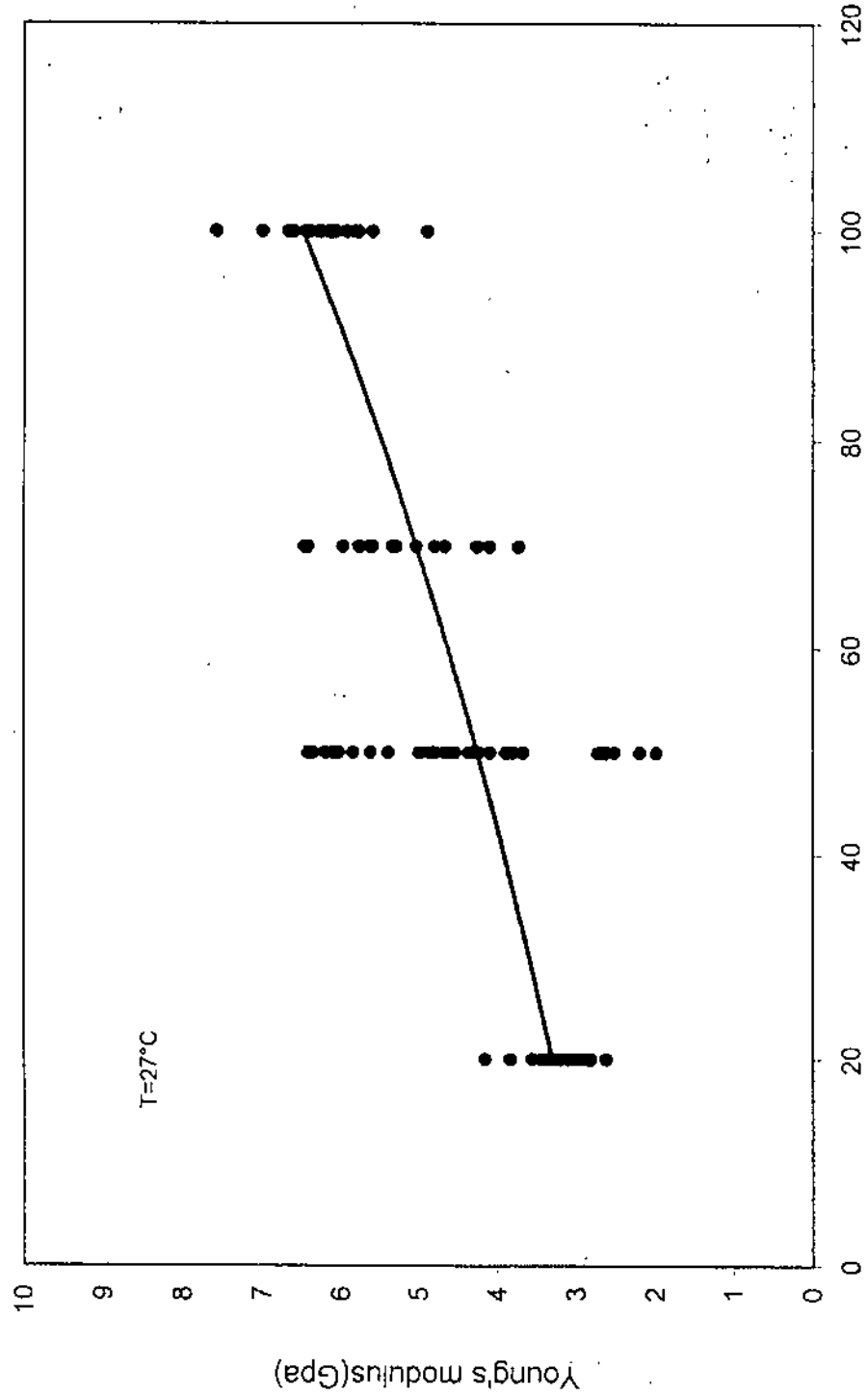
Figure(4-17) Strength versus gauge length for carbon fiber

