

Review

From Melt to Structure: The Science and Technology of Flat Soda–Lime–Silicate Glass for Structural Engineers

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Abstract

Flat soda–lime–silicate glass is the dominant glass type used in contemporary buildings. This paper provides a comprehensive and integrated review of the scientific principles and technological processes that underpin its manufacture, processing, and structural performance. The discussion spans the glass transition and the nature of the glassy state; the network structure of soda–lime–silicate glass, with its inherent lack of long-range order; and its physical and mechanical properties, including fracture. The industrial production of flat soda–lime–silicate glass (melting, float-forming, and annealing) and its subsequent processing (thermal tempering, chemical strengthening, and coating) are described in detail, with emphasis on how they influence residual stresses, surface and edge quality, and structural reliability. Environmental considerations and ongoing advances in energy efficiency and decarbonisation are also examined. By linking the fundamentals of glass science to modern structural design standards, particularly the forthcoming Eurocode 10 for glass structures, the article seeks to equip structural engineers with an informed understanding of glass as a high-performance material for innovative and sustainable design.

Keywords: glass structures; soda–lime–silicate glass; flat glass; float glass; thermal tempering; chemical strengthening; Eurocode 10; sustainability

1. Introduction

Flat glass is among the oldest man-made building materials. The Romans were producing window glass as early as the first century CE [1]. Paradoxically, flat glass is also one of the most modern, multifaceted, and forward-looking building materials, shaping the appearance of contemporary architecture perhaps unlike any other.

In the latter half of the 20th century, the convergence of three major technological innovations—the float process of glassmaking, the development of tempered glass, and glass lamination with “stiff” interlayers—combined with a deeper scientific understanding of glass, unlocked a wide range of new possibilities for the use of glass in structural applications. A broad distinction can be made between primary and secondary structural glass elements, based on their role in the performance of the structure and the consequences of failure [2]. Secondary elements do not contribute to the overall stability of a structure; they must withstand and transmit only those actions, direct and indirect, that are applied to them (and, if necessary, provide a sufficient degree of fall-through safety). These elements are not responsible for carrying loads from other parts of the structure. Curtain walls and large windows—which became emblematic of 20th-century modernist architecture, with its



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emphasis on light, transparency, openness, and utility and its rejection of ornamentation—are prime examples of secondary glass members. Other examples include balustrades, horizontal glazing on a supporting grid and frameless glazing systems, in which glass panels are point-fixed to a supporting structure without perimeter framing.

Over the past three or four decades, the structural use of glass in buildings has expanded rapidly, driven by the desire to increase transparency and openness and to create the illusion of a non-existent structure. This has led to the use of glass in primary elements such as beams, beam-columns, and walls. These elements play an integral part in the overall stability of the structure (structure-forming elements) and are typically subjected to significant in-plane loads, with buckling often being a critical consideration. Glass as a structural material is also increasingly making its way into interior elements, like stairs and railings.

Building renovation—an area where balancing the respect for the identity of the existing construction with the need to meet contemporary standards of spatial quality, efficiency, functionality and safety is essential—provides immense opportunities for the structural use of glass, by virtue of its distinctive ability to contrast with the features to be preserved and thus to enable an open dialogue between the old and the new. The same applies to the safeguarding of heritage buildings and sites, with an additional emphasis on the reversibility of the interventions.

The forthcoming Eurocode 10 for the design of glass structures is expected to further promote the adoption of glass as a structural material. It currently comprises three parts. The first part, entitled “General rules”, addresses glass-specific basis of design items [3]. The second part, “Out-of-plane loaded glass components”, presents structural design rules for glass components and assemblies primarily subjected to lateral loads (and which do not usually take loads from other elements); it also covers the design of commonly used supports, namely continuous edge supports, point supports and supports for cantilever systems [4]. The third part is dedicated to “In-plane loaded glass components” and the joints connecting them, which are often integrated into the primary structure in terms of overall performance and stability [5]. A fourth part, with guidance on “Glass selection relating to the risk of human injury”, retains the status of European “Technical Specification”, with the character of recommendation [6]. It is currently not intended to be elevated to full Eurocode status in the future [7], as it deals with issues of “safety and accessibility in use”—basic requirement 4 for construction works of the Construction Products Regulation [8] (not to be confused with basic requirement 1, “mechanical resistance and stability”).

This paper surveys the science and technology of flat soda–lime–silicate glass—by far the most widely used glass in construction—with particular emphasis on its relevance to structural engineering. The review brings together glass science fundamentals, the structure and properties of soda–lime–silicate glass, the main industrial processes (melting, float-forming, annealing, thermal tempering, and chemical strengthening), and key environmental considerations. Laminated glass (including building-integrated photovoltaic systems) and connections are outside the scope of this study, as each of these subjects would require a separate article for adequate treatment. Through this synthesis, structural engineers may gain a deeper understanding of glass as a structural material and its expanding role in innovative, sustainable, and reliable building solutions, thereby supporting more informed decisions in design, detailing, and specification.

2. The Nature of Glasses

It is not the purpose of this work to formulate a comprehensive definition of the term *glass*. Instead, it is sufficient to adopt the definition provided in the ASTM Standard Terminology of Glass and Glass Products, which states that “Glass is an inorganic product of

fusion which has been cooled to a rigid condition without crystallisation" [9] (Conradt [10] critiques the ASTM definition of glass as overly narrow; he argues that (1) glasses are not necessarily inorganic materials and (2) the formation routes for glass extend well below melt-cooling).

2.1. The Glass Transition and the Glassy State

To better understand glasses, it is instructive to examine the volume–temperature relationship of a single-component glass-forming liquid cooled at constant pressure (e.g., [11])—shown in Figure 1. This relationship encapsulates the essential characteristics of the glassy state—its nature, its formation process, and its distinction from other states of matter. We begin our analysis at point “a”, in the (equilibrium) liquid state. As the liquid cools, its volume decreases gradually along the linear path “abc”, consistent with a positive volumetric thermal expansion coefficient. Point “b”, at temperature T_m , corresponds to the melting point of the crystalline phase. If the cooling rate is sufficiently low, crystallisation occurs at this temperature, or just below (point “c”), leading, in general, to an abrupt decrease in volume (water is a notable exception)—a jump discontinuity—due to the formation of an ordered structure. This constitutes a first-order phase transition. Continued cooling of the crystalline material causes further volume contraction along the crystal path “de”, governed by its volumetric thermal expansion coefficient (smaller than that of the liquid). However, if the cooling rate is sufficiently high, crystallisation is bypassed and the liquid enters a supercooled liquid state, following the path “cf”—the extrapolation of “abc”. The supercooled liquid is metastable with respect to the corresponding crystalline phase. As cooling progresses, the viscosity of the supercooled liquid increases rapidly. At sufficiently low temperatures, atomic mobility becomes too slow relative to the cooling rate, preventing the system from reaching the volume of the supercooled liquid structure corresponding to each temperature before that temperature is lowered further. As a result, the graph begins to deviate smoothly from the “cf” path (without any discontinuity) and eventually transitions into a near-straight line, often approximately parallel to “de” [12], ending at point “g” when cooling is faster (less time is allowed for atomic rearrangements), or at “h” when cooling is slower. The material in this near-straight, low-temperature part of the curve behaves essentially as a solid—it has reached a glassy state, with a structure characterised by the absence of long-range order (as indicated, there is not a unique glassy state; the properties of a glass depend on its thermal history [12]). The region between the onset of deviation from the supercooled liquid path and the attainment of a seemingly rigid condition corresponds to the so-called glass transition (or freezing-in process [13]). It does not occur at a single sharply defined value of the temperature, but over a temperature range. Nonetheless, a glass transition temperature T_g , dependent on the cooling rate, is customarily defined to mark the transition. The glass transition temperature is determined experimentally, typically using dilatometry or calorimetry, with results depending on the specific operational definition and testing procedure used [14]. The dependence of T_g on the cooling rate is usually modest [12]. The IUPAC Compendium of Chemical Terminology defines T_g as the temperature at which the viscosity of glass is 10^{13} dPa · s [15]. For soda–lime–silicate glasses (described below), T_g is typically in the range 530–580 °C, depending on the specific composition and thermal history [16]. The glass transition is not strictly a second-order phase transition—there is a characteristically kinetic dimension to it [17]—and glass is a non-equilibrium system (writing in 1995, physicist and Nobel laureate Philip Warren Anderson described the glass transition as “probably the deepest and most interesting unsolved problem in solid state theory”). However, this latter point warrants nuance; once formed, glasses can persist without a practical limit under ordinary conditions.

