6 Mechanical Behavior

6.1 General Deformation Behavior

The mechanical characteristics of plastics are more sensitive than those of metals to external influences such as temperature, time, load magnitude and type, UV-radiation, and chemicals. In general, the elastic modulus of plastics is approximately two orders of magnitude, and the strength one order of magnitude lower than that of metals. Obviously, the very good strength to weight ratio of plastics and polymer composites makers them a plausible alternative to metals for lightweight constructions. Due to the ability of tailoring the mechanical properties of plastics, design engineers are required to have a full understanding of the processing-dependent material properties to adequately utilize the inherent benefits of plastics.

The basic formulas for design calculations were developed for linear stress-strain behavior. With this assumption of a linear stress-strain response, parts of simple geometry can be numerically determined. If this assumption cannot be accepted, one has to use the more advanced models that account for the specific response on the excitation in polymers. The type and magnitude of the response of polymeric materials to externally applied forces or deformations determine their mechanical behavior. For plastic materials, the deformation reactions are highly coupled to the material temperature as well as the rate and magnitude of the applied load.

One special behavior of plastics is that the macromolecules do not immediately deform when a load is applied. Rather, the individual molecular chains are inclined to distribute the stress to an equilibrium value via physical relocation. The response of the individual chains to an applied load via molecular motion is statistically irregular and thus impossible to exactly predict beforehand. However, the rates of these relocation processes are known to be dependent on the magnitude of the applied load. The physical or chemical bonds (bulky side-groups, hindered rotation of the main chain, etc.) also influence the mobility of the chain segments. The material temperature also helps to facilitate the molecular relocation process by increasing both the molecular vibrational energy and the total free volume.

Aside from the material specific parameters (molecular weight, degree of branching or cross-linking, chain mobility) and the environment parameters (humidity, chemicals, temperature, rate of loading, as well as the type and magnitude of the applied stress), it is known that the thermomechanical history can often influence the deformation behavior. Residual stress, molecular or filler orientation, degree of crystallinity, etc. can become noticeable when the material is subjected to subsequent loading conditions such as an elevated temperature. Additionally, these individual influences can chance during the service life of the part. In practice, time and temperature are the most influencing factors on the deformation behavior.

Plastics will exhibit rigid and brittle behavior if the duration of the applied load is relatively short in comparison to the time of the molecules require to relocate. If the molecular chains have sufficient time for adjustment to an equilibrium state when subjected to loading, plastics will become ductile and tough. Therefore, the same plastic part may be either brittle or tough within the same application, depending on both the temperature and the loading rate.

Brittle behavior of a plastic is the result of the inability of the molecules to relocate fast enough when the specimen is subjected to either a high rate of loading or deformation at low temperature. Contrary to this response, plastics exhibit a **tough** behavior at elevated temperatures or slow loading rates. Due to the network structure, the behavior of thermoset materials tends to be more brittle at ambient temperatures. Considering the two major types of thermoplastic materials, semi-crystalline materials are generally tougher than amorphous materials. For example, amorphous polymers such as polystyrene (PS) and styrene-acrylonitrile copolymer (SAN) are particularly brittle. Elastomers are classified as a tough polymeric material due to their rubber-like nature. A general overview of the deformation behavior of plastics is shown in Fig. 6.1.

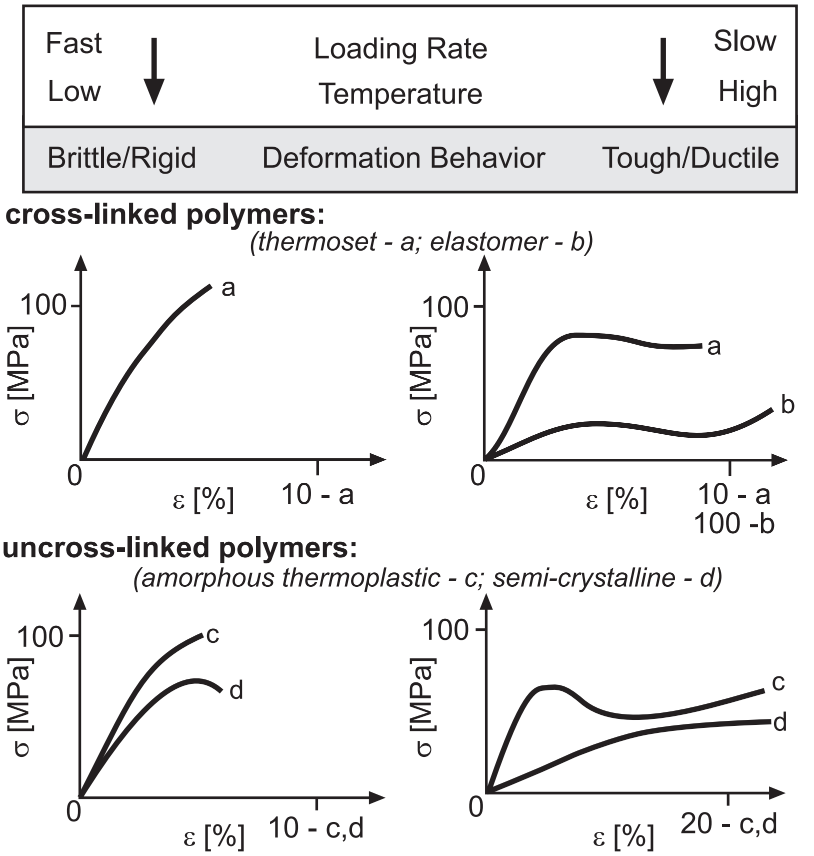

Fig. 6.1: Mechanical behavior of thermosets, elastomers, and both amorphous and semi-crystalline thermoplastics

The molecular and supermolecular structure[44] directly determines the material properties. The processing conditions, chemical structure, external loading (heat and aggressive chemicals) as well as the time-dependent aging characteristics can influence the material properties and therefore the quality of the finished part as illustrated in Fig. 6.2. The design of plastic parts must take into consideration the type

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[44] **Structure** refers to the physical and chemical organization of the individual macromolecules as well as their supermolecular arrangement. In contrast, **morphology** refers to the phenomenological appearance of the polymeric superstructure.

of loading, part geometry, material selection, processing method, conditioning, as well as all of the interactions between these variables.

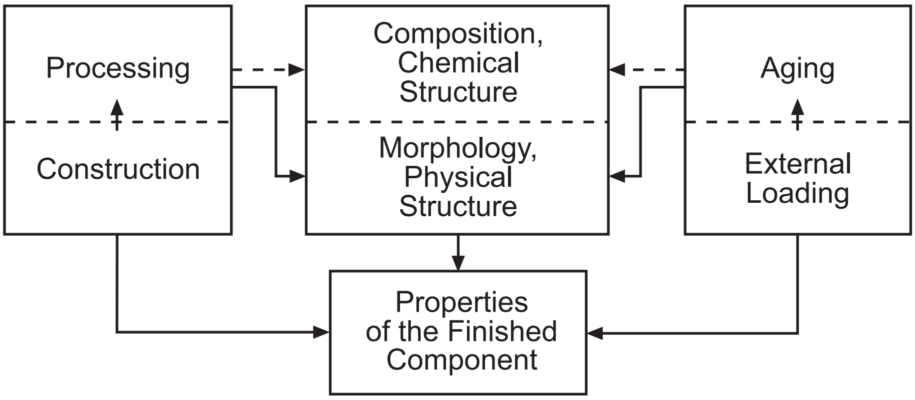


Fig. 6.2: Factors that influence the design cycle for plastic parts (courtesy Retting)

6.1.1 Mechanical Properties

The mechanical response of polymeric materials to an applied load can be divided between fully reversible and irreversible deformation behavior. Reversible deformation is described as the ability of a specimen to geometrically recover the original shape after a load has been removed. In comparison, irreversible deformation describes the accumulation of permanent damage within the specimen such that the original geometry cannot be recovered. Furthermore, the time-dependent deformation response can be used to subdivide reversible and irreversible deformation responses into the following classifications:

- instantaneous elastic deformation (reversible)
- time-dependent viscoelastic deformation or relaxation (reversible)
- and time-dependent viscous deformation (irreversible)

The three classifications of deformation are schematically illustrated in Fig. 6.3. The instantaneous modification of the atomic distances and the distortion of the valence angles between fixed chemical bonds both characterize purely **elastic** deformation. In contrast the hindered movement of the molecular chains in relation to one another describes the **viscous** or **plastic** response. The **viscoelastic** or **relaxation deformation** is positioned between the elastic and viscous deformations in which the individual molecules or molecular groups require a certain time interval to deform via molecular relocation when subjected to an external stress.

Therefore, plastics react to an applied load with a certain time delay that is characterized by the material-dependent **relaxation time** τ. The relaxation is described, in general, as the time-dependent tendency of the material to reach an equilibrium state when externally loaded. Since the load leads to a definite strain and a definite stress, the ability to equilibrate the load is defined as either strain or stress relaxation, respectively. The relaxation time can be obtained from the viscous and elastic response by the equation:

$$\tau = \frac{\eta_{rel}}{E_{rel}} \tag{6.1}$$

Realistic values for the relaxation time for polymeric materials range between 1 and 100 hours.