C.R.S=constant strain rate
C.R.E=constant rate of extension

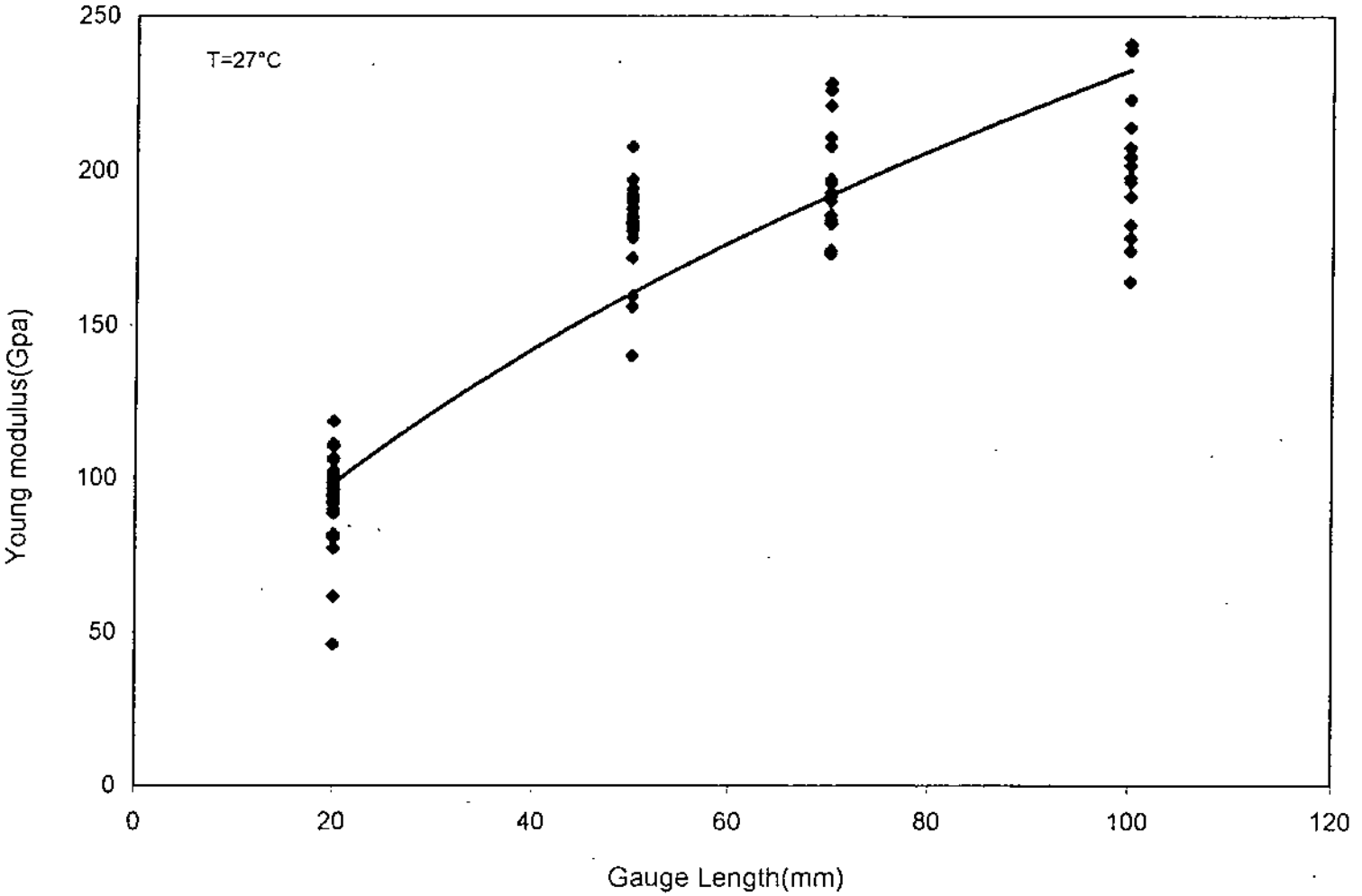


Figure(4-18) Breaking-Strain versus gauge length for carbon fiber

C.R.S=constant strain rate
C.R.E=constant rate of extension



Figure(4-19) Young's modulus versus gauge length for PP films



Figure(4-20) Young's modulus versus gauge length for carbon fiber

4.5 Weibull Model

The study of size effect on the mechanical behavior through the Weibull model is important for investigating the deformation and rupturing processes during loading of a fibrous structure (Pan *et al*.,1997).

Table (4.3) shows the Weibull parameters for strength, and table (4.4) shows the Weibull parameters for breaking strain. It is clear that the Weibull parameters vary with the fiber length. The scale parameter (α) increases as the length decreases, and changes from sample to another (Gurvich and Dibendetto, 1997). Some researchers (Ping and Hsieh, 1997) assume the fiber's breaking strain is a Weibull type variable as well. The fitness of the Weibull model to the experimental data has to be tested statistically to verify whether both strength and breaking strain of fibers obey the Weibull model, the Kolomgorov suitability and goodness-of-fit of performed tests was used (Mann *et al.*, 1974, Bury, 1975) following the procedure outlined below:

1) Rank all experimental data Xi in an ascending sequence.

2) Calculate the sample statistical distribution using (i-0.5) /n, where n is the total sample number.

3) Calculate the theoretical distribution according to equation (2.9), using the estimated Weibull parameters.

4) Compare the corresponding pairs from 2 and 3 points and find the maximum difference between them denoted as d_n.

5) Calculate the critical values of dn: for significant level a=0.05,dnc=$1.36/n^{1/2}$.

6) If $d_n<d_{nc}$, the statistical model is considered a good representation of the data distribution. The computer program for this procedure is written in Gwbasic and given as program II in appendix. The d_n and d_{nc} for both the strength and breaking strain, for the two types of the polymers, are thus calculated and listed in Table (4.3) and Table (4.4).

Table (4-3) Weibull distribution for strength property.

FIBERS	GAUGE LENGTH (mm)	α(GPA)	β	d_{nc}	d_n
PP	20	0.390	10.144	0.258	0.056
	50	0.375	6.936	0.233	0.098
CARBON	20	2.861	9.644	0.258	0.101
FIBERS	50	2.210	26.137	0.304	0.114

We can conclude, by comparing d_n and d_{nc} values in the above tables, that the Weibull model is an adequate representation of the data of both

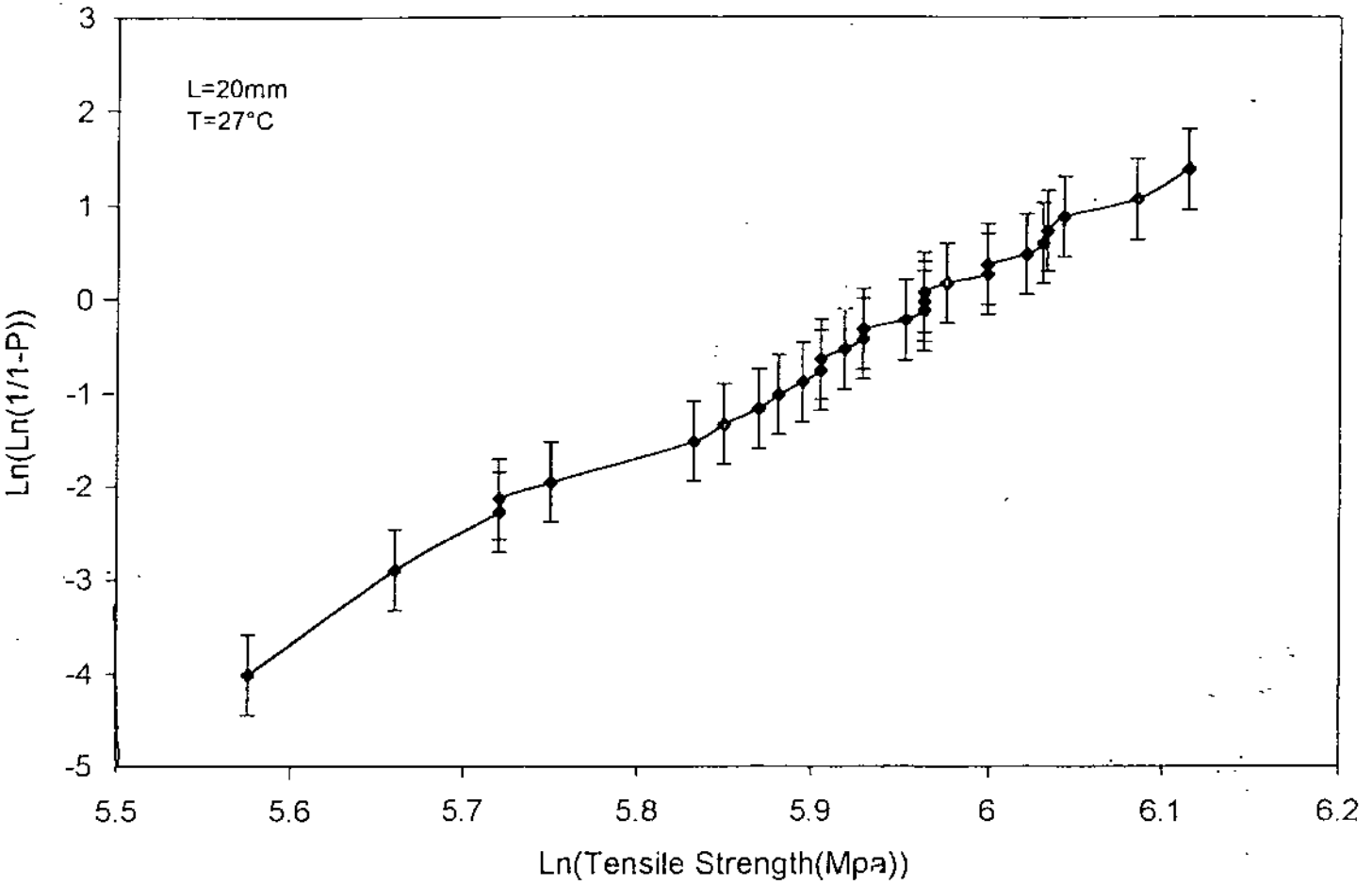
strength and breaking strain. Similar results are reported by other workers (Nakomura and Mikito, 1997).