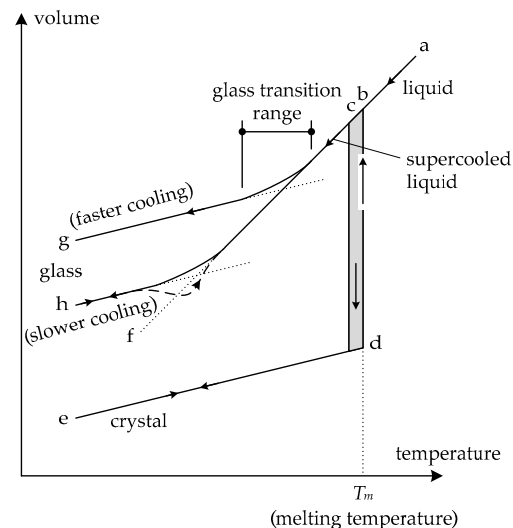


Figure 1. Schematic volume–temperature diagram (at constant pressure) for a glass-forming substance and the corresponding crystal (after [11]), illustrating that glass formation requires crystallisation to be bypassed.

When the glass at “h” is reheated, the state smoothly moves through the transition region along the dashed curve to the supercooled liquid state and ultimately to the liquid state. If the crystalline material at “e” is heated, it would follow the crystal path up to “d” (at the melting temperature T_m), then melt to reach point “b”, and subsequently follow the liquid path “ba”.

2.2. Viscosity of Glass-Forming Melts

According to the kinetic perspective described above, the relationship between viscosity and temperature—which varies significantly with chemical composition—plays a central role in determining the glass-forming ability of a liquid. Glasses are easily formed when either (1) the viscosity is very high at the melting temperature of the crystalline phase that would otherwise form or (2) the viscosity increases very rapidly as temperature decreases. In both cases, crystallisation is impeded by the kinetic barrier to atomic rearrangements created by high viscosity [14]. The viscosity–temperature relationship is also fundamentally important in all facets of glassmaking technology, as will be discussed shortly in the context of the float process of flat glass production. Technologists define reference points on the viscosity–temperature curve of glass-forming systems, each linked to a specific viscosity value and processing event (Table 1).

Table 1. Viscometric fixed points.

	Viscosity (Pa·s)	Typical Temperature Ranges for Soda–Lime–Silicate Glasses (°C) [18]	Significance	Standards
Working point	10^3		The majority of glass-forming operations are carried out between the working and the softening points	[19,20]
Softening point (Littleton temperature)	$\sim 10^{6.6}$ (1)	720–730	Temperature at which a round fibre of glass, nominally 0.65 mm in diameter and 235 mm long, elongates under its own weight at a rate of 1 mm/min when the upper 100 mm of its length is heated at the rate of $(5 \pm 1)^\circ\text{C}/\text{min}$	[21,22]

Table 1. Cont.

	Viscosity (Pa·s)	Typical Temperature Ranges for Soda–Lime–Silicate Glasses (°C) [18]	Significance	Standards
Annealing point	$\sim 10^{12.2}$ ⁽²⁾	540–550	Temperature at which internal stresses in a glass are substantially relieved in a matter of minutes	[23,24]
Strain point	$\sim 10^{13.7}$ ⁽²⁾	500–510	Temperature at which internal stresses in a glass are substantially relieved in a matter of hours	[23,24]

⁽¹⁾ The exact viscosity at the softening point depends on density and surface tension; the value given here corresponds approximately to a density of 2.5 g/cm³ and a surface tension of 300 mN/m. ⁽²⁾ At temperatures corresponding to the annealing and strain points, the viscosity of glass is highly time dependent.

3. Structure and Properties of Soda–Lime–Silicate Glass

3.1. Composition and Network Structure

Figure 2 provides an overview of the major glass families, categorised by their chemical composition. The most commonly used glass in construction is soda–lime–silicate glass, a member of the family of silicate glasses, the main components of which are silicon dioxide (SiO₂), sodium oxide (Na₂O), and calcium oxide (CaO).

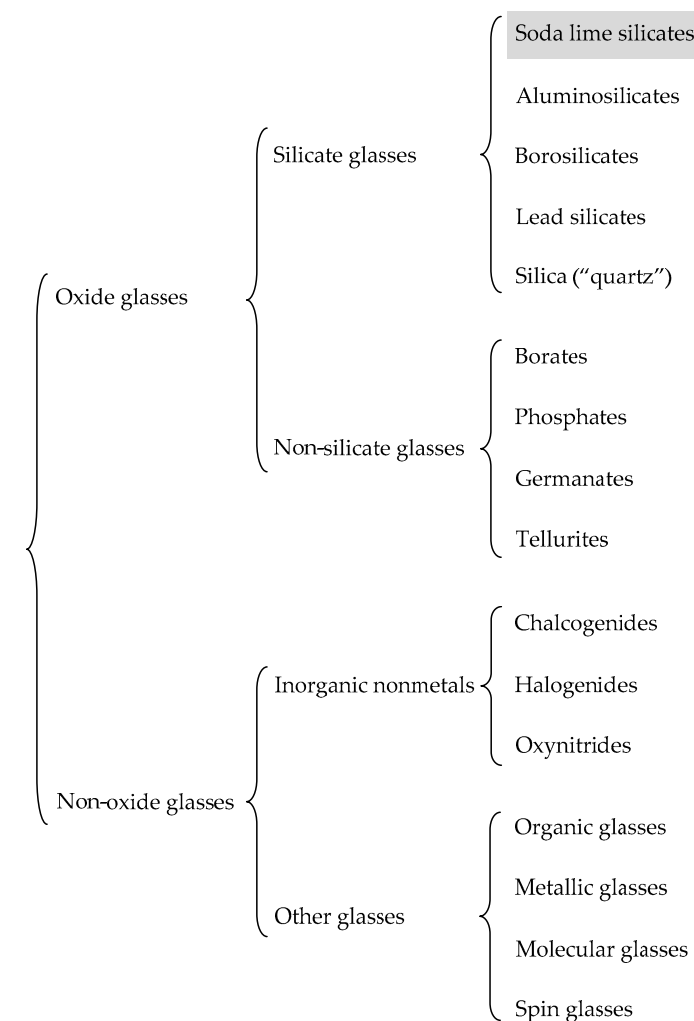


Figure 2. Glass compositional families (after [25]). The diagram includes glasses that do not fit the ASTM definition given earlier. Soda–lime–silicate glasses, the focus of this article, are highlighted.

The backbone, or skeleton, of silicate glasses consists of a disordered three-dimensional network of corner-linked SiO_4 tetrahedra, with each Si–O bond being primarily covalent in character. In soda–lime–silicate glass, some of the corner oxygens are bonded to a sodium ion or to a calcium ion, rather than to another silicon atom, splitting up the network (Figure 3)—thus, silicon is the network former, while the sodium and calcium ions act as network modifiers [11,13,16]. Oxygen atoms that participate in Si–O–Si bonds are known as bridging oxygens (they act as a bridge between neighbouring silicon atoms), while those that participate in Si–O–Na or Si–O–Ca bonds are known as non-bridging oxygens. Each sodium ion leads to the formation of one non-bridging oxygen and each calcium ion leads to the formation of two non-bridging oxygens, with the Na–O and Ca–O bonds being chiefly ionic in character. The addition of the alkali oxide Na_2O to silica is very effective in lowering the viscosity of the glass-forming melt [11,26]. However, it reduces the chemical durability of the resulting binary Na_2O – SiO_2 glass, due to the relatively high mobility of sodium ions in the glass structure. To provide for improved chemical durability, the alkaline earth oxide CaO is also added [13]. Moreover, commercial soda–lime–silicate glass may contain small amounts of magnesium oxide (MgO) and aluminium(III) oxide (Al_2O_3), both of which have a positive influence on chemical durability [13], as well as other minor constituents. Aluminium(III) oxide is often referred to as an intermediate oxide; aluminium enters the glass network as AlO_4^- tetrahedra, with charge balance provided by Na^+ ions. Consequently, its addition decreases the number of non-bridging oxygens and increases network connectivity [16]. Table 2 gives some examples of typical commercial compositions of soda–lime–silicate float glass, expressed as weight percentages of the various oxides.