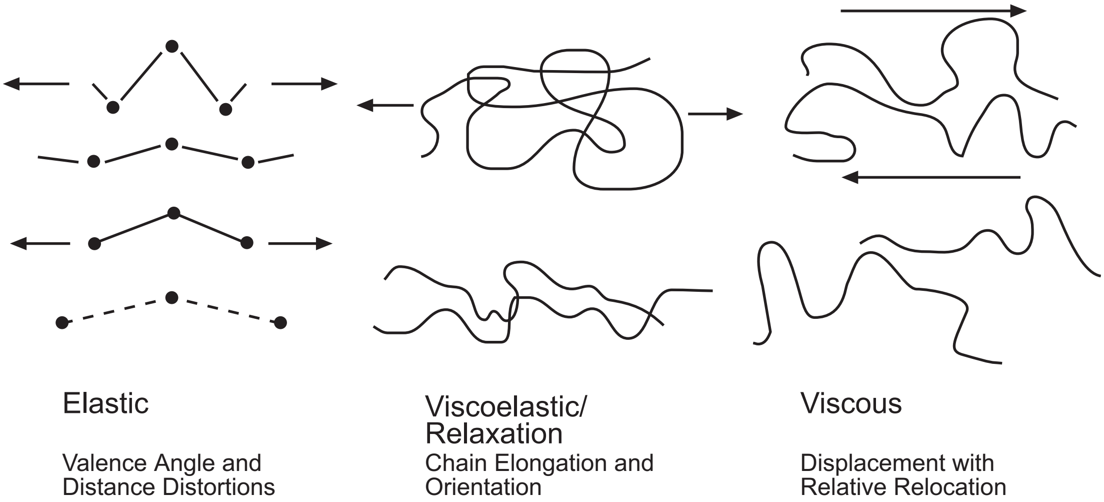
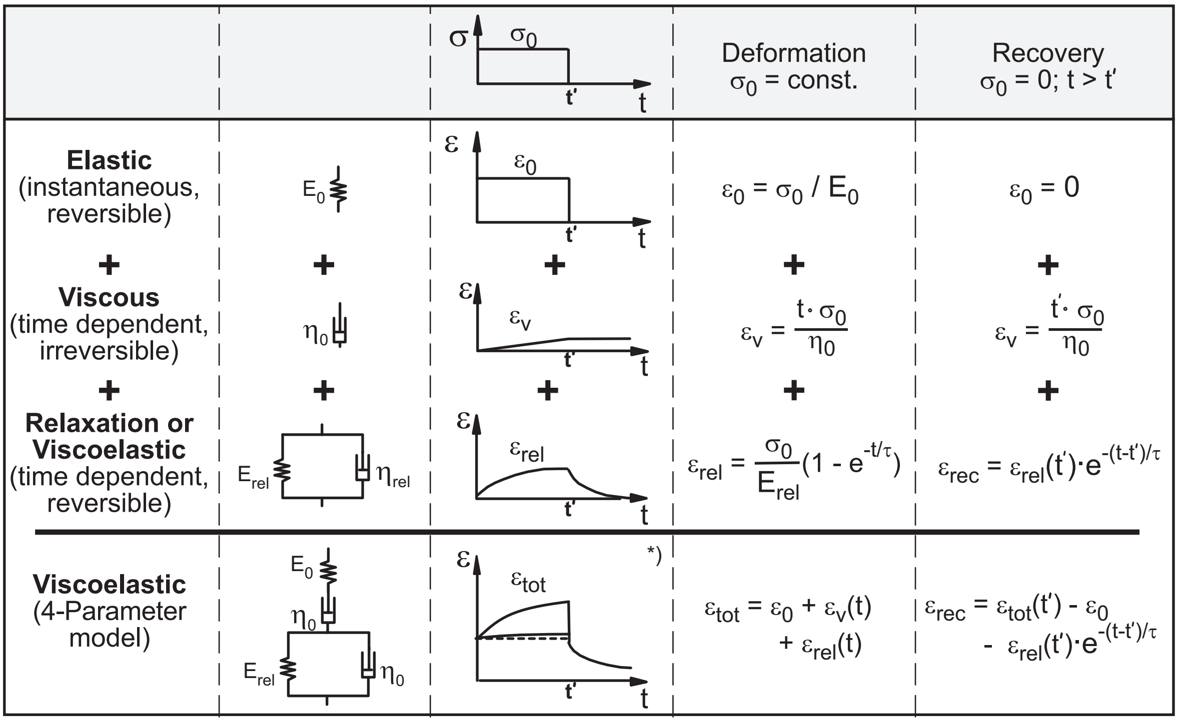


Fig. 6.3: Molecular description of the elastic, viscoelastic/relaxation and viscous deformation of polymers [29]



			Deformation σ_0 = const.	Recovery $\sigma_0 = 0;\ t > t'$
Elastic (instantaneous, reversible)			$\varepsilon_0 = \sigma_0 / E_0$	$\varepsilon_0 = 0$
+	+	+	+	+
Viscous (time dependent, irreversible)			$\varepsilon_v = \frac{t \cdot \sigma_0}{\eta_0}$	$\varepsilon_v = \frac{t' \cdot \sigma_0}{\eta_0}$
+	+	+	+	+
Relaxation or Viscoelastic (time dependent, reversible)			$\varepsilon_{rel} = \frac{\sigma_0}{E_{rel}}(1 - e^{-t/\tau})$	$\varepsilon_{rec} = \varepsilon_{rel}(t') \cdot e^{-(t-t')/\tau}$
Viscoelastic (4-Parameter model)			$\varepsilon_{tot} = \varepsilon_0 + \varepsilon_v(t) + \varepsilon_{rel}(t)$	$\varepsilon_{rec} = \varepsilon_{tot}(t') - \varepsilon_0 - \varepsilon_{rel}(t') \cdot e^{-(t-t')/\tau}$

*) see also Fig. 2.4

Fig. 6.4: Description of the deformation behavior of polymeric materials via a 4-parameter model
Index: 0 = elastic, v = viscous, rel = relaxation, rec = recovery

When a polymeric specimen is deformed to a constant strain, the first response is the development of a relatively large localized stress. The stress magnitude begins to diminish over a time interval as a result of the slow relocation of the individual molecular chains into the preferred coiled equilibrium state. This type of time-dependent deformation is known as **stress relaxation**. In comparison, the graduate term **strain relaxation** of polymeric materials is used to describe the deformation of a polymeric specimen subjected to a constant stress. The time-dependent deformation of a specimen is also known as **retardation** or **creep** as illustrated in Fig. 6.4.

Subjected to a constant stress, the viscoelastic response can be determined from the superposition of the time-dependent reversible (elastic) and irreversible (viscous) deformation. A large proportion of the total deformation can be elastically retrieved after the load is removed if damage does not accumulate. The purely elastic contribution to the total deformation can be recovered immediately, however the relaxation-dependent contribution is time-dependent.

The formal distinction between the load-dependent stress relaxation (ε = const.) and creep (σ = const.) has little importance for practical applications. This is true because in reality there is no single time value but time- and temperature-dependent relaxation- and retardation-spectra are present, which are rarely not available due to the difficulty in obtaining consistent and accurate experimental results. However, there are also hardly any reliable data for the simplified single time values on hand, available for common polymers. For example, even to maintain a constant level of strain within a test specimen by controlling the stress level would theoretically require a perfectly rigid clamping device. In comparison, monitoring the strain as a function of a constant stress can be achieved rather easily and accurately. The term relaxation is commonly used generically to describe all such deformation processes.

The superposition of both the relaxation and retardation spectrum is necessary when, for example, loading a cylindrical test sample. The cross-sectional area of the cylindrical specimen will become larger in time when subjected to a compressive load and would therefore lead to a decrease in the stress level within the material. Currently, it is more common and much easier to experimentally determine the strain relaxation at a constant stress rather than the stress relaxation at a constant strain.

The relaxation behavior of plastic materials can be described via a 4-parameter viscoelastic model that is comprised of a parallel-series arrangement of springs and dash pots as represented in Fig. 6.4. The basic elements of the viscoelastic model consist of a **spring** with a characteristic elastic modulus E_0 and a **dash pot** with a characteristic viscosity η_0 combined in series. The remainder of the model is the serial combination of the previously defined element with a parallel combination of a spring E_{rel} and a dash pot η_{rel}. The result is a model that can be used to predict the time-dependent stress-strain response of polymeric materials.

The most familiar parameter within the model is the elastic modulus E_0, which is experimentally measured at very low loads and within a time interval less than one minute. The deformation behavior of the individual elements within the model when subjected to a constant load σ_0 results in:

elastic:

$$\varepsilon_0 = \frac{\sigma_0}{E_0} \tag{6.2}$$

viscoelastic/relaxed[45]*:*

$$\varepsilon_{rel} = \frac{1}{E_{rel}}\left(1 - e^{-t/\tau}\right) \cdot \sigma_0 \text{ where } \tau = \eta_{rel} / E_{rel} \tag{6.3}$$

viscous:

$$\varepsilon_v = \frac{t \cdot \sigma_0}{\eta_0} \tag{6.4}$$

The summation of the individual elements is used to calculate the total deformation as:

$$\varepsilon_{tot} = \varepsilon_0 + \varepsilon_v(t) + \varepsilon_{rel}(t) \tag{6.5}$$

$$\varepsilon_{tot}(t) = \left(\frac{1}{E_0} + \frac{t}{\eta_0} + \frac{1}{E_{rel}}\left(1 - e^{-\frac{t}{\tau}}\right)\right) \cdot \sigma_0 \tag{6.6}$$

For plastics, it is common to define the ratio between the applied stress and the time-dependent strain by the **creep modulus E_c** as:

$$E_c = \frac{\sigma_0}{\varepsilon_{tot}(t)} \tag{6.7}$$

Since both ε_{rel} and ε_v increase as a function of time, the corresponding time-dependent creep modulus will decrease. Though the interpretation of the elastic modulus should always remain as a material inherent characteristic, the time-dependent modulus of the material must be considered for design calculations.