Table (4-4) Weibull distribution for breaking strain property.

FIBERS	GAUGE LENGTH (mm)	α	β	d_{nc}	d_n
PP	20	28.911	7.062	0.258	0.147
	50	25.098	5.612	0.233	0.079
CARBON FIBERS	20	2.895	10.392	0.258	0.146
	50	1.222	11.388	0.304	0.193

Figs. (4.21) and (4.22) show the relation between Ln (Ln (1-P)) and Ln (σ) for Polypropylene drawn films at 20, 50mm gauge lengths, respectively. Figs.(4.23) and (4.24) show the relation between Ln (Ln (1-P)) and Ln (σ) for Pan-based carbon fibers at 20, 50 mm gauge lengths, respectively, Figs. (4.25) and (4.26) show the relationship between Ln (Ln (1-P)) and Ln(ε) for Polypropylene oriented films and Figs. (4.27) and (4.28) for Pan-based carbon fibers. The relation is not a straight line as predicted by equation (2.11). Similar results are reported from other workers (Gao, 1993; Gurvich and Dibenedetto, 1997). The strong evidence can be deduced that the failure of Polypropylene oriented films and Pan-based carbon fibers is a distribution of failure strength and behavior of breaking strain. It can be

concluded that the Weibull model is adequate to describe the strength and breaking strain for polymeric fibers. Two pictures for Polypropylene films and carbon fibers at different gauge lengths for deformation are illustrated in Figs. (4-29) and (4-30). Large gauge length gives rises to increasing the defects, this means that the strength is lowered while the elongation is increased. A general examination of shown photographs reveals that the Polypropylene films and carbon fibers at large gauge lengths is more ductile than those deformed at relatively small gauge lengths.



Figure(4-21) Probability- Strength curve for PP films Cumulative Weibull distribution