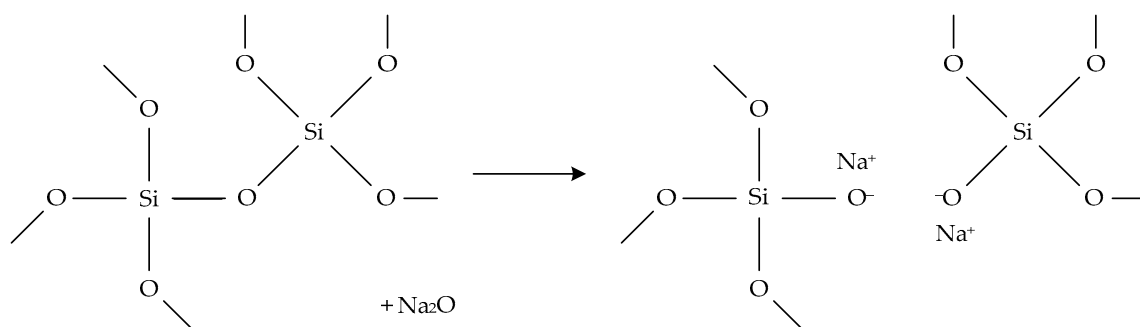


Figure 3. Schematic illustration of the splitting up of the network of corner-linked SiO_4 tetrahedra by the addition of Na_2O (network modifier), with each sodium ion leading to the formation of one non-bridging oxygen. The relatively high mobility of sodium ions in the glass structure lowers melt viscosity and reduces chemical durability.

Table 2. Examples of commercial soda–lime–silicate float glass compositions from different manufacturers in the UK (Float01–03) and Turkey (Float04) [27].

Oxide (wt %)	Float01	Float02	Float03	Float04
SiO_2	72.279	72.394	71.776	72.137
Na_2O	13.723	13.249	13.475	13.934
CaO	8.780	10.192	8.688	9.729
MgO	4.016	3.011	4.231	3.420
Al_2O_3	0.800	0.626	1.130	0.337
K_2O	0.110	0.209	0.368	0.125
SO_3	0.205	0.217	0.217	0.232
Fe_2O_3	0.061	0.062	0.066	0.054
TiO_2	0.025	0.039	0.048	0.031
Cr_2O_3	0.001	0.001	0.001	0.001

The above compositional and structural characteristics have direct implications for the physical and mechanical properties of soda–lime–silicate glass.

3.2. Physical and Mechanical Properties

The most striking feature that contributes to the widespread use of soda–lime–silicate glass in buildings is its transparency to visible light. This transparency is the result of two highly non-trivial characteristics: (1) the existence of an electronic bandgap, i.e., an energy range where practically no absorption of electromagnetic radiation occurs, and (2) the (nearly) complete absence of light scattering, due to the disordered nature of glass on the length scales that are relevant for the interaction with visible light [28].

Because of their inherent lack of long-range order, glasses are mechanically homogeneous and isotropic materials (at scales larger than a few nanometres), with an almost perfect linearly elastic behaviour in tension up to brittle failure. The most important physical and mechanical properties of basic soda–lime–silicate glass products are summarised in Table 3. The relatively high coefficient of thermal expansion makes soda–lime–silicate glass susceptible to thermal shock failure.

Table 3. Physical and mechanical properties of basic soda–lime–silicate glass products: generally accepted values for calculations not requiring high accuracy, according to EN 572-1 [29].

Density	$\rho = 2500 \text{ kg/m}^3$
Knoop hardness	$HK_{0.1/20} = 6 \text{ GPa}$
Young modulus	$E = 70 \times 10^3 \text{ N/mm}^2$
Poisson ratio	$\nu = 0.2$
Specific heat capacity	$c_p = 0.72 \times 10^3 \text{ J/(kg} \cdot ^\circ\text{C)}$
Average coefficient of linear thermal expansion between 20 °C and 300 °C	$\alpha_T = 9 \times 10^{-6} \text{ }^\circ\text{C}^{-1}$
Resistance against temperature differential and sudden temperature change	40 °C
Thermal conductivity	$\lambda = 1 \text{ W/(m} \cdot ^\circ\text{C)}$
Mean refractive index to visible radiation (at 589.3 nm)	$n = 1.5$
Corrected emissivity	$\varepsilon = 0.837$
Minimum light transmittance (measured according to EN 410) for designating a transparent glass product as clear glass	0.89 (2 mm nominal thickness) 0.87 (4 mm nominal thickness) 0.85 (6 mm nominal thickness) 0.81 (10 mm nominal thickness)

3.3. Fracture

As with all brittle materials, the tensile strength of glass (as determined in standardised tests) depends strongly on inevitable surface flaws or cracks (arising from manufacturing processes, post-manufacturing handling, and service life), not necessarily visible to the naked eye, which give rise to stress concentrations that cannot be redistributed by plastic deformation. Linear elastic fracture mechanics provides a good model to interpret the brittle failure of glass. Of the three basic modes of crack-surface displacement, mode I (opening mode) is by far the most relevant, with the contributions of modes II (sliding mode) and III (tearing mode) being considered negligible [30]. Virtually instantaneous failure occurs when the (mode I) stress intensity factor K_I at a crack reaches the critical value K_{Ic} , or (plane strain) fracture toughness, which is regarded as a material constant (in the case of soda–lime–silicate glass, $K_{Ic} = 0.75 \text{ MPa} \cdot \text{m}^{1/2}$ can be adopted for practical

purposes [31]). The size of the tensile-loaded surface area or edge length affects glass strength; the larger the area or length, the greater the probability of finding a crack of critical size for the applied stress state (size effect). Similarly, the state of stress is also important; the probability of finding a significant crack at right angles to the maximal tensile stress is greater under biaxial stress than under uniaxial stress [31–33].

Due to environmental effects, crack growth is still possible when $K_I < K_{Ic}$, and this is of major significance for structural applications. Indeed, glasses experience delayed fracture under sustained load in the presence of water, a phenomenon known as static fatigue [34–38] (even in a vacuum, the tensile strength of several glasses is slightly time-dependent [39], a phenomenon known as inert fatigue). Static fatigue in soda–lime–silicate glass can be explained by the slow (sub-critical) growth of cracks caused by a three-step strain-enhanced chemical reaction at the crack tip between water in the environment and glass [40,41]: (1) a water molecule is adsorbed to a strained Si–O–Si bridge at the crack tip (through hydrogen-bond formation with the bridging oxygen and the interaction of a lone electron pair on the oxygen atom of the water molecule with a silicon atom); (2) the Si–O bond and the O–H bond in the adsorbed water cleave to form two hydrogen-bonded silanol groups (Si–OH); (3) the remaining hydrogen bond cleaves to leave fracture surfaces covered by silanol groups (Figure 4). The entire process is called dissociative chemisorption, and a basic prerequisite is the deformation of the network of corner-linked SiO_4 tetrahedra in the vicinity of the crack tip due to an applied tensile stress. The time to failure is determined by how long it takes for a crack to grow to its critical size, at which point failure is virtually instantaneous.

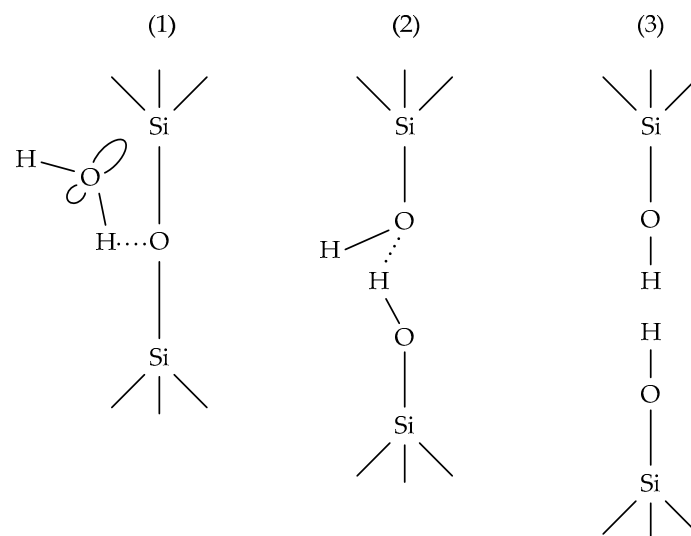


Figure 4. Static fatigue of soda–lime–silicate glass in tension. Three-step chemical reaction between water molecule and strained crack-tip siloxane bond leading to sub-critical crack growth [40]: (1) adsorption, (2) formation of two hydrogen-bonded silanol groups, (3) separation.