After the removal of the stress at the time event t', the sample will immediately recover by the elastic contribution ε_0 and then slowly retrieve via relaxation by the viscoelastic contribution ε_{rel}. This deformation response is defined as **recovery deformation**. The viscous irreversible contribution ε_{eq} is negligible for elastomers and highly cross-linked thermosets since they are incapable of flow in the cured state. Therefore, the recovery once the stress has been removed can be modeled by the expression:

$$\varepsilon_{tot}(t > t') = \varepsilon_{tot}(t') - \varepsilon_0 - \varepsilon_{rel}(t') \cdot e^{-\frac{(t-t')}{\tau}} \tag{6.8}$$

The deformation behavior of a polyoxymethylene (POM) specimen subjected to the step load of both 10 and 15 MPa and the corresponding recovery is represented in Fig. 6.5. The x-axis in the graph is plotted on a linear scale between 0 and 10 hours and then again on a different linear scale between 10 and 1,000 hours[46]. The x-axis during recovery is illustrated in the same fashion.

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[45] The retardation time is the time for which $\varepsilon_{rel} = 0.632 \cdot \sigma_0 / E_{rel}$, therefore when $\tau = t$ in Eq. 6.3.

[46] Which causes the discontinuity during the course of deformation.

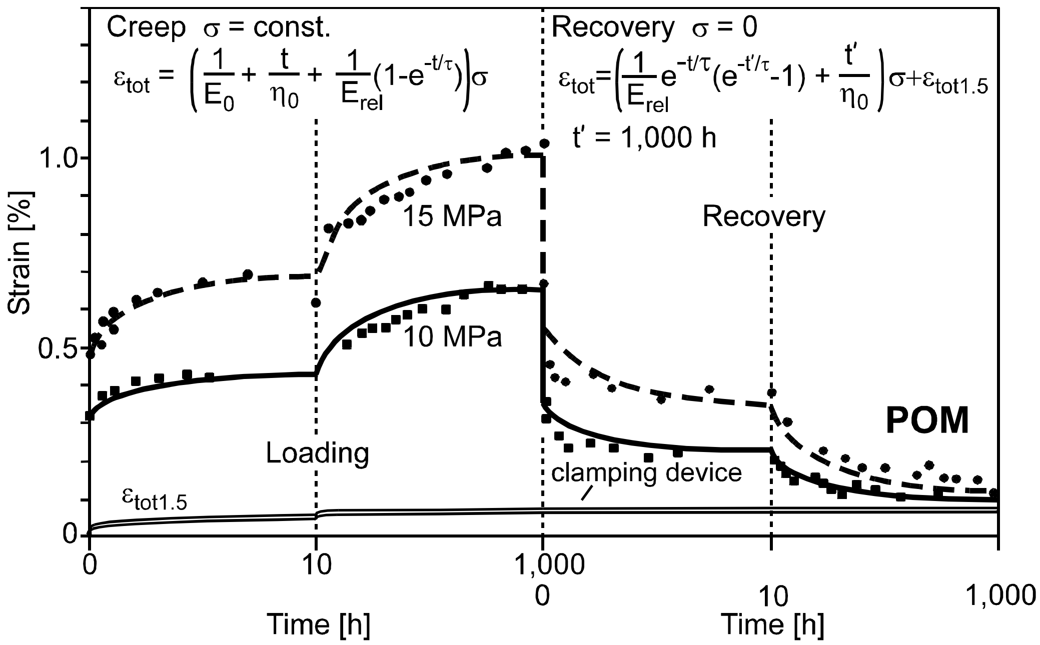

Fig. 6.5: Deformation behavior of polyoxymethylene (POM) to loading and recovery. Additional loading by the clamping device: 1.5 MPa

Such time-dependent investigations are typically conducted over extended time intervals. Therefore, any time-dependent influences on the testing conditions or instruments must be taken into consideration. Since the magnitude of the loads are above the threshold of linear viscoelastic behavior, E_{rel}[47] and η_0 are dependent on the stress level.

Only the viscous or elastic-viscous deformation is characterized by irreversible **plastic flow**, while the viscoelastic part contains a reversible **creep** behavior. However, these definitions are sometimes not consistently applied. The definition of the creep modulus is one such example in which the total strain $\varepsilon_{tot}(t)$ contains not only a creep but also a flow part.

Principally, a similar behavior is observed if a constant strain is applied instead of a constant stress. A time-dependent deformation of the material when subjected to a constant stress is referred to as **retardation**. In comparison, the decay in the stress when the material is subjected to a specified degree of deformation is referred to as stress relaxation or simply **relaxation**. According to these definitions, the time-dependent relation between stress-strain can also be described by the relaxation or retardation modulus. When subjected to a large stress or a stress condition over extended time intervals, the magnitude of the **relaxation modulus** will always be smaller than the **retardation** or **creep modulus**. Correspondingly, the relaxation time will be less than the retardation time.

6.1.1.1 Linear Viscoelastic Behavior

The individual deformation parts can be superimposed provided that the admissible stress level is not exceeded. Beyond the stress limit, the comparisons between the

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[47] $E_{rel} = \sigma_0 / \varepsilon_{rel}$ where ε_{rel} is the ultimate value for relaxation deformation

modeled and experimental results dramatically diverge with an increase in the load magnitude. The point at which the linear mathematical models and the physically measured response diverge signifies the end of the **linear viscoelastic** region.

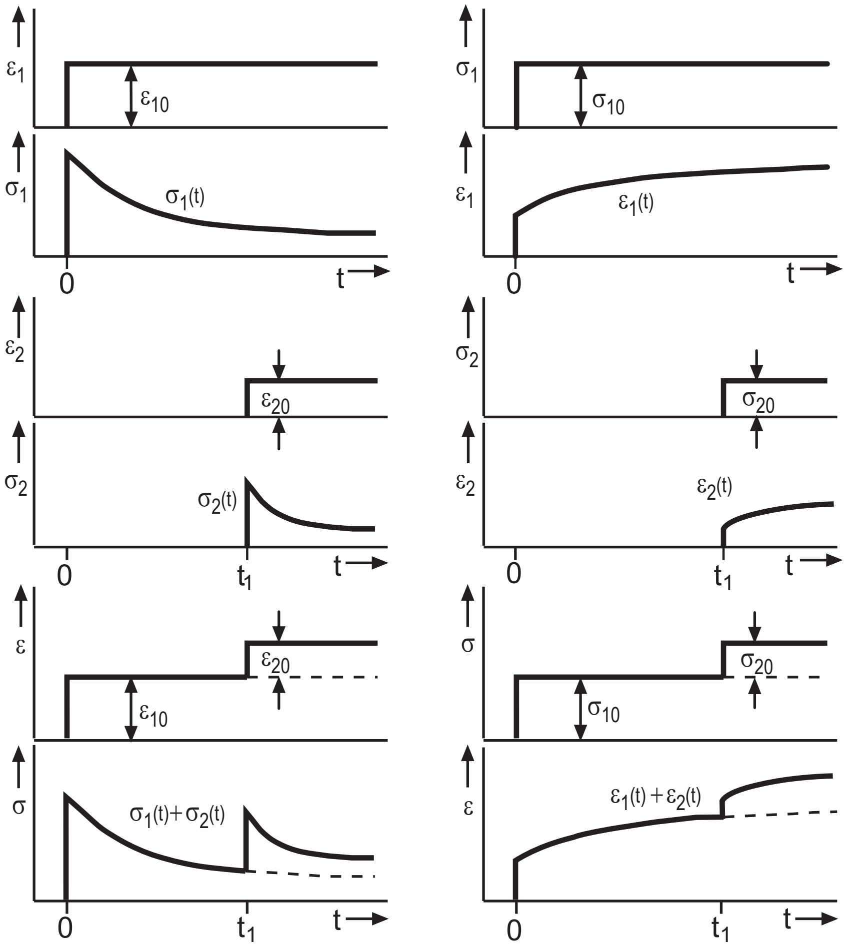


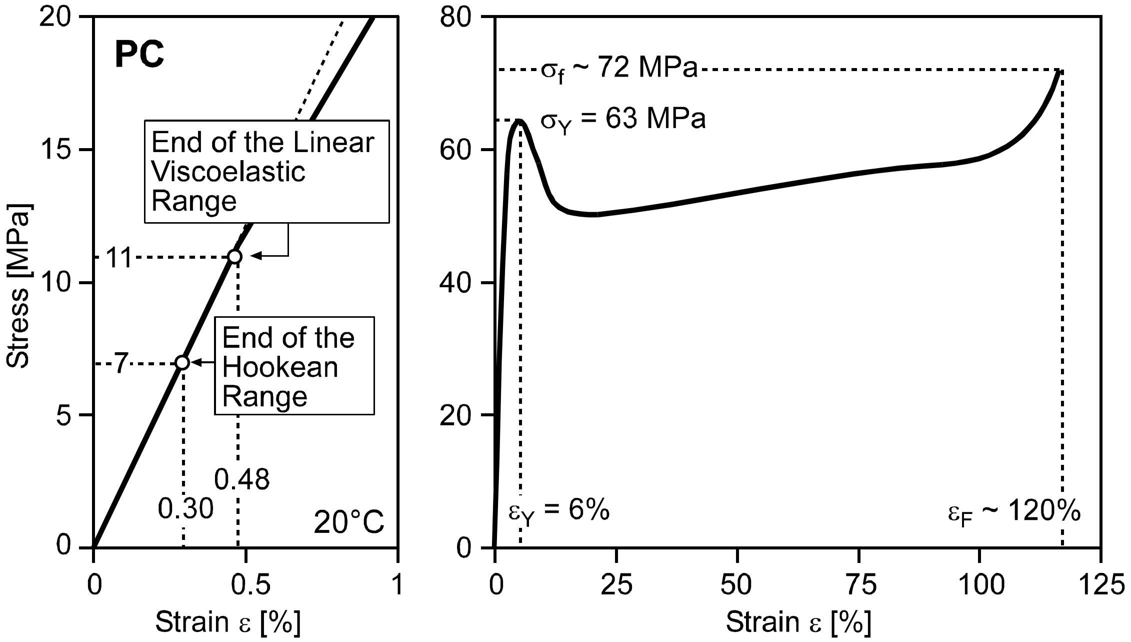
Fig. 6.6: Illustration of the Boltzmann superposition principle
left: Case 1 right: Case 2

The Boltzmann superposition principle has been successfully applied for the modeling of the linear viscoelastic behavior of polymeric materials as long as the material remains within the linear viscoelastic regime as illustrated in Fig. 6.6. The figure is divided between two loading schemes. The left column illustrates the stress relaxation response when the strain is applied in a step-wise manner at $t = 0$ and the superposition of the step strain event at $t = t_1$.