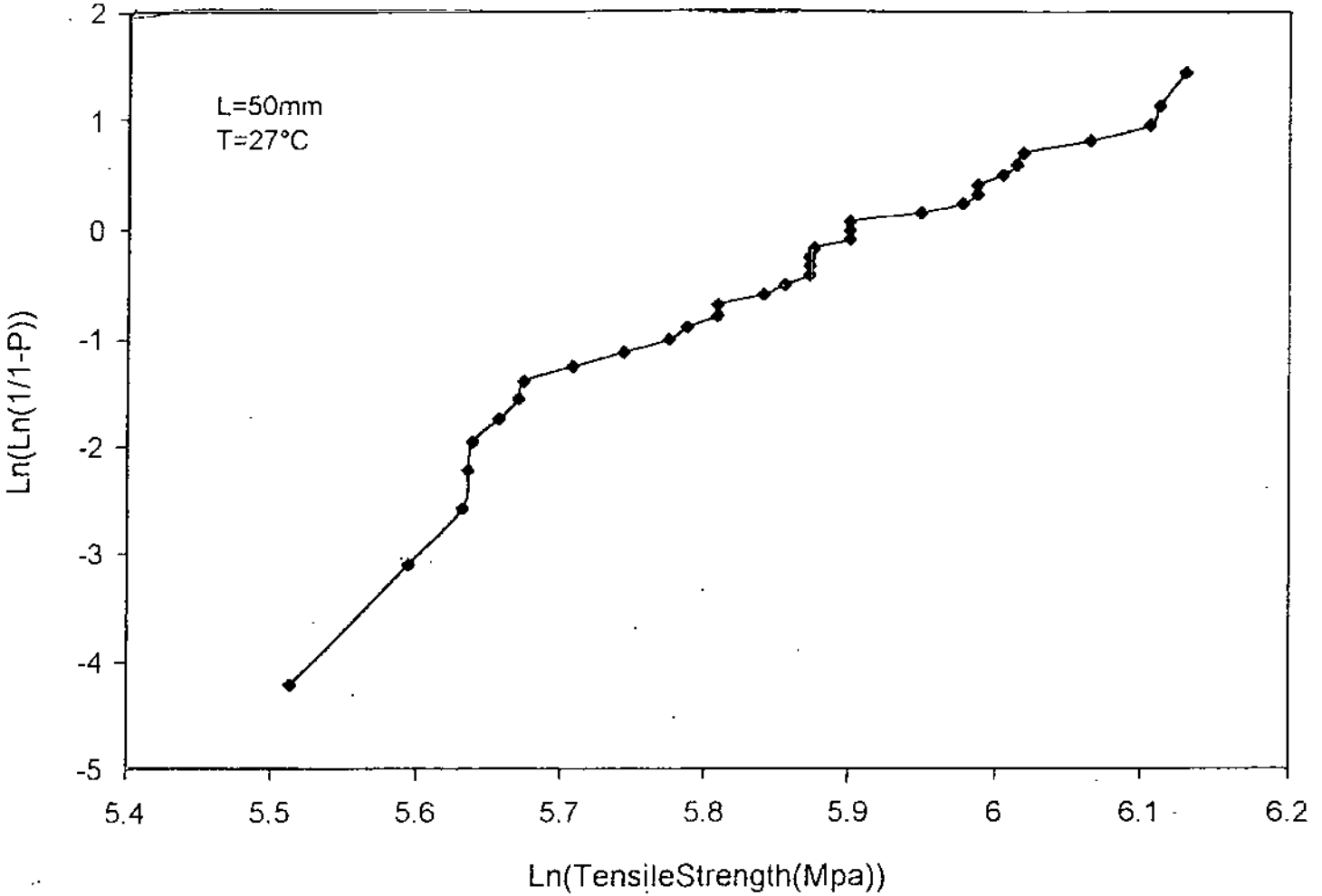


Figure (4-22) Probability-Strength curve for PP films

Cumulative Weibull distribution

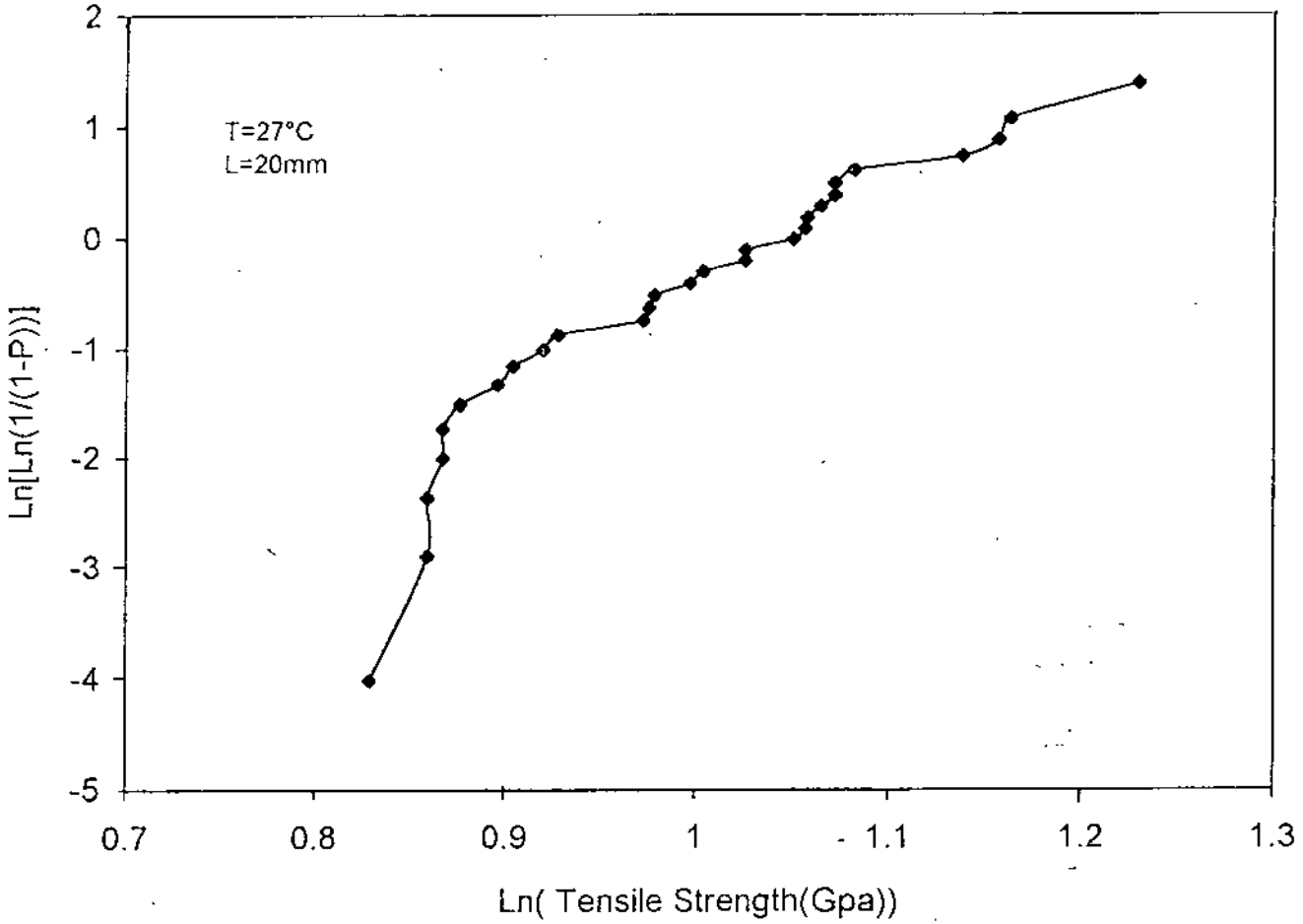


Figure (4-23) Probability-strength curve for carbon fiber Cumulative Weibull distribution

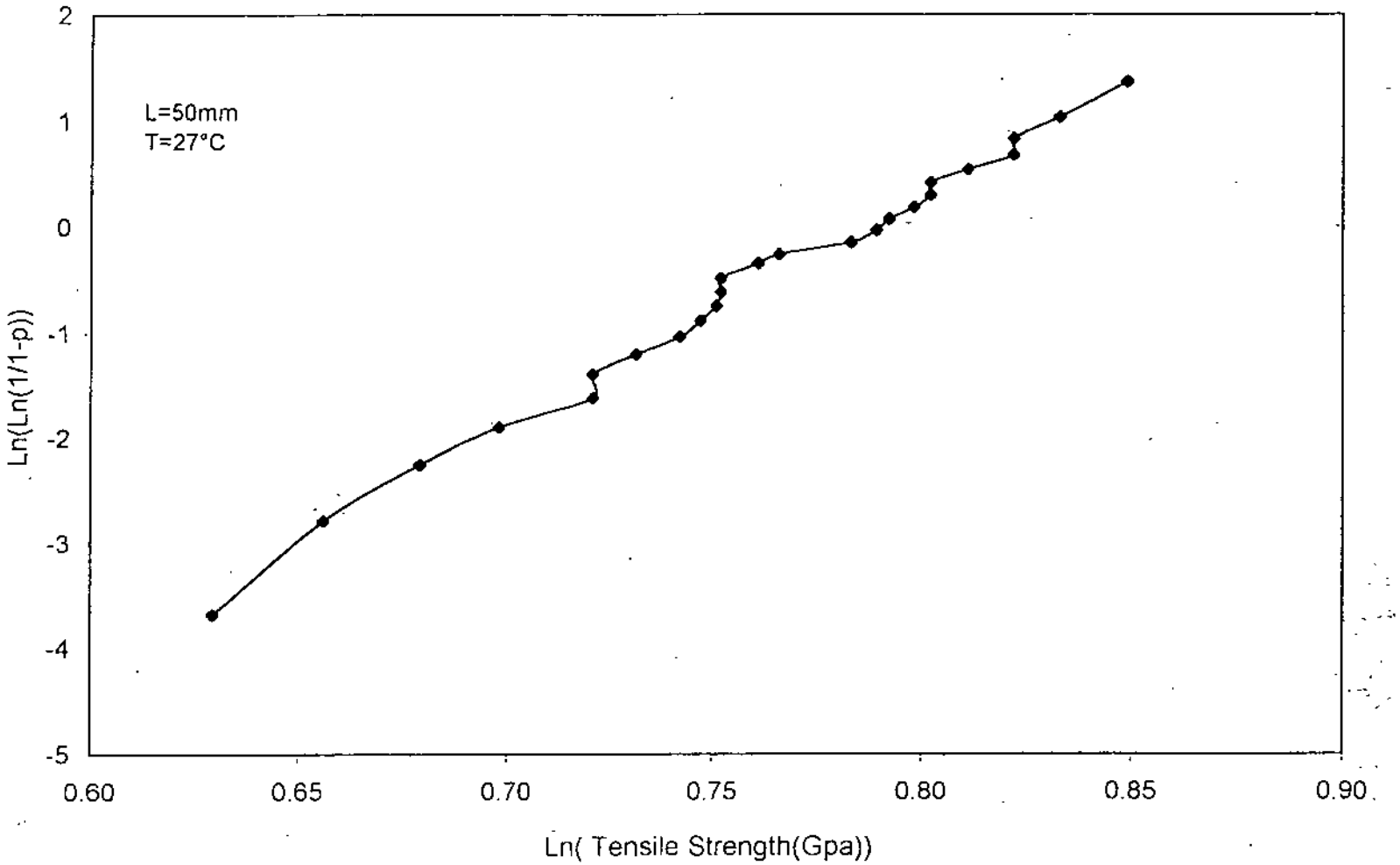


Figure (4-24) Probability-strength curve for carbon fiber Cumulative Weibull distribution