Figure 5 shows an idealised log–log plot of crack velocity v as a function of the stress intensity factor K_I for soda–lime–silicate glass in a moist environment. The representation abstracts the following features exhibited by experimental results [34]:

- (1) The data tend to fall into four regions (labelled 0, I, II, and III in Figure 5), implying different rate-controlling processes.
- (2) In region I, the strong dependence of v on K_I can generally be described by an empirical power law of the form.

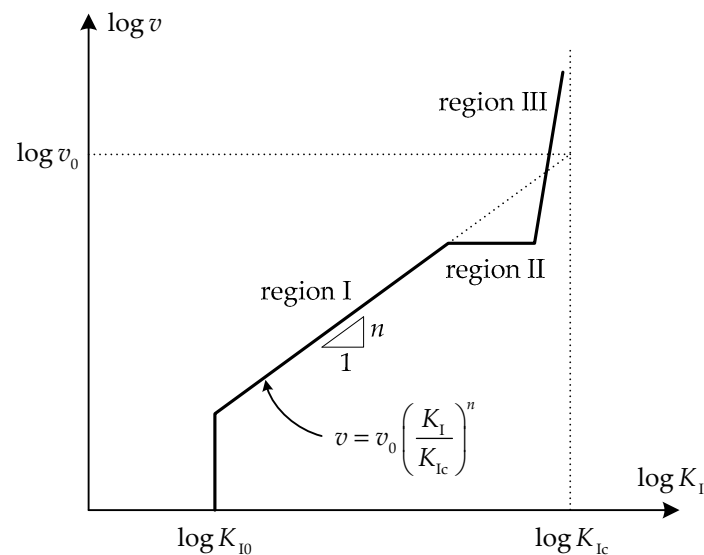


Figure 5. Idealised representation of crack velocity as a function of the stress intensity factor K_I (log–log plot) for soda–lime–silicate glass in a moist environment.

$$v = v_0 \left(\frac{K_I}{K_{Ic}} \right)^n \quad (1)$$

where n and v_0 are parameters (whose geometrical meaning is indicated in Figure 5), as proposed by Evans and Wiederhorn [42]. It is believed that crack velocity in this region is reaction-rate controlled.

- (3) In region II, the crack velocity is essentially independent of K_I , which suggests that v is controlled by the transport of the active species to the crack tip.
- (4) In a narrow region just below K_{Ic} (region III), the velocity does not depend on the environment and is even more sensitive to K_I than in region I. In inert environments, this part of the curve would extrapolate to lower crack velocities.
- (5) There exists an environment-sensitive minimum value K_{I0} of the stress intensity factor, called the static fatigue limit or threshold, below which no observable crack growth occurs in soda–lime–silicate glasses, even in the presence of a reactive medium (region 0). Moreover, cracks held below this limit exhibit a delayed resumption of growth upon reloading [43–45]. Experimental evidence supports the hypothesis that the static fatigue limit results from the formation of a leaching layer at the crack tip, caused by interdiffusion between alkali ions in the glass (Na^+) and hydronium ions (H_3O^+) in the water, rather than from crack tip blunting due to network dissolution. The leached layer exhibits higher fracture toughness than the bulk material.

The following parameters have a significant influence on the relationship between stress intensity and crack velocity in soda–lime–silicate glass [35]:

- (1) Increasing the chemical activity of water in the environment causes a parallel shift in regions I and II of the log–log plot of the $v - K_I$ curve towards higher crack velocities (thus mainly affecting v_0 without significantly changing the exponent n in Equation (1)). It is not the absolute amount of water that matters, but rather the ratio of the actual partial pressure of water to that at saturation [46]. For structural applications, this implies that it is the relative humidity of the air that governs the velocity of sub-critical crack growth [31].
- (2) An increase in temperature in liquid water causes a parallel shift in region I of the $v - K_I$ curve towards lower stress intensities [47]; there is no indication that the

threshold K_{I0} is significantly affected [48]. The $v - K_I$ curve in vacuum, corresponding to region III, also exhibited a shift towards lower stress intensities with increasing temperature [39].

- (3) In aqueous media, the exponent n in region I increased with decreasing pH [49]. There is also a strong effect of pH on the static fatigue limit [50].
- (4) All aspects of sub-critical crack growth are influenced by the chemical composition of the glass. Although there is generally a correlation between fatigue susceptibility and alkali content, this relationship is not systematic [35].

For soda–lime–silicate glass, reviews of available experimental data on the parameters v_0 and n appearing in Equation (1) can be found in [31,51,52]. For design purposes, in view of the considerable scatter, Haldimann [31] suggests the conservative adoption of $n = 16$, regardless of environmental conditions; this author further suggests the adoption of $v_0 = 6$ mm/s in general applications, with higher values being required for elements permanently immersed in water (e.g., $v_0 = 30$ mm/s). The Italian guidelines CNR-DT 210/2013 [53] recommend $n = 16$ and $v_0 = 2.5$ mm/s. For the environmental conditions relevant in the construction industry, the static fatigue limit lies in the range of $K_{I0} \approx 0.2K_{Ic} - 0.27K_{Ic}$ [51].

Because region II occurs only at relatively high crack velocities, the total time to failure is dominated by crack growth in region I [42].

As surface cracks do not grow when in compression, the compressive strength of glass is far higher than its tensile strength; however, this is irrelevant for virtually all structural applications.

4. The Industrial Production of Flat Soda–Lime–Silicate Glass

4.1. Raw Materials

The selection of raw materials for the industrial production of soda–lime–silicate glass depends on factors such as availability, cost, impurity levels, particle size distribution (which influences reaction kinetics), preferred reaction paths, and environmental considerations [54–57]. Silica sand and sandstone are usually the main sources of SiO_2 . Alkali and alkaline earth oxides can be introduced in the batch as carbonates, notably soda ash (Na_2CO_3) and limestone (CaCO_3). Dolomite, a double-carbonate mineral ($\text{CaMg}(\text{CO}_3)_2$), is often used to supply both calcium and magnesium oxides. Aluminium(III) oxide (Al_2O_3) is usually added to the batch via alkali feldspars, which also contribute alkali oxides to the batch. Fining agents are added to promote the removal of bubbles from the melt. The most commonly used fining agent in soda–lime–silicate glass production is sodium sulphate (Na_2SO_4), commonly known as salt cake, which also accelerates the dissolution of sand. Given the wide range of potential raw materials, this list is indicative rather than exhaustive.

In addition to raw materials, the so-called batch includes cullet (broken and crushed glass), which may be sourced internally, from glass discarded during the manufacturing process, or externally, from customers' scrap cuts and waste glass processed at recycling centres. Cullet reduces the demand for raw materials, decreases gas emissions, and serves as a flux, promoting the melting of the batch at lower temperatures. However, it can also introduce substantial impurities and contaminants. Moreover, variations in the composition of cullet—particularly in the case of externally sourced cullet—can complicate the control of the final glass composition.

4.2. Melting, Fining, and Conditioning

The raw materials are thoroughly mixed to form the batch, which is fed into the melting chamber of a furnace, typically as a blanket or islands on top of a bath of already molten

glass. Almost all furnaces in the flat glass sector (nearly 90%, according to Zier et al. [58]) are cross-fired regenerative furnaces fuelled mainly by natural gas—now the predominant energy carrier—and, to a lesser extent, by fuel oil; an electrical boost may provide up to 10% of the energy input [55]. A regenerator is a heat-recovery system consisting of a chamber filled with a checkerwork of refractory bricks. Regenerative furnaces are equipped with pairs of regenerators that alternate between heat-recovery and heat-restitution cycles. During one cycle, the refractory bricks are heated by the flue gases from combustion; during the subsequent cycle, the stored heat is transferred to the combustion air. In industrial furnaces, switching between the heat-recovery and heat-restitution cycles typically occurs every 20–30 min [56,59]. The batch is heated from above, by flames and hot flue gases in the combustion space, and from below, by the hot glass melt. Temperatures in the chamber reach up to 1600 °C.