In comparison, the right column illustrates the strain (or creep) response when the stress is applied in a step-wise manner at $t = 0$ and the superposition of the step stress event at $t = t_1$ as summarized below:

Case 1: If the strain ε_{10} results in the stress $\sigma_1(t)$
and the strain ε_{20} results in the stress $\sigma_2(t)$,
then $\varepsilon_{10} + \varepsilon_{20}$ results in the total stress $\sigma_1(t) + \sigma_2(t)$

Case 2: If the stress σ_{10} results in the strain $\varepsilon_1(t)$
and the stress σ_{20} results in the strain $\varepsilon_2(t)$,
then $\sigma_{10} + \sigma_{20}$ results in the total strain $\varepsilon_1(t) + \varepsilon_2(t)$



Material	σ_0 [MPa]	σ_{lv} [MPa]	σ_Y [MPa]	E [MPa]
PP	1.5	2.5	35	1,450
PA 66 (std.)	3	5	50	1,300
PBT	3	5	55	2,600
POM	5	7	65	2,900
ABS	5	7	45	2,400
PC	7	11	63	3,700
SAN	7	10	71[1]	3,700

[1] σ_F = stress at failure, no yield limit

Fig. 6.7: Stress at the end of the elastic (σ_0) and linear viscoelastic (σ_{lv}) regions as well as yield strength (σ_Y) for polycarbonate (PC). Values for other thermoplastics are listed in the corresponding table
The linear viscoelastic boundary in the upper-left diagram, is determined by isochronous stress-strain curves.

The material deformation behavior above the linear elastic limit is indicated by a strong inelastic deformation part and thus identifies the onset of **nonlinear viscoelastic** behavior. Within the nonlinear viscoelastic regime, the ability to analytically calculate the stress-strain relation becomes increasingly complicated by the time-dependent effects. Thus, the mathematical models to predict this type of deformation cannot be posed in a closed form. Unfortunately, most structural parts are loaded within this deformation regime. Therefore, the behavior of highly loaded parts is difficult to predict and can only be approximated.

The boundary between the elastic and linear viscoelastic deformation regimes for most thermoplastics and elastomers is positioned far below their measured strength limits as listed in Fig. 6.7. The elastic deformation boundary is determined by the deviation from a linear response represented in the stress-strain diagram (elongation rate = const.).

In comparison, the linear viscoelastic deformation boundary is determined by the deviation from a linear response in an isochronous stress-strain diagram (time = const.). The limiting deviation tolerance from the linear response is not specified. The location of the limiting deviation is normally decided upon by visual inspection of the experimental results. In Fig. 6.7, a 1% deviation from gradient is considered as the end of both the elastic and linear viscoelastic deformation regions. The selection of the appropriate deviation is dependent on whether the diagram is presented on a constant rate or constant time scale.

6.1.1.2 Strength Properties

The mechanical properties of polymeric materials depend on both the manufacturing process and testing conditions. Therefore, the experimental methods used for the determination of the engineering constants of plastics are of great concern. Without proper consideration of processing, the mechanical properties of plastics within the same polymer grade cannot be adequately evaluated for material selection purposes.

The material selection process is often further complicated by an abundance of experimental results from which a limited number of meaningful characteristic values must be selected that are dependent on the testing conditions. Additionally, the manufacturing process, environmental conditioning, and test geometry of each sample specimen must be defined.

A methodological standard has been accepted by many plastic manufacturers based on the material database **CAMPUS** (**C**omputer **A**ided **M**aterial **P**reselection by **U**niform **S**tandards). The method of sample preparation and testing according to the CAMPUS rules is designed for the unequivocal comparison of the material properties confined to a set of uniform geometry, equipment, manufacturing, and testing conditions based on international standards.

The material properties that can be used to compare the stress-strain characteristics between various plastics is shown in Fig. 6.8 and listed as follows:

- **Yield stress** σ_Y defined by the location in which the upward gradient of the stress-strain curve takes the value of zero for the first time;
- **Tensile strength** σ_t defined by the location of maximum stress;
- **x%-Offset yield stress** $\sigma_{x\%}$ defined by the location in which the stress-strain curve deviates from the linear response by x%-strain;
- and **stress at failure** σ_f refers to the situation when no yield stress is present or when no x%-yield stress can be determined.

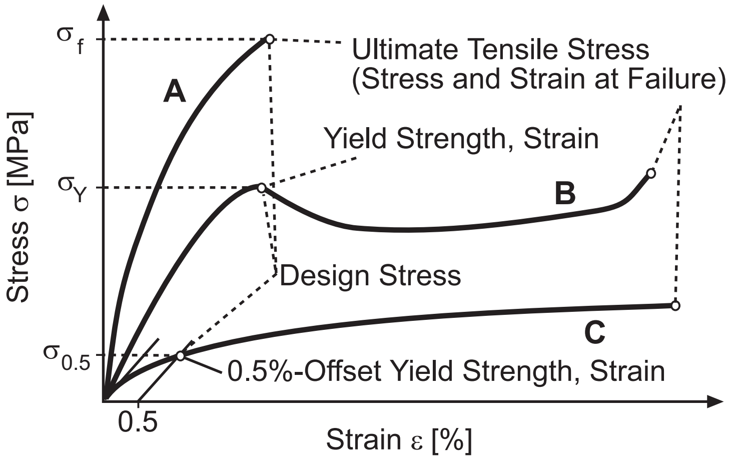


Fig. 6.8: Characteristic values determined from typical stress-strain curves
A: brittle plastics
B: tough plastics that exhibit a yield point
C: tough plastics that do not exhibit a yield point

The x%-offset yield stress can be used as a characteristic value provided that the stress-strain curve of the material does not indicate a distinct yield stress and the failure occurs at relatively high values of strain. Therefore, the x%-offset yield stress is commonly used for materials that exhibit significant nonlinear behavior. All short-time characteristic values can be inferred from standard stress-strain curves similar to Fig. 6.8. The most important strength parameter that is retrieved from such curves is the tensile strength σ_t. The tensile strength is insensitive to the cross-sectional geometry and thus gives the best overview of the tensile properties.

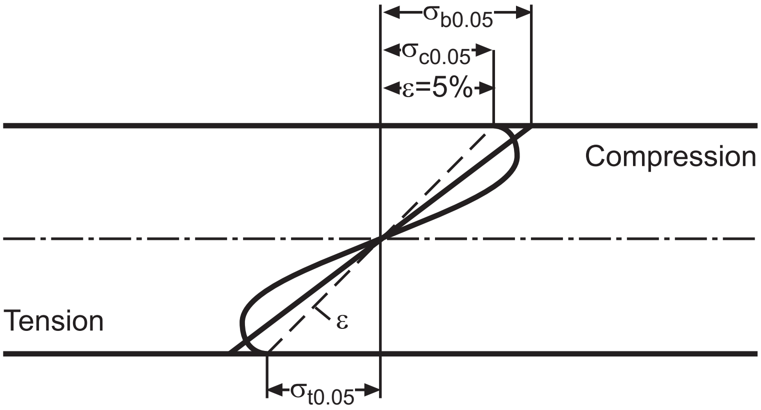


Fig. 6.9: Stress and strain distribution through the thickness direction subjected to bending
$E_{b0.05}$ = 1140 MPa, bending modulus via 5% edge strain
$\sigma_{b0.05}$ = 57.1 MPa, via theory of elasticity for bending
$\sigma_{c0.05}$ = 41.5 MPa, via compression test
$\sigma_{t0.05}$ = 35.0 MPa, via tensile test

Besides data from tensile testing, data from bending tests are often used because bending is a frequent mode of loading in practice. In spite of the popularity of this approach, several arguments against the specification of the **bending strength** σ_b

exist. One major argument is that the stress-strain distribution through the cross-sectional thickness is linear only for small deflections.

At higher load levels, the linear strain distribution is maintained within the deflected element, however the stress distribution becomes "S"-shaped in both tension and compression as shown in Fig. 6.9. This nonlinear stress distribution is not considered in standard calculations for the stress at the surface from the bending moment M and resistive the moment of inertia W (σ_b = M/W). Therefore, the yield stress σ_Y determined by bending is typically larger than that determined by tensile test investigations.