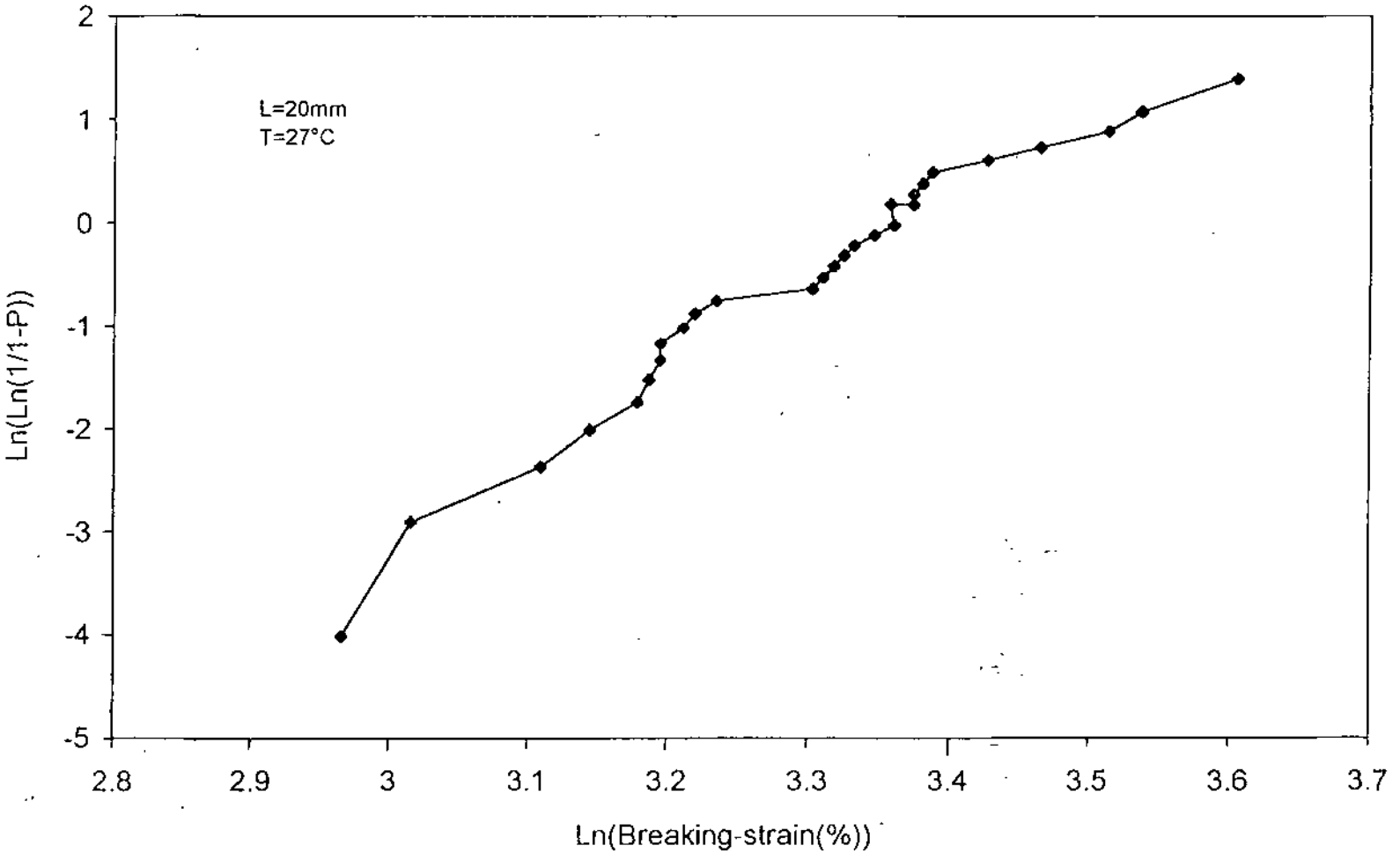


Figure (4-25) Probability- breaking strain curve for PP films Cumulative Weibull distribution

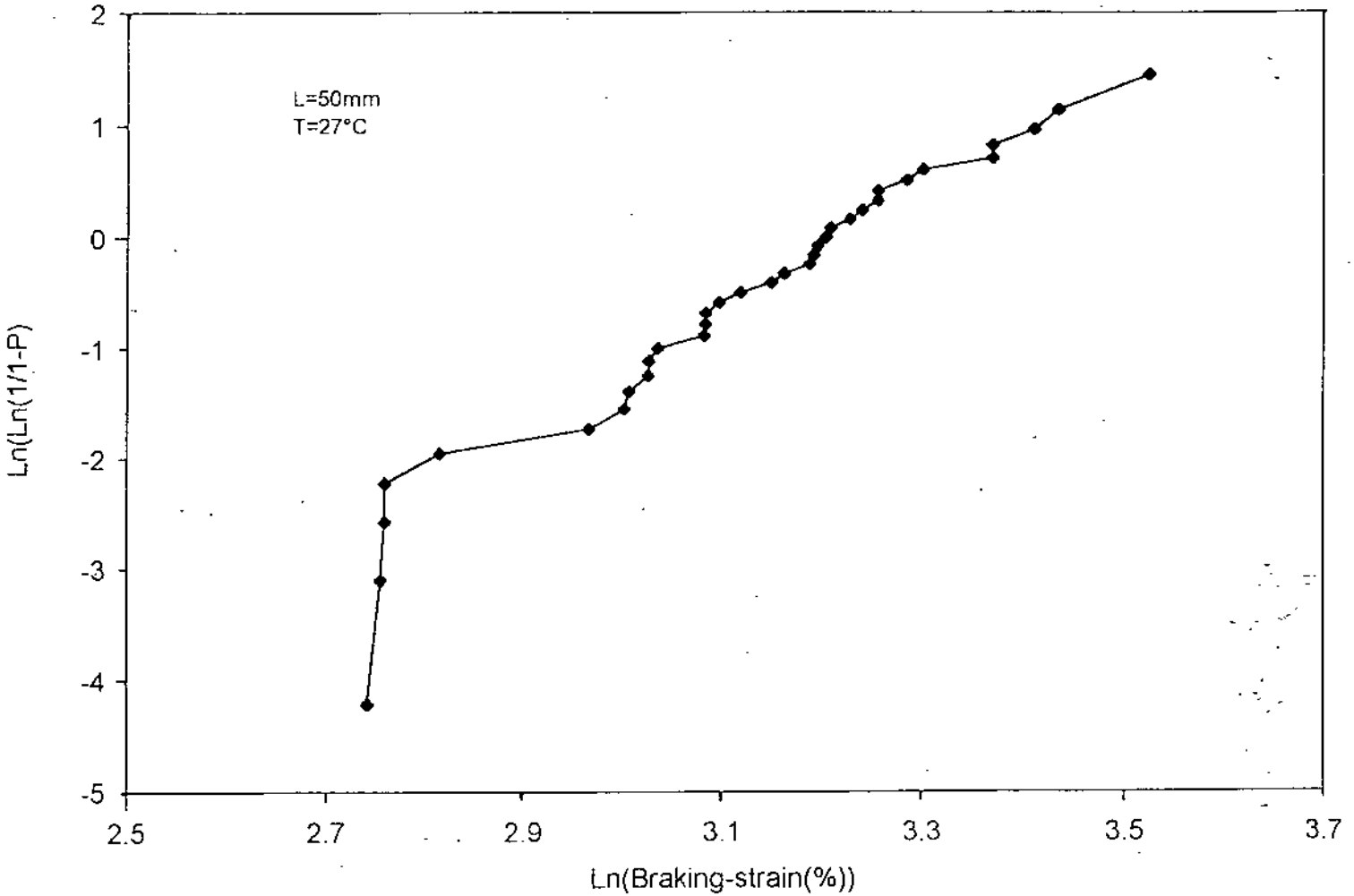


Figure (4-26) Probability-Braking-strain curve for PP films Cumulative Weibull distribution