The batch-to-melt conversion involves a series of complex chemical and physical phenomena, including decarbonation, dehydration, solid-state reactions, the formation of low-melting eutectics, and dissolution [54,60]. It is accompanied by the release of large volumes of gases (e.g., CO₂ from carbonate decomposition), resulting in a bubble-filled melt. Most of these bubbles are removed by buoyancy effects, in a process called (primary) fining. Fining agents release gases at high temperatures, which diffuse into the existing bubbles, causing them to expand and rise more quickly to the surface of the melt. This process produces a clearer melt, with only a few small bubbles, called seeds, remaining [54,56,61,62].

The melt is then gradually cooled and chemically and thermally homogenised (primarily by diffusive mixing) in the working end of the furnace, a stage known as melt conditioning. During this stage, the remaining seeds are reabsorbed, as most gases become more soluble in the melt with decreasing temperatures, in a process referred to as secondary fining or refining.

4.3. The Float Process

Float soda–lime–silicate glass is the most widely used form of basic flat glass product. Its name derives from the process by which they are produced.

The homogeneous molten glass exits the working end of the furnace through a canal, where its flow rate is precisely regulated by an adjustable gate (tweel), and is continuously poured, at about 1050 °C, onto a shallow pool of molten tin whose oxidation is prevented by a reducing atmosphere of hydrogen and nitrogen. Because it has a lower specific gravity, the molten glass floats on the tin, spreads out, and progresses down the tin bath, forming a ribbon with perfectly flat and parallel surfaces (except near the edges) under the combined effects of gravity, surface tensions, and the longitudinal tractive force applied by the roller conveyor at the downstream end of the float bath [63–68]. Molten glass simply poured atop molten tin adopts an equilibrium thickness determined by the glass/atmosphere, tin/atmosphere, and glass/tin surface tensions and by the molten glass and tin densities (Figure 6). For soda–lime–silicate glass, this equilibrium thickness is approximately 7 mm, a value that is largely insensitive to limited changes in the chemical composition of the glass, metal, or atmosphere. By increasing the speed of the roller conveyor at the downstream end of the float bath (i.e., by increasing the applied longitudinal tractive force), a thinner glass ribbon is obtained, but this also results in an unacceptable decrease in the ribbon's width. To achieve the desired thickness (substantially less than the equilibrium value) and width, a series of top roll machines—each equipped with a single water-cooled toothed wheel—is used to grip the ribbon at its edges, gently stretch it, and set its speed at the point of contact. In contrast to the complexity of producing ribbons thinner than the equilibrium value, thicker glass can be produced by a basic, simple system where the initial outward spread of the molten glass is limited by non-wetted parallel longitudinal guides; the glass

is then cooled to an appropriate temperature in the downstream area of the guides so that its viscosity becomes high enough to prevent further dimensional changes. In the terminal zone of the float bath, the glass ribbon is cooled to around 600 °C, at which point it can be lifted onto a roller conveyor without incurring surface damage (and while the tin remains liquid). The two faces of the ribbon (commonly referred to as the tin side and the air side) are chemically dissimilar, having been formed in contact with different environments [69]. Although first envisioned as a finisher of glass surfaces, the float bath is thus seen to be “a remarkable entity which also functions as a container, a conveyor, a forming unit, a chemical reactor and a heat exchanger” [66].

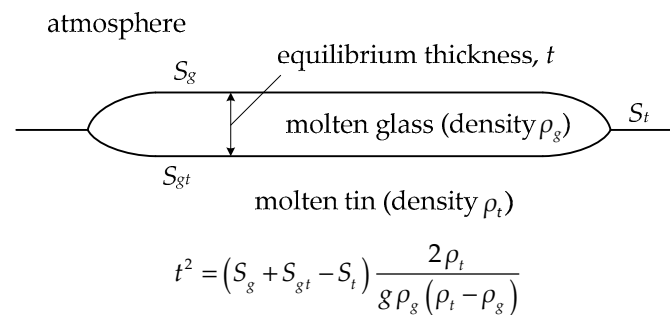


Figure 6. Vertical section of a pool of molten glass floating on molten tin illustrating the concept of equilibrium thickness under gravitational and surface tensional forces— S_g : glass/atmosphere surface tension; S_t : tin/atmosphere surface tension; S_{gt} : glass/tin surface tension; g : gravitational acceleration (after [63]).

4.4. Annealing

After the tin bath, the glass passes through an annealinglehr. Annealing is the process of cooling glass from above its transition range in a slow, controlled manner to minimise the development of residual stresses. Poorly annealed glass may crack or shatter during handling or even spontaneously. A typical annealinglehr used in float glass production is divided into two parts according to the dominant mode of heat transfer: a radiative cooling section and a convective (air-blast) cooling section. To modulate and more precisely control the cooling rates, the radiative cooling section is further divided into “zones”.

Let us briefly consider the mechanisms of stress generation during the cooling of a glass ribbon. Assume that the ribbon, initially at a temperature above the glass transition range, is cooled in such a way that its surface temperature decreases at a uniform rate. A parabolic temperature distribution is rapidly established across the thickness of the ribbon, with an overall amplitude proportional to the cooling rate and to the square of the ribbon’s thickness. The ribbon remains stress-free because the glass is sufficiently soft to prevent the buildup of stress. In other words, any stress that forms is immediately relaxed due to the low viscosity of the material. As long as the cooling rate remains constant, there is no change in thermal gradient across the thickness and the glass remains stress-free. At some temperature below the transition range, cooling is halted and the thermal gradient decays to zero. At this point, the glass responds elastically to the vanishing thermal gradient, developing residual stresses (Figure 7). This “classical” view on annealing, first proposed by Adams and Williamson [70], has long served as a foundational interpretative framework. Nevertheless, it represents an oversimplification of the actual process. Experimental investigations by Gardon and Narayanaswamy [71]—in which the authors tracked the evolution of surface temperature and stress-induced birefringence in the middle plane of flat glass plates cooled at a constant rate—revealed that stresses can, in fact, develop during cooling through the transition range, even when the cooling rate is constant. This phenomenon is attributed to structural changes within the glass. Although constant-rate cooling ensures that the final structural state of

both the surface and interior regions is ultimately the same, at any given moment during the cooling process through the transition range, the structural evolution near the surface precedes that in the interior. As a result, transient differences arise between the specific volumes of layers of glass near the surface and in the interior, leading to the development of strains and, consequently, stresses. Due to a rapidly increasing (temperature-dependent) viscosity, the relaxation of these stresses developed during cooling through the transition range is only partial.