The bending strength calculated according to this standard method (ASTM 790) is unrealistically high especially for tough thermoplastics. This error can be compensated for with the use of the bending strength values from shaped parts if the wall thickness corresponds to the thickness of the piece to be analyzed. The same concept applies for the determination of the elastic modulus via bending. For identical materials, the measured stress-strain diagrams in tension, compression, and bending deviate from one another as shown in Fig. 6.10.

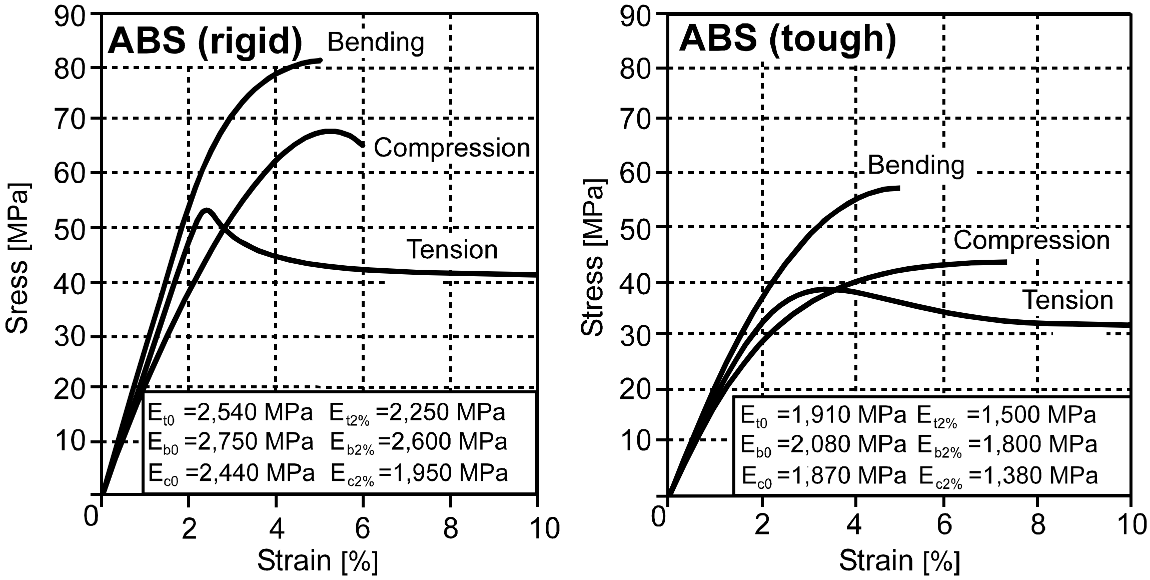


Fig. 6.10: Stress-strain curve for a rigid (left) and a tough (right) acrylonitrile-butadiene-styrene copolymer (ABS) subjected to (c) compression, (t) tension, and (b) bending

E_0 = standard elastic modulus

$E_{2\%}$ = secant modulus measured between 0 and 2% strain

Compression increases the risk of specimen failure by buckling, the **compression strength** σ_c is measured on short samples. The difficulties in obtaining sufficiently parallel and plane surfaces and the formation of a complex multiaxial deformation and wear problems in the contact area between the testing machine and the specimen surface frequently lead to a misrepresentation of the experimental results. In contrast to tensile loading, an inherent benefit of compression loading is that small notches and surface cracks become insignificant. Therefore, it is common to find that the strength of the same plastic measured in compression is higher than the values obtained in tension. Long fiber reinforced plastics, particularly thermosets, are an exception since the fiber tensile strength dominates the strength properties.

The **shear strength** τ_f can be determined by applying a torsional load on a cylindrical test specimen. Irregular deformation of the test sample (bulging, buckling, etc.) during testing should be avoided. This may be difficult since most thermoplastic materials exhibit a low stiffness. The shear strength is more commonly determined from the tensile strength by the use of the shear stress failure hypothesis as:

$$\tau_f = 0.5 \cdot \sigma_{tf} \tag{6.9}$$

If the tensile state is exactly known, the change of shape hypothesis (Huber, von Mises, Hencky) can be applied as:

$$\tau_f = 0.58 \cdot \sigma_{tf} \tag{6.10}$$

Therefore, the following ranking of the relative strengths can be generally applied:

Shear Strength	**<**	**Tensile Strength**	**<**	**Compression Strength**	**<**	**Bending Strength**

There is a further reason for not specifying several characteristic strength values. The geometry and manufacturing conditions of the test specimens can lead to anisotropic properties, for example via molecular or filler orientation during tool filling. The anisotropy developed during processing may have a stronger influence on the strength than the type of load. More importantly, orientation during tool filling and the associated difference between the strength properties parallel and perpendicular to the flow direction has to be considered. With the use of the tensile strength, the numerical calculation of the stress within a part contains an inherent margin of safety since (except for fiber-reinforced plastics) the values are lower than those measured via compression or bending investigations. Thus, the utilization of tensile test investigations to determine the strength characteristics is not only technically preferred but safer as well.

6.1.1.3 Deformation Properties

Similar to the definition of the stress properties, the corresponding strain properties shown in Figs. 6.8 and 6.11can be listed as follows:

- **Yield strain** ε_Y defined by the yield stress σ_Y;
- **Strain at failure** ε_f defined by the stress at failure σ_f;
- **Elastic modulus** (Young's modulus) E_0 defined by Hooke's law, it describes the stress-strain relation within the linear region;
- **Secant modulus** E_s describes the stress-strain relationship at a specific load level;
- **Tangent modulus** E_t describes the resistance against further deformation at a specific load level;
- and **Creep modulus** E_c defined for time-dependent loads as the relation between the stress and the time-dependent deformation, see Fig. 6.4.

The deformation of plastics results from the superposition of the elastic, relaxation, and viscous responses. Since the relaxation and viscous responses are time and load dependent, a plastic will exhibit a more linear elastic behavior the faster it is loaded. For normal static stress cases, the stress level in which the stress-strain curve deviates from a linear behavior is very low. In particular at, elevated material temperatures a nonlinear response is observed even for small magnitudes of deformation.

Due to equipment-related issues, the elastic modulus is analyzed not as a tangent modulus through the zero point, but more commonly as a secant modulus between the 0.05 and 0.25% strain values. A small amount of pre-load is necessary for the compensation of free play inherent to the clamping device. Therefore, the determination of the elastic modulus is complete after a strain of 0.25% is achieved and can be expressed by the relation:

$$E_0 = \frac{\sigma_{0.25} - \sigma_{0.05}}{\varepsilon_{0.25} - \varepsilon_{0.05}} \tag{6.11}$$

This definition for the elastic modulus is necessary for plastics that exhibit a nonlinear deformation behavior in order to enable an accurate means for material comparison. Nevertheless, the deformation behavior cannot be characterized sufficiently for the calculation of a designed part. Therefore, three distinct definitions for the elastic modulus can be defined and are given in Fig. 6.11.

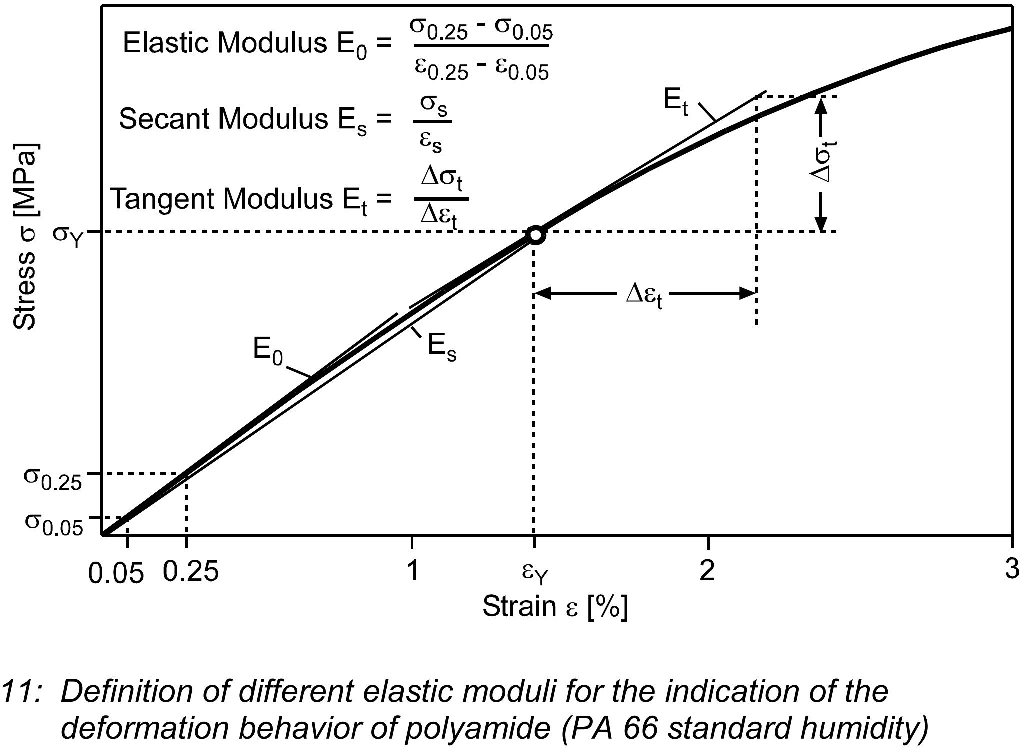


Fig. 6.11: Definition of different elastic moduli for the indication of the deformation behavior of polyamide (PA 66 standard humidity)

Technological designers use the measured **elastic modulus** E_0 only for material comparison purposes or for material selection. Since typical parts designed with plastic are strained more than 0.25%, the calculation of these parts should be conducted with the use of the **secant modulus** E_s. The secant modulus is defined as the ratio between the stress σ and the strain ε generated by a known load relative to the origin by the relation:

$$E_s = \frac{\sigma_s}{\varepsilon_s} \tag{6.12}$$

Therefore, the value of the secant modulus will decrease with an increase in the applied load. This simplification, applied to the analysis of the nonlinear deformation behavior of polymeric materials, is helpful and should be considered. A third possibility arises from the concept of the **tangent modulus** E_t, which is defined by the slope of the stress-strain curve at a specified load level. The tangent modulus is not commonly used in practice, however it can be used for the description of further deformation behavior, e.g. bulging of preloaded parts.