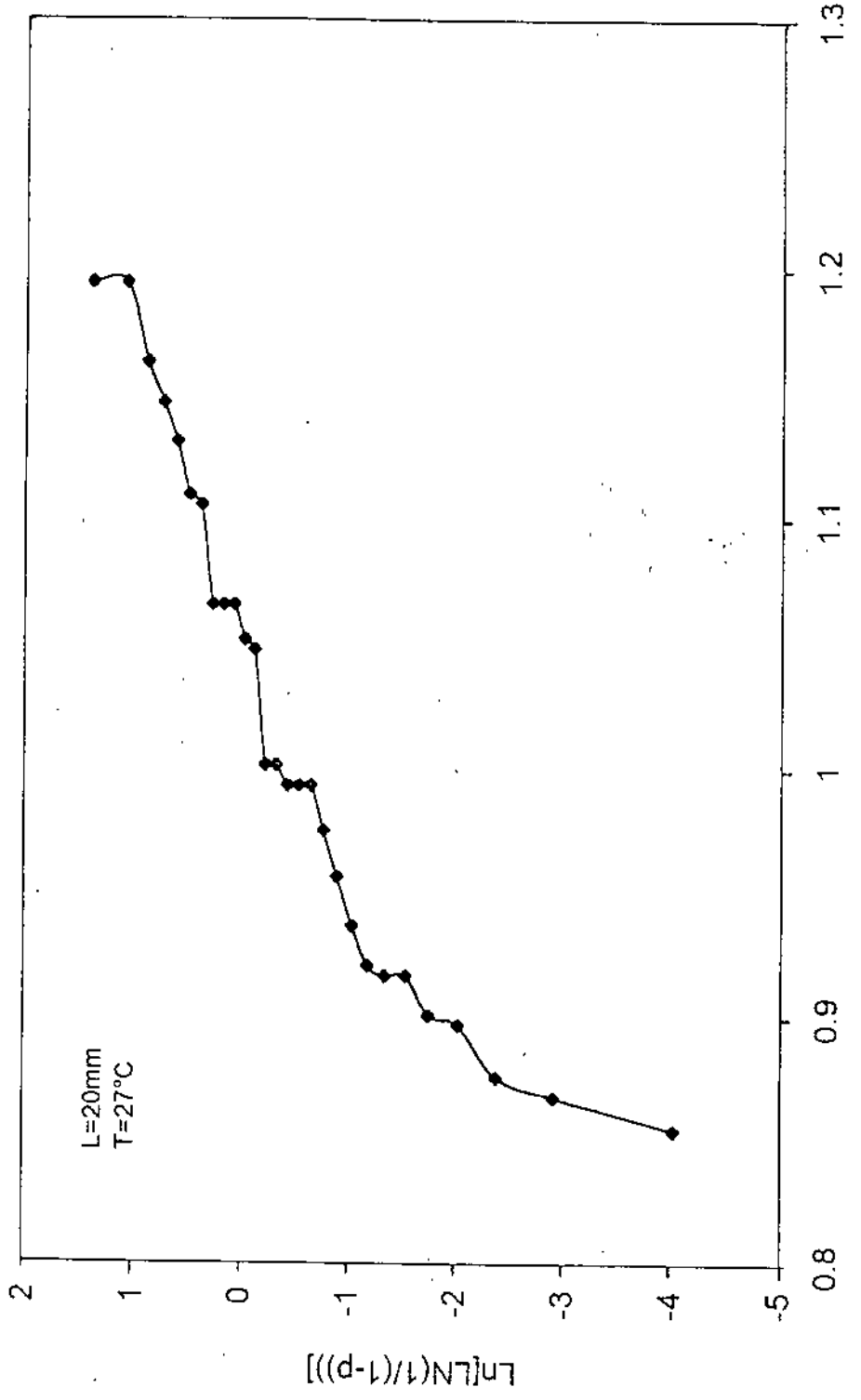


Figure (4-27) Probability-breaking strain curve for carbon fiber Cumulative Weibull distribution