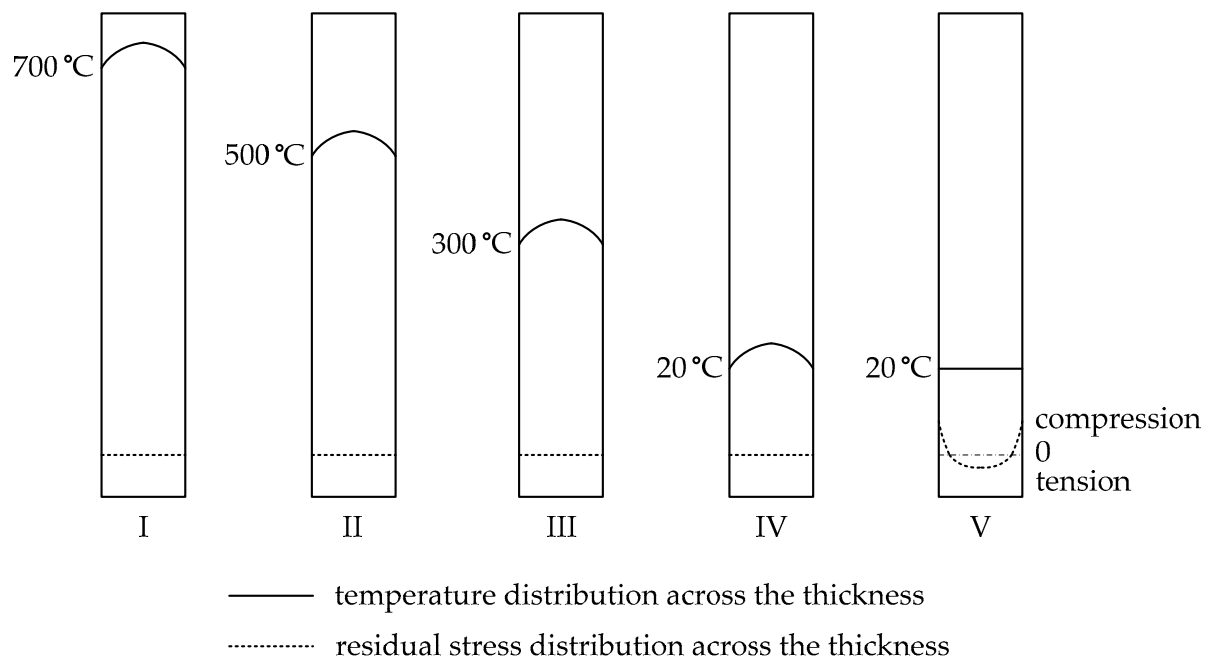


Figure 7. Illustration of the “classical” view on stress generation in a glass ribbon cooled at a constant rate from a sufficiently high temperature, well above the glass transition range (after [70,72])). In this view, stresses develop only when the thermal gradient decays to zero after the glass has reached ambient temperature and responds elastically (stage V).

4.5. Coatings

A variety of functional coatings—e.g., self-cleaning, anti-reflective, and low-emissivity (low-E) coatings—can be applied to the cooling glass ribbon, either in the float bath chamber, between the float bath and the annealinglehr, or inside the annealinglehr itself. This is achieved using on-line chemical vapour deposition techniques, which are compatible with the continuous nature of float glass production. Chemical vapour deposition is a class of synthesis methods in which thin films are formed on a heated substrate via chemical reactions of gas-phase precursors [73]. The resulting coatings are commonly referred to as “hard” coatings due to their exceptional mechanical durability and resistance to abrasion.

Alternatively, coatings may be applied off-line using sputtering techniques, a class of physical vapour deposition methods. In sputtering, atoms or molecules are ejected from a solid source material by bombardment with energetic ions in a vacuum chamber and subsequently deposited onto the glass surface [73]. The resulting sputtered coatings are referred to as “soft” coatings due to their lower mechanical and chemical durability relative to “hard” coatings. As a result, they require careful handling and are typically used in applications where the coated surface is protected from direct exposure to weathering agents.

4.6. The Cold End

The cold end is the final part of the production line. It includes quality control, edge trimming, cutting, and stacking. The float process is continuous and, in theory, can

produce extremely long glass panes. However, for practical reasons, such as transportation, handling, and mounting, a standard format of $600 \times 321 \text{ cm}^2$, known as ribbon size, is used. Alternative lengths, up to 20 m, may be custom-made upon request, albeit at a significant additional cost. The nominal thicknesses of standard float glass range between 2 mm and 25 mm, with tolerances ranging from $\pm 0.2 \text{ mm}$ up to $\pm 1.0 \text{ mm}$ [74].

5. Thermal Tempering

Basic flat glass is further processed to obtain the shape, size, appearance, and performance characteristics required to meet particular needs.

For structural applications, thermal tempering is the most important glass-processing method [75,76]. Float soda–lime–silicate glass plates are heated to a temperature approximately 75°C higher than T_g and then quenched (cooled rapidly) by jets of cold air applied uniformly and symmetrically to both faces (Figure 8). The glass may rest on a ceramic roller conveyor (horizontal process), or it may be held upright and supported either at its bottom edge or by mechanical grippers along the top edge (vertical process). This has the effect of cooling and solidifying first the surface and then the interior of the glass plates (the solidification of a glass plate starts when the surface passes the upper limit of the transformation range and ends when the middle plane passes the lower limit. It is only from this point on that the glass is wholly elastic), inducing in regions away from the edges an isotropic plane stress state (parallel to the middle plane of the plates), with a nearly parabolic through-the-thickness profile. Compressive residual stresses develop near the surface (over a depth of approximately $1/5$ of the plate thickness, which exceeds that of surface flaws), while balancing tensile residual stresses develop in the core (which is essentially flaw-free); the magnitude of the surface compressive stress is approximately twice that of the maximal tensile stress. Compressive stresses also develop in a narrow region along the edges of a plate due to a locally higher surface-to-volume ratio, which causes the edges to cool more rapidly than the interior; for additional details, the reader is referred to [77–80]. It should be recognised that during tempering, the surfaces of a glass plate are usually cooled through the transition range more rapidly than the middle plane. This gives rise to a structural heterogeneity, which, amongst other effects, manifests itself as spatial variations in the (stress-free) density of the glass. These density differences, of structural origin, in turn give rise to stresses [81]. Any stresses generated in the transformation range that are not relaxed by the time the glass emerges from this range (and the rate of stress relaxation decreases rapidly with decreasing temperature, therefore, most rapidly at the surfaces) form a distinct additional component of the final permanent stresses in the glass.

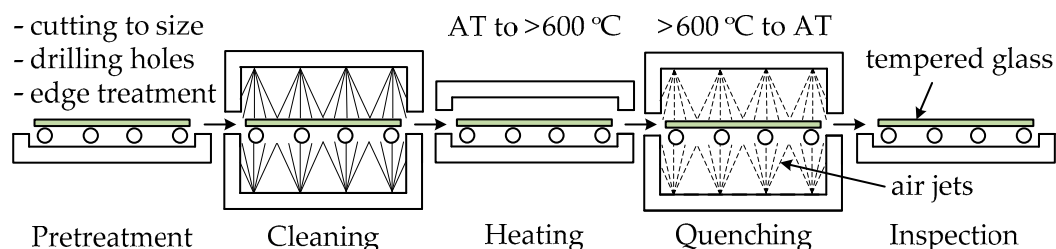


Figure 8. Schematic of the thermal tempering process for float glass, showing the typical temperature ranges during the heating and quenching stages (AT: ambient temperature).

As glass fracture starts from tensile stresses around flaws in the glass surface, putting the surface into compression results in a higher load-bearing capacity; the surface compression has to be overcome before tensile stresses can develop. Another major advantage of the surface compression layer is that it provides significant protection against static fatigue, as the surface only comes into tension under severe loading conditions.

By altering the cooling rate during the quenching phase, the permanent residual stress values can be manipulated (without modifying the depth of the compression layer). Depending on the level of surface compression (Figure 9) and the associated fracture patterns, tempered glass is classified as thermally toughened safety glass or as heat-strengthened glass. The former, also referred to as fully tempered glass, exhibits high levels of surface compressive stresses, which typically vary between 90 MPa and 150 MPa, and fractures into numerous small pieces, the edges of which are generally blunt. The latter, also known as semi-tempered glass, exhibits moderate levels of surface compressive stresses, typically ranging from 35 MPa to 55 MPa, and fractures in a manner similar to annealed glass, with larger fragments than for thermally toughened safety glass. The fracture pattern depends on the elastic energy stored in the glass (from the residual stresses and external actions) at the moment of breakage. Part of this energy is expended in generating fracture surfaces; the greater the stored energy, the smaller the resulting fragments [82]. It should be emphasised that the term “safety glass” has a precise technical meaning; it refers to types of glass that are designed and tested to reduce the risk of injury if they break or if a person collides with them. The designation “safety glass” therefore does not imply unbreakability or superior fracture toughness, but refers primarily to the manner in which the glass behaves upon breakage and how this behaviour mitigates injury risk (see, in this regard, the definition of fracture limit state given in Section 7). Nevertheless, the term may be misleading, since even relatively small glass fragments falling from a height of several metres can cause serious injury.