The material temperature plays an instrumental role in the deformation process. Its relation to the glass transition temperature is of particular importance. Plastics are typically ductile and tough if the application temperature is above the glass transition temperature, however they become rigid at low temperatures. This is true particularly for semi-crystalline thermoplastics whose amorphous phase can soften without loss of useful properties of the material. The material will retain cohesion through the crystalline regions. The duration and magnitude of the applied load are also important, as previously discussed.

The strains are directly coupled to the stress during deformation. The strain observations at the characteristic points for the strength were indicated in Fig. 6.8. For the characterization of the behavior of polymeric parts, the **strain at failure** is suitable for brittle plastics and the **0.5%-strain** can be specified for tough materials that do not exhibit a defined yield strength. The strain at failure of tough plastics results in a deformation that is far above admissible values. Nevertheless, the strain at failure is of interest for the qualitative evaluation of plastics.

The elongation strain and the total deformation can be used as an indication of the materials tolerance to damage during critical loading conditions. The ability to undergo large deformation when subjected to a high loading condition characterizes the **toughness** of a material. The total deformation alone cannot be used to describe the capacity of the material to absorb energy; in particular this is not possible over small strain intervals. Typically, a plastic with a small strain at failure cannot absorb much energy. The yield strain is indicated by a unique load limit. Above this load limit, the part geometry cannot be maintained and deformation will occur, for example by necking.

It is obvious that no material deformation will occur in the absence of an applied force at a constant temperature. However, the absence of a measurable deformation must not be construed as absence of a load. For multi-axial states of loading, the strain in the transverse direction can significantly reduce the observed strain in the axial direction. The evaluation of plastics through the **allowable strain** limit is not as arduous as that by means of the admissible stress limit. This is true because strain characteristics are inherently less sensitive to external influences than either the stress or strength characteristics.

The stress-strain curves for both an amorphous impact-resistant polystyrene (PS-I) and a semi-crystalline polyamide 66 (PA 66 dry) are shown in Fig. 6.12. While the yield stress is a clear function of the loading rate and temperature, the change in the elongation at yield strain exhibits a dependence that is not as pronounced. Therefore, the admissible strain values are more meaningful and much easier to handle than the admissible stress. A further argument in favor of the applicability of the admissible strain is that the loading of parts can be more readily measured with the use of strain gauges.

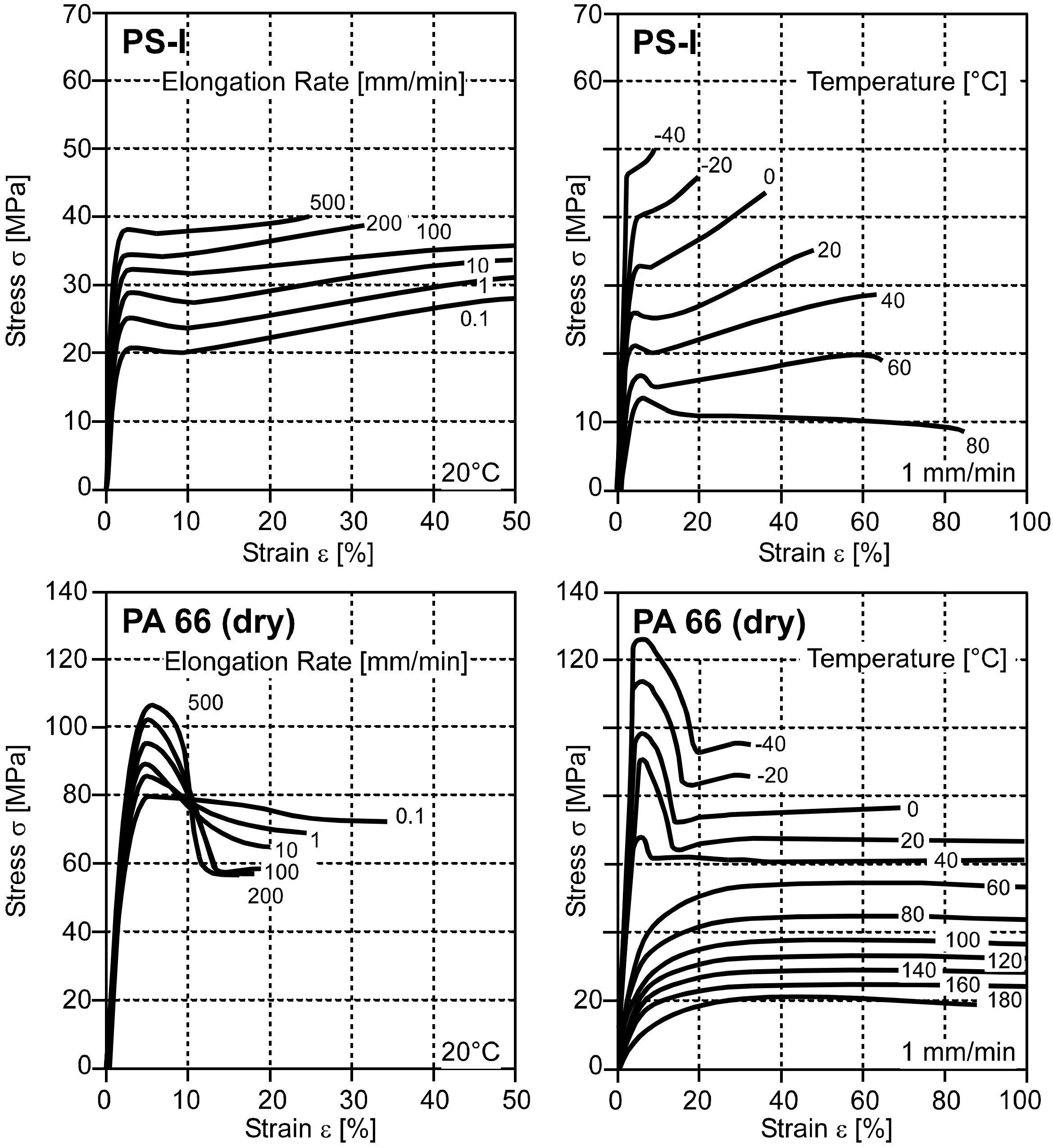


Fig. 6.12: Stress-strain curves of two thermoplastic materials as a function of the elongation rate, and temperature (courtesy BASF)
above: amorphous impact-resistant polystyrene (PS-I)
below: semi-crystalline polyamide 66 (PA 66 dry)

On the other hand, it must be realized that the parts constructed from metals are more commonly characterized by their admissible stress limits rather than by exceeding their admissible strains. Thus, interpretation via the admissible strain limit is much more important for plastics than for metals. This is due to the fact that the strength of metals is greater than that of plastics by a factor of 10 on average.

Likewise, the elastic modulus is greater by a factor of 100, i.e. at a specific load the deformation is larger by a factor of 100. Significant complications arise during dimensioning of plastic parts. Thus, some engineers trained for steel design lack the understanding of deformation processes that plastic engineers must have.

A substantial problem for the analysis of the stress-strain response of a complex geometry is the lack of a suitable and well-established strain failure hypothesis that would allow the determination of the equivalent strain. Strain hypotheses are useful for the description of imposed deformations (e.g. thermal strain), ductile deformation (creep and flow), and to gain fundamental knowledge of the deformation state through experimental measurements (including the determination of the stress state from strain measurements).

A hypothesis that can be used to determine the failure boundary can be determined by use of the **maximum strain theory** in which an equivalent strain ε_{eq} can be calculated from the maximum shearing strain γ_{max} and the use of Poisson's ratio ν:

$$\varepsilon_{eq} = \frac{\gamma_{max}}{1+\nu} \tag{6.13}$$

From the equivalent strain, the Huber-von Mises-Hencky analogy can be used to determine the equivalent stress.

A useful approach for the determination of the strain limits was made by Menges, who suggested a relationship between critical strain and maximum allowable elongation. It was found that a critical strain could be used to define the maximum allowable elongation for various load types as $\varepsilon_{eq} < \varepsilon_{crit}$. As an approximate value, particularly for uniaxial deformation, the critical strain should be considered.