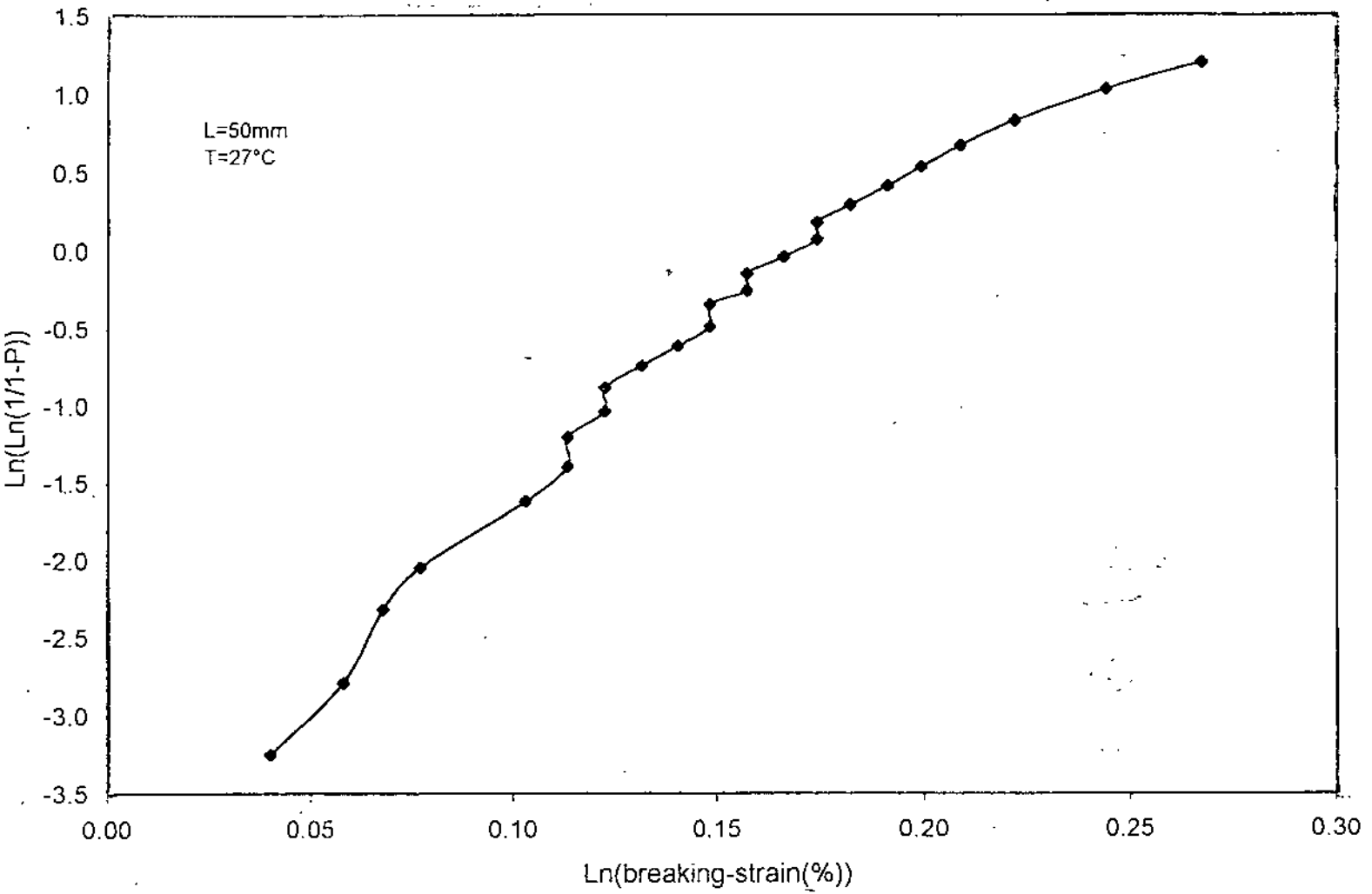


Figure (4-28) Probability-Braking strain curve for carbon fiber Cumulative Weibull Distribution

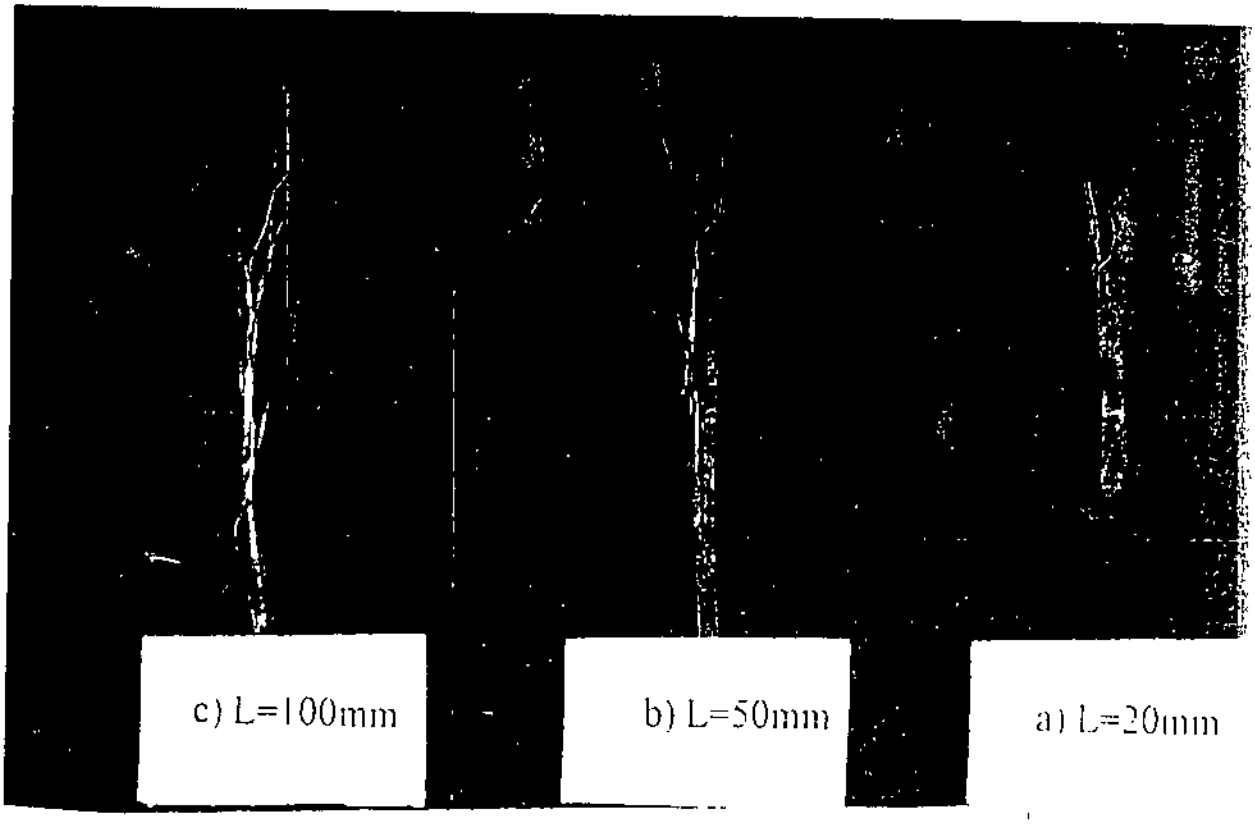


Figure 4.29 Mode of deformation of Polypropylene of different gauge lengths.

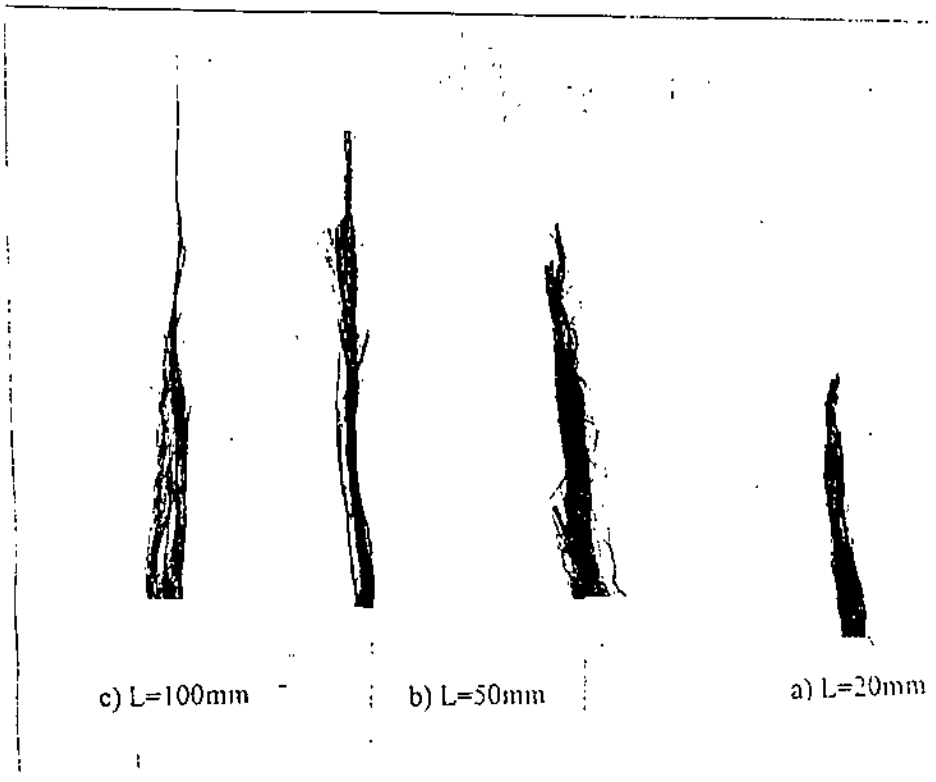


Figure 4.30 Mode of deformation of carbon fibers of different gauge lengths

CHAPTER FIVE

CONCLUSION & FUTURE WORK

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CUNCLUSION & FUTURE WORK

5.1 Conclusion

The research work presented in this thesis covers some of the mechanical properties of both Pan-based carbon fibers and Polypropylene drawn films (fibers) under different gauge lengths. An attempt to fit the Weibull model on the deformation behavior of these oriented polymers is presented.