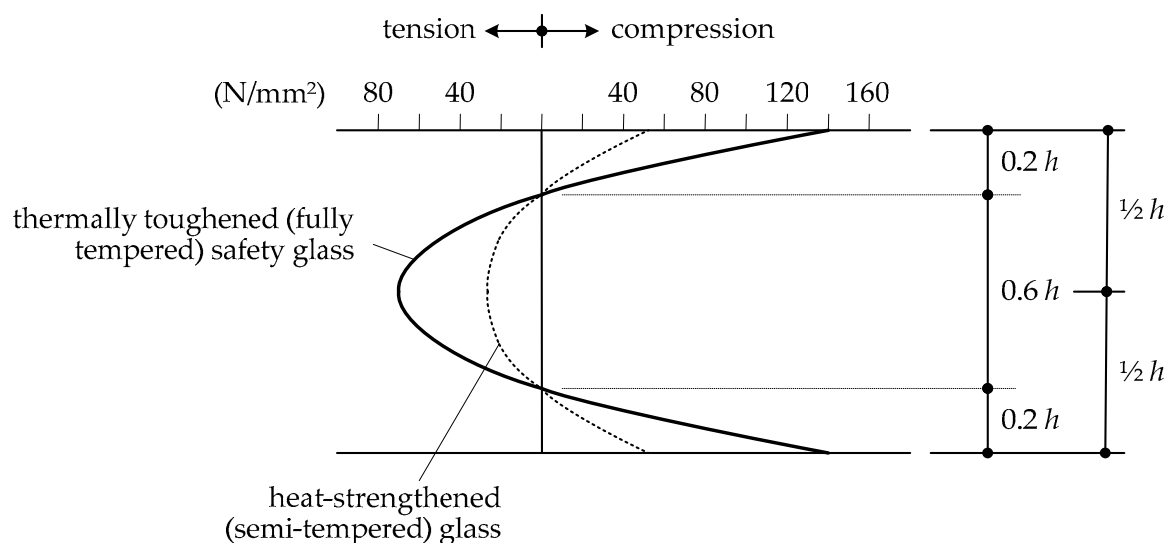


Figure 9. Representative parabolic through-the-thickness residual stress profiles of (symmetrically) heat-strengthened and thermally toughened glass plates in regions away from the edges; h denotes the plate thickness. The maximum residual tensile stresses are approximately half the magnitude of the surface compressive stresses.

Any cutting, drilling, or grinding must be completed prior to thermal tempering. Due to flaws caused by drilling and due to relatively high stress concentrations upon load introduction, point-supported glass panels with holes should always be thermally toughened [83] (cf. clause 7.3.2(1) in Part 1 of the forthcoming Eurocode 10 [3]).

Owing to the very nature of the thermal tempering process, it is not possible to obtain a product as flat as (annealed) float glass. Deviations from flatness may cause discernible distortions in objects observed through, or reflected on, the glass.

There is a practical lower limit of approximately 3 mm to the thickness of soda–lime–silicate glass plates that can be thermally tempered using the conventional process described

above. Below this limit, it becomes difficult to establish the required through-the-thickness temperature gradient and the risk of thermal shock breakage and unacceptable distortion increases. Nevertheless, specialised tempering systems are capable of handling thicknesses down to about 2 mm [76].

Thermally toughened (soda–lime–silicate) safety glass is also known to carry a small but non-negligible risk of spontaneous breakage due to the presence of nickel sulphide (NiS) inclusions. During the heating phase of the tempering process, the NiS particles are transformed into the high-temperature hexagonal α -phase. The subsequent rapid quenching prevents the reverse transformation into the low-temperature rhombohedral β -phase to take place, effectively “freezing” the nickel sulphide in an unstable form. Over a period of time ranging from a few minutes to several years, the α -phase gradually converts back to the β -phase, a transformation that involves an increase in volume [76,84]. Combined with the tensile stresses in the glass core due to tempering, this expansion can lead to sudden breakage with no apparent cause, in a typical “butterfly” fracture pattern. The risk is drastically reduced, though not entirely eliminated, by subjecting the glass to a heat soak procedure before it leaves the factory. The procedure is designed to accelerate the $\alpha \rightarrow \beta$ phase transformation, inducing failure under controlled conditions when dangerous inclusions are present. Although the temperatures during the heat soak procedure remain well below the transition range of commercial soda–lime–silicate glass, the process induces some stress relaxation (of the order of 5 to 10%, according to Schneider et al. [85]) and alters the fracture pattern to one with significantly larger shards, both of which are undesirable side effects. The reliable optical detection of NiS inclusions during glass production is not yet feasible.

6. Chemical Strengthening

There is no single, generally accepted definition of chemical strengthening. The term has been used to designate any method of producing surface compression other than thermal tempering, including the creation, at temperatures above the glass transition temperature, of a surface layer with a lower thermal expansion than the bulk glass [86,87]. Chemically strengthened (alkali-containing) glasses are typically produced by an ion-exchange process in which smaller alkali cations in the glass (e.g., Na^+ in soda–lime–silicate glass) are partially replaced by larger alkali cations (typically K^+) from a bath of molten salt [86,88–93]. The exchange process is confined to a thin layer near the glass surface and is carried out at temperatures at least 50 °C below the strain point [94]. The resulting “stuffing” or “crowding” effect causes the surface layer to expand while constrained by the unaltered glass core, generating high surface compression balanced by interior tension. Using molecular dynamics simulations, Varshneya et al. [95] demonstrated that a K^+ -stuffed soda–lime–silicate glass has a stress-free volume substantially smaller than that of the compositionally equivalent as-melted glass (see also [11]), partially explaining why elastic stuffing models overpredict the magnitude of the surface compressive stress. In addition, ion exchange can increase the thermal expansion coefficient of the surface glass, further reducing the compression upon cooling to room temperature.

The ion-exchange process is an interdiffusion-controlled phenomenon (potassium ions diffuse inward and sodium ions diffuse outward). The depth of the compression layer (or case depth) depends on the glass and molten bath compositions, temperature, and immersion duration; the build-up of surface compressive stress itself impedes the kinetics of ion exchange. For effective strengthening, this depth must exceed the size of surface flaws expected from production, handling, and use (it should be borne in mind that some surface flaws are accessible to salt penetration, whereas others are not [88]). Severe flaws typically have depths of the order of tens of micrometres. For typical commercial soda–lime–silicate

glass compositions, LaCourse [96] reported case depths between 27 μm and 51 μm after immersion in potassium nitrate (KNO_3) bath at 450 $^\circ\text{C}$ for 25 h; achieving a case depth of 50 μm required immersion durations of 24–85 h. Therefore, the process is currently too slow, and thus too costly, to be well-suited for large-scale continuous industrial production of soda–lime–silicate flat glass (LaCourse [96] suggests that the K^+ for Na^+ exchange kinetics in soda–lime–silicate glass can be markedly improved through relatively small compositional changes, though these changes may also impact other glass properties and industrial processing conditions). Moreover, immersion in molten salt imposes practical constraints on the maximum dimensions of glass panes that can be strengthened using this technique. Nevertheless, chemical strengthening introduces minimal optical distortion and enables the strengthening of very thin glass (with a thickness of less than 2 mm), unlike thermal tempering.

A typical through-the-thickness residual stress profile obtained after chemical strengthening is shown in Figure 10. Considerably higher surface compression (usually about 400–600 N/mm^2 in soda–lime–silicate glass, according to Varshneya et al. [95]) than that achievable by thermal tempering can be obtained, but the compression zone extends only to a limited depth. The interior tensile stresses exhibit a fairly uniform distribution, possibly with a slight maximum just beneath the compression zone. The tin side of chemically strengthened float glass has lower surface compression and a smaller case depth than the air side [97]. Because the case depth is shallow, the tensile stresses in the core are low (except when the glass is very thin, with a thickness of the order of a few hundreds of micrometres [89]) and the fracture pattern of chemically strengthened glass is similar to that of annealed glass. It therefore does not, by itself, qualify as a safety glass. The low interior tensile stresses also mean that cutting of chemically strengthened glass is possible to a limited extent, but the freshly cut edge, deprived of compressive stresses, will be no stronger than annealed glass.