6.1.1.4 Temperature

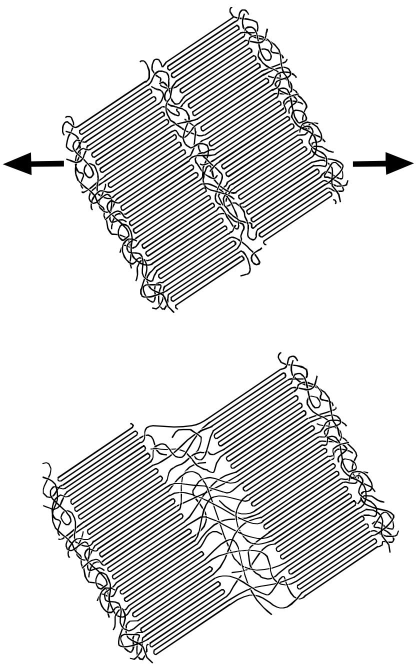
The glass transition temperature T_g and the melting temperature T_m are the two most important characteristic temperatures that should be considered when evaluating plastics for load bearing applications (see also Sections 5.3 and 5.4). The glass transition temperature is particularly important for amorphous thermoplastics in which the intertwined molecular chains are the dominating contributors to the geometric stability of the part. However, the polymeric structure is capable of molecular stability above the glass transition temperature either through the crystalline order within semi-crystalline thermoplastics or else through chemical cross-links, for both thermosets and elastomers.

The crystalline regions within semi-crystalline materials do not exhibit melting until well above the glass transition temperature. In comparison, with cross-linked thermosets and elastomers will not melt but degrade thermally at higher temperatures, where the process is, of course, irreversible. The degradation temperature is, with only a few exceptions (e.g. aramid fibers), higher than the melting temperature. Thus, the characteristic temperatures can be ranked in the following order:

Glass Transition Temperature	<	Crystallization Temperature	<	Melting Temperature	<	Thermal Degradation Temperature

Within the crystalline regions, the physical cohesive forces are optimal and can be destroyed only by increasing the temperature above the melting temperature. In comparison, the amorphous regions exhibit a transformation at much lower temperatures, as shown in Fig. 6.13.

The glass transition temperature is the temperature of largest change of stiffness prior to melting.



Polymer	Moisture	T_g	T_m
	[°C]	[°C]	[°C]
PE		- 125	135
PP		20	170
POM		- 65	178
PET		80	255
PBT		43	223
PA 6	dry 0.2	78	223
	std 3.0	28	223
	sat 8.0	- 8	223
PA 46	dry 0.2	94	287
	std 3.5	31	287
	sat 9.7	-10	287
PA 66	dry 0.2	50	264
	std 2.7	39	264
	sat 7.2	- 6	264
PA 610	dry 0.1	77	222
	std 1.5	48	222
	sat 3.2	19	222
PA 66/6	dry 0.2	81	243
	std 2.7	29	243
	sat 7.4	- 6	243
PA (amor.)	dry 0.3	152	-
	std 2.9	114	-
	sat 5.0	97	-

Fig. 6.13: Deformation of the amorphous phase within semi-crystalline thermoplastics above the glass transition temperature (T_g) [30]

T_m = melting temperature, dry= dry sample
std = conditioned sample (23°C and 50% relative humidity)
sat = saturated (maximum water content)

Below the softening or glass transition temperature, the amorphous and the semi-crystalline domains are brittle and rigid. Therefore, amorphous thermoplastics can only be used for constructive purposes below the glass transition temperature. However, semi-crystalline thermoplastics are capable of remaining relatively rigid well above their glass transition temperature.

Polymeric materials become tough and ductile above the glass transition temperature range (approximately 10°C). The same behavior can be realized for cross-linked thermosets and loosely cross-linked elastomers. The chemical bonds provide the cohesion of the material, the physical bonds determine the stiffness. The softening of the latter make the material tougher.

Depending on the temperature, the same plastic can assume different stress-strain properties. This can occur over a relatively narrow temperature range for semi-crystalline thermoplastics, as shown in Fig. 6.14.

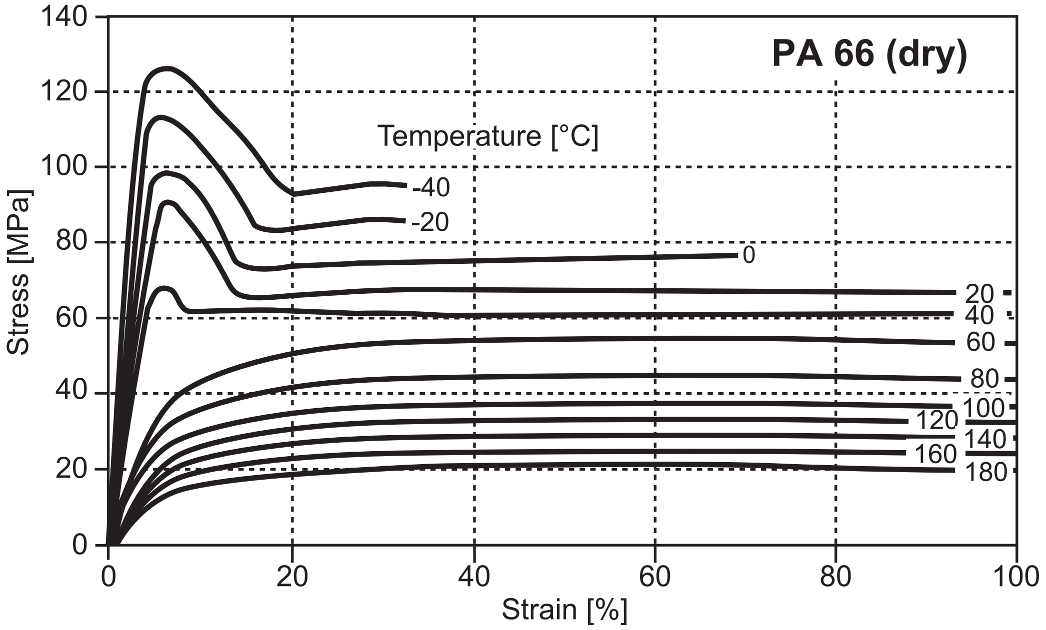


Fig. 6.14: Stress-strain curves of a semi-crystalline polyamide (PA 66 dry) at different testing temperatures, v = 1 mm/min

As long as the material temperature is below the glass transition temperature, the amorphous areas are vitrified and a yield strength is evident. Otherwise, the material is ductile and a well-defined yield strength cannot be observed. Likewise, a pronounced temperature dependence of both the stiffness and characteristic dimensional values is present.

The influence of moisture content may be superimposed on the temperature effects. For example, the glass transition temperature of polyamide can vary from -8 to 90°C depending on the moisture content, while the crystalline phase melts at about 220°C (PA 6) or 260°C (PA 66). The moisture not only influences the stiffness values, but the entire deformation behavior as well.

With increasing elongation rate, the time-dependent inelastic deformation response is reduced, i.e. the stress-strain curve becomes steeper, and therefore the magnitude of the elastic modulus is increased. Likewise, both the yield stress and the tensile strength increase when subjected to faster elongation rates. On the other hand, the strain at failure decreases as shown on the left in Fig. 6.15.

In general, the influences of the temperature on the strength and elastic modulus are more significant than the elongation rate. This is evident for polyamide (PA 66 standard humidity) when comparing either the influence of temperature or duration of loading up to a strain of 4% as shown in Fig. 6.16. The conversion shown by the right graph in Fig. 6.16 illustrates the stress-strain relationship for a fixed time. Such diagrams are called isochronous stress-strain diagrams.

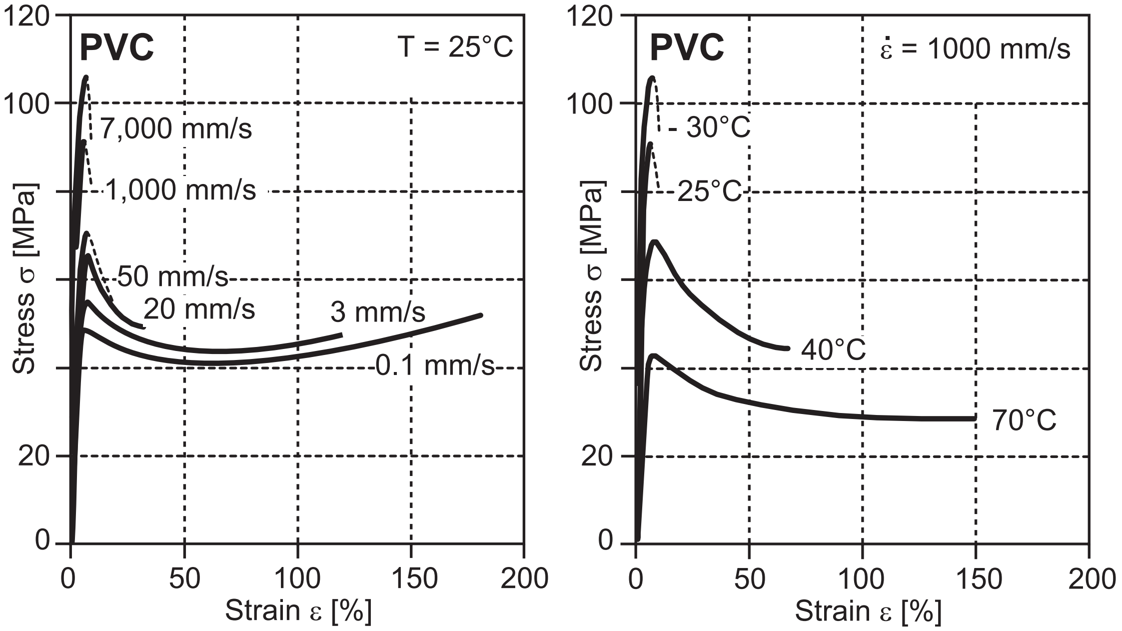


Fig. 6.15: Stress-strain curves of polyvinyl chloride (PVC) via tensile testing (courtesy Retting)
left: temperature (25°C) right: load rate (1000 mm/s)