The following conclusions can be drawn from the obtained results:

1) Modulus, strength and breaking strain of the tested polymers showed size effect dependence.

2) The Weibull model is an important approach for statistical representation of polymers deformation.

3) The Weibull model is adequate to describe the observed deformation of both strength and breaking strain of polymeric fibers.

4) The Weibull parameters of polymer change with the gauge length of the tested fibers and differ from material to another.

5) Photography reveals that the deformation behavior of tested Polypropylene films and carbon fibers at large gauge length are more ductile than those deformed at relatively small gauge length.

5.2 Future Work

Polypropylene films and Pan-based carbon fibers have good mechanical and electrical properties, which enable them to be used successfully in many technological applications. Also the Weibull model is an adequate model to describe the deformation of the polymeric fibers. The future work may cover the followings:

1) Studying the Weibull model for heat treated polymeric fibers.

2) Studying the fitness of Weibull model on deformed polymers composites.

3) Studying the Weibull model under both compressive and the tensile tests for strength and breaking strain.

4) Studying other statistical distributions for strength and breaking strain and compare them with the Weibull distribution.

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```
10 REM                          Program # (1)
20 REM THIS PROGRAM IS TO CALCULATE THE PARAMETERS
        OF WEIBULL DISTRIBUTION FOR
30 REM THE EXPERIMENTAL DATA BY USING THE BISECTION METOD.
40 INPUT N
50 REM N IS THE NUMBER OF THE VALUE OF EXPERIMENTAL DATA.
60 REM X(I) IS THE VALUES OF STRENGTH OR BREAKING STRAIN.
70 DIM X(40)
80 FOR I=1 TO N
90 INPUT X(I)
100 NEXT I
110 FOR F=1 TO 98
120 A1=0:A2=0:A3=0
130 FOR J= 1 TO N
140 A1=A1+LOG(X(J))
150 A2 =A2+X(J)^F
160 A3=A3+(X(J)^F)*LOG(X(J))
170 NEXT J
180 G=N/F+A1-(N/A2)*A3
190 IF G<0 THEN GOTO 210
200 NEXT F
210 M=F
220 S=F-1
230 P=S+(M-S)/2
240 A1=0:A2=0:A3=0
250 FOR J=1 TO N
260 A1=A1+LOG(X(J))
270 A2=A2+X(J)^P
280 A3=A3+(X(J)^P)*LOG(X(J))
290 NEXT J
300 G=N/P+A1-(N/A2)*A3
310 IF ABS(G)<.01 GOTO 370
320 IF G>0 THEN GOTO 350
330 M=P
340 GOTO 230
350 S=P
360 GOTO 230
370 MLA=P
380 S1=0
390 FOR H=1 TO N
400 S1=S1+X(H)^MLA
410 NEXT H
420 MLB =(S1/N)^(1/MLA)
430 REM WHERE a IS THE SHAPE PARAMETER AND
      b IS THE SCALE PARAMETER.
440 PRINT"a=";MLA
450 END
```

```
10 REM                PROGRAM # (2)
20 REM  THIS PROGRAM IS TO TEST THE FITNESS OF THE
        WEIBULL MODULL FOR EXPEREMENTAL DATA STSTISTICALLY
30 REM  FOR BOTH STRENGTH AND BREAKING STRAIN.
40 REM WHERE N IS THE NUMBER OF  EXPERIMENAL  DATA
50 REM WHERE A IS THE SHAPE PARAMETER
        & B IS THE SCALE PARAMETER.
60 INPUT N,A,B
70 DIM T(N)
80 DIM S(N)
90 DIM P(N)
100 DIM D(N)
110 FOR I= 1 TO N
120 REM T(I) ARE THE VALUES OF EXPERIMENTAL
130 REM DATA FOR BOTH  STRENGTH & BREAKING STRAIN.
140 INPUT T(I)
150 NEXT I
160 FOR I=1 TO N-1
170 FOR J =1 TO N-I
180 IF T(I)<T(I+J) THEN GOTO 220
190 W=T(I)
200 T(I)= T(I+J)
210 T(I+J)=W
220 NEXT J
230 NEXT I
240 FOR I= 1 TO N
250 P(I)=1-(EXP(-(T(I)/A)^B))
260 PRINT"p("I")=";P(I)
270 NEXT I
280 FOR I=1 TO N
290 S(I)=(I-.5)/N
300 PRINT"s("I")=";S(I)
310 NEXT I
320 FOR I=1 TO N
330 D(I)=P(I)-S(I)
340 PRINT"d("I")=";D(I)
350 NEXT I
360 H=0
370 FOR I= 1 TO N
380 IF ABS(H)>ABS(D(I)) GOTO 400
390 H=D(I)
400 NEXT I
410 PRINT"h=";ABS(H)
420 END
```

الملخص

دراسة التشوه اللدن لمبلمرات خيطيه بواسطة نموذج ويبل

إعداد

غازي محمود جوهر

إشراف

الأستاذ الدكتور عواد الزحلف

تتضمن هذه الاطروحه دراسة التشوه اللدن لكل من أشرطة البـــولي بروبلـــين و خيـــوط الكربون من خلال نموذج ويبل لاطوال مختلفة لعينات البحث .وقد تم تطبيق التحليل الإحصـــائي لكل من المقاومة و انفعال القطع باستخدام القيم العملية لفحص مطابقة نموذج ويبل لهذه الألياف.و بالإضافة إلى ذلك تمت دراسة الخصائص الميكانيكية لاشرطة البولي بروبلين المتجهـــة الجزئيـــات و خيوط الكربون ,مثل المقاومة و انفعال القطع و معامل المرونة و علاقتـــهم بـــالطول و معـــدل الانفعال. وقد وجد أن زيادة الطول يؤدي إلى تقليل كل من المقاومة و انفعال القطع و زيـــادة في معامل المرونة. و يعزى ذلك إلى زيادة نسبة العيوب عند زيادة الطول . و تستخلص الدراســـة إن نموذج ويبل مناسب إحصائيا لوصف كل من المقاومة و انفعال القطع , وكذلك يصلح لدراســـة التشوه اللدن لهذه المبلمرات الخيطية.