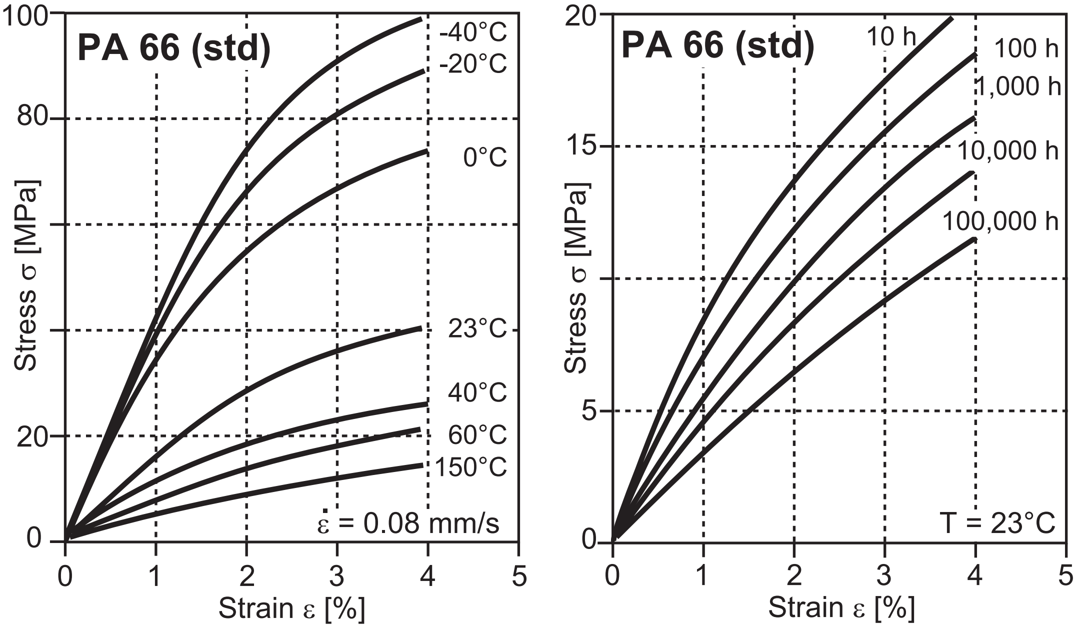


Fig. 6.16: Magnification of the polyamide (PA 66 standard humidity) stress-strain curves at various temperatures and after specific times
(strain at failure is approximately 20-40%)

6.1.1.5 Water Absorption

Water absorption is of particular interest for materials such as polyamides. Polyamides are of technical interest since they are commonly used in parts that must resist both high thermal and mechanical loading conditions in the automotive, the electric, and electronic industries. Water diffuses into polyamides, becomes located within the loosely packed amorphous regions and preferably replaces the hydrogen bonds. Therefore, three states are specified for polyamides when reporting their properties i.e. dry, standard humidity, and saturated as listed in Table 6.1.

State	Environment	Water Content PA 6 / PA 66 [weight-%]	Glass Transition Temperature PA 6 / PA 66 [°C]	Melting Temperature PA 6 / PA 66 [°C]
Dry	no humidity	< 0.2	78 / 90	223 / 264
Standard Humidity	23°C / 50% relative humidity	3 / 2.7	28 / 39	
Saturated	water	8 / 7.2	-8 / -6	

Table 6.1: Influence of the water content on the glass transition temperature and the melting temperature of polyamides (PA 6 and PA 66)

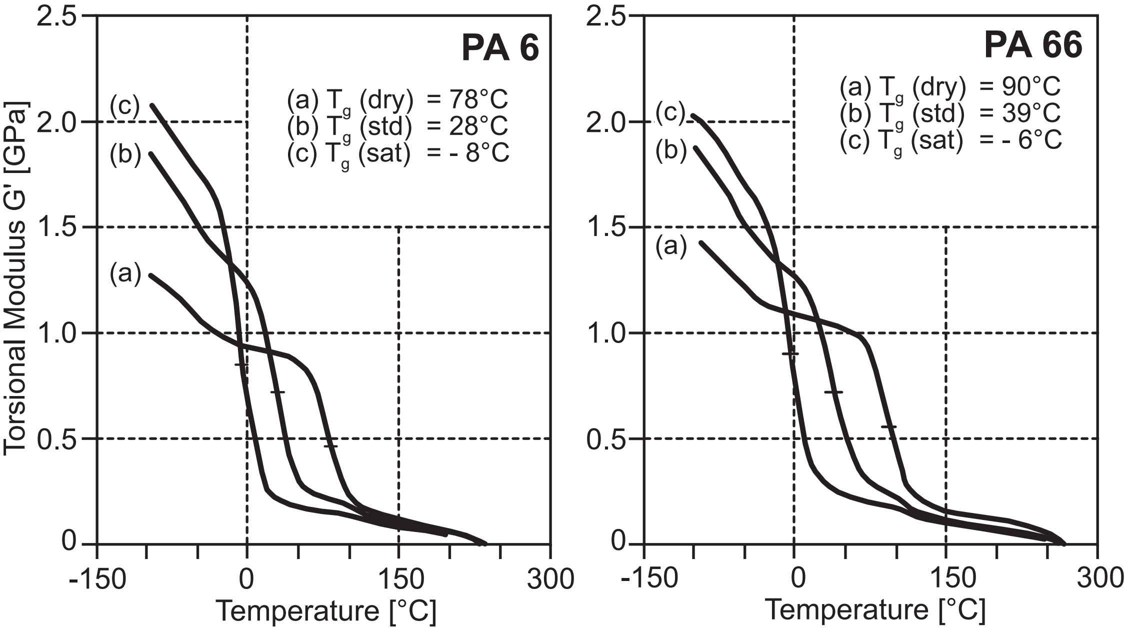


Fig. 6.17: The temperature-dependent torsional modulus (low loading) of polyamides (PA 6 and PA 66) with the glass transition temperature for dry, standard humidity, and saturated conditioning

These absorption states are obtained by storing the material at the specified conditions until no further change in the percent water content over time can be observed. Extended time intervals are necessary to reach a uniform humidity content, for example a previously dried part with a thickness of 2 mm can require months of environmental conditioning to obtain the normal, standard humidity state.

The absorption process can be substantially accelerated when the specimen is subjected to elevated temperatures. However, short-term conditioning as well as daily fluctuations in the humidity lead to only a small modification in the water content. Though the properties of polyamides are strongly dependent on the moisture content, daily fluctuations in the moisture content will lead to only minor alterations in their properties.

One significant influence of an elevated water content on the properties of polyamides is the shift in the glass transition temperature to lower temperatures due to the absorption of water as shown in Fig. 6.17. The glass transition temperature for a polyamide (PA 6 when dry) is determined to be 78°C. However, the glass transition temperature quickly reduces to 28 and 8°C with a water concentration of 3 and 8%, respectively. While water content increases the ductility of polyamide at temperatures below freezing of water, the ductility decreases.

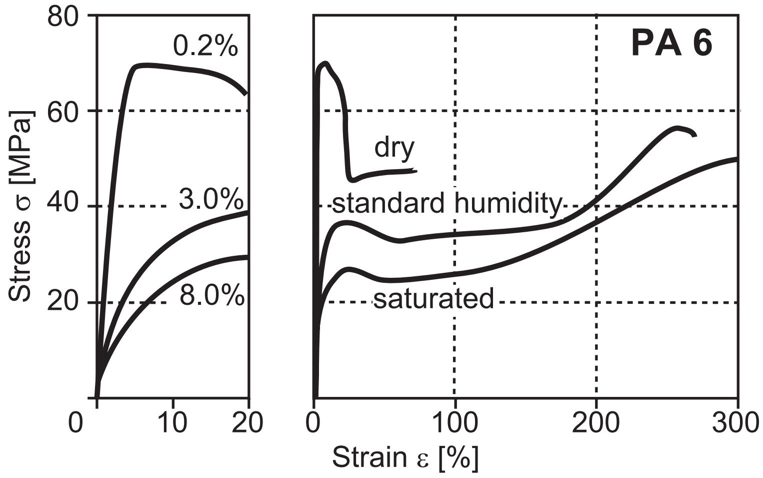


Fig. 6.18: Stress-strain diagrams of dry, standard humidity, and saturated polyamide (PA 6)

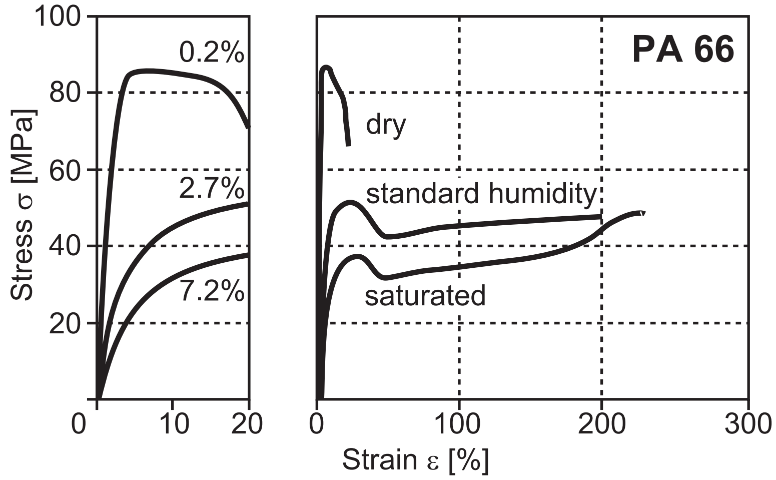


Fig. 6.19: Stress-strain diagrams of dry, standard humidity, and saturated polyamide (PA 66)