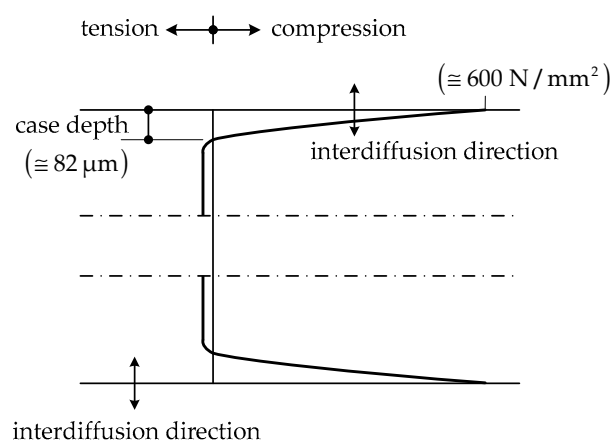


Figure 10. Schematic residual stress profile through the thickness of a chemically strengthened glass plate. The depth of surface compression (case depth) is small and largely independent of the plate thickness. Numerical values shown in parentheses were obtained from photoelastic measurements on a chemically strengthened flat soda–lime–silicate glass specimen by Aben and Guillemet [98].

The strengthening process inherently alters the surface chemistry of the glass, so the procedures for, and the performance of, post-processing steps—such as laminating or coating—can differ from those for non-chemically strengthened glass.

Currently, the use of chemically strengthened soda–lime–silicate glass is restricted to high-value applications where thinness and weight reduction are critical. It is rarely used in buildings. Table 4 presents a comparison of the methods for producing surface compression in glass discussed above.

Table 4. Comparison of different glass prestressing methods.

	Heat-Strengthened (Semi-Tempered)	Thermally Toughened (Fully Tempered)	Chemically Strengthened
Surface compression	35–55 N/mm ²	90–150 N/mm ²	400–600 N/mm ²
Compression depth	~20% of thickness	~20% of thickness	shallow, depends on treatment duration
Fracture pattern	similar to annealed glass	small, generally blunt, fragments	similar to annealed glass
Treatment temperature	~75 °C higher than T_g	~75 °C higher than T_g	at least 50 °C below strain point
Treatment duration	minutes	minutes	hours
Cost	low	low	high
Optical distortion	heating and quenching must be carefully controlled to avoid distortion	heating and quenching must be carefully controlled to avoid distortion	minimal
Minimum thickness	~2 mm (with specialised tempering systems)	~2 mm (with specialised tempering systems)	applicable to very thin articles (<2 mm)
Shape limitations	applicable only to shapes with all surfaces freely accessible to quenching medium	applicable only to shapes with all surfaces freely accessible to quenching medium	applicable to unusual shapes, provided surfaces can contact molten salt
Glass compositions	applicable to most glasses	applicable to most glasses	alkali-containing glasses
Post-processing (cutting, drilling, edge working)	not allowed	not allowed	limited
Standards	[99–101]	[101–105]	[106–108]

7. Implications for Structural Design: The Upcoming Eurocode 10

Glass structures, just like any other structure, are designed to have adequate resistance (including adequate resistance for a required period of time in case of fire), serviceability, durability, and robustness. Moreover, they should be designed to limit the adverse impact on non-renewable environmental resources, on society, and on economy during their entire life cycle. For the purpose of reliability differentiation, glass *components* are assigned to one of the five “consequence classes” defined in Eurocode 0 [109], according to the human-life, socio-economic, and environmental impacts of failure. The provisions of the Eurocodes cover consequence classes CC1 to CC3. For consequence class CC0 (the lowest), either the Eurocodes or alternative provisions may be used—for example, in the case of glass components, those given in the standards EN 16612 [110] or ASTM E1300 [111]. For consequence class CC4 (the highest), additional provisions to those given in the Eurocodes may be required.

Due to the brittleness of the material, the robustness requirement, i.e., the requirement that the structure, during its design service life, shall not be damaged by unforeseen events to an extent disproportionate to the original cause [112], plays a very special role in glass structures. To emphasise this aspect, the design of glass structures according to the new Eurocode 10 shall always consider situations in which the entire glass component, or parts thereof, fractures [3]. These (accidental) design situations are associated with two specific (ultimate) limit states, to be checked by testing or theoretical assessment—the fracture limit state and the post fracture limit state. The fracture limit state is defined as the condition

beyond which, during accidental fracture of a glass component or part of glass component, the following requirements are no longer satisfied: (1) the prevention of injuries caused by glass fragments, (2) the prevention of being passed through by a body, and/or (3) the ability to limit the failure to that glass component or part of glass component. The post fracture limit state is the condition beyond which, in case of accidental failure of a glass component, an adequate residual load bearing capacity, provided by (1) the undamaged plies of that glass component and/or (2) alternative load paths, is no longer maintained. Sufficient safety after fracture is to be verified for a limited period of time [4,5]. For structural applications, glass elements should be made accessible for inspection, included in a regular maintenance schedule, and designed for possible replacement, with documented replacement procedures.

The “classical” serviceability and ultimate limit states are concerned with intact (i.e., unfractured) glass components [3]. It is worth noting that three deformation classes are introduced according to the relevance of deflections—whether they are purely aesthetic; affect the integrity, functionality, or durability of the glass member in the unfractured state (serviceability-related); or affect safety (e.g., leading to loss of support—ultimate limit state-related) [3].

The determination of the glass design bending strength is of the utmost importance for the practising engineer and should reflect current scientific knowledge and industrial practices. In the forthcoming Eurocode 10 (Part 1, Annex A), the design bending strength (which refers to the tensile strength of glass components subjected to moments, in-plane forces or combinations thereof) is given formally as the sum of two terms:

$$f_{g,d} = \lambda_A \lambda_l k_e k_{mod} \frac{k_{sp} f_{g,k}}{\gamma_M} + k_P k_{e,P} \frac{f_{b,k} - f_{g,k}}{k_i \gamma_P} \quad (2)$$

the first term is the design strength of annealed glass, while the second term considers the contribution of surface prestress induced by thermal tempering or chemical strengthening. The meanings of the symbols in the first term are as follows:

- (1) $f_{g,k}$ is the characteristic (5% fractile) bending strength of annealed glass, corresponding to quasi-static loading over a very short time (Table 5).
- (2) k_{sp} is a surface profile factor.
- (3) γ_M is a partial safety factor for resistance; its value depends on the design situation and the consequence class, thereby allowing different target reliability levels [113]. It should be noted that model uncertainties are not explicitly included in γ_M , which is thus not distinguished from γ_m [114].
- (4) k_{mod} is a modification factor that accounts for load duration and thus reflects, in a highly condensed form, the phenomenon of static fatigue (sub-critical crack growth)—cf. Section 3.3. The values of k_{mod} are given in Table 6 and, as a simplification appropriate for a design code, they are independent of environmental conditions. Analytical approaches for the definition of k_{mod} , based on linear elastic fracture mechanics and explicit modelling of sub-critical crack growth (using Equation (1)), are presented in, e.g., [32,53], with Meyland et al. [115] providing a recent comparison of different normative provisions for load-duration effects. The reference duration adopted in the Eurocode, which corresponds to $k_{mod} = 1$, is 3 s.
- (5) k_e is an edge or hole finishing factor, used for verifications near edges or holes under tension. The bending strength at glass edges—plate or hole edges—can be considerably lower than that of the interior, due to damage introduced by the cutting or drilling operation. Glass is typically cut along a straight line in a two-step process. A cutting wheel is used to score the glass surface and initiate a controlled crack. Gentle bending or tapping is then applied to propagate the crack and fracture the

glass along the score, resulting in sharp and uneven edges. Curved cuts can be made using this method but are problematic, as the fracture path often deviates unintentionally from the intended line. Complex shapes are usually obtained by the abrasive waterjet cutting or, less commonly, by laser-assisted thermal or ablation cutting techniques. Holes are generally produced under CNC control using a water-cooled, diamond-tipped hollow drill, with drilling often carried out simultaneously from both sides to minimise chipping at the edges. Where the drills meet, a small burr may form and must be removed by re-drilling or grinding. Cylindrical-conical holes, which allow fasteners to sit flush with the glass surface, are made in two stages, with the conical portion being milled from an initially cylindrical hole. Undercut holes represent another common type, with the undercut being created by eccentric rotation of the tool. Subsequent finishing work, leading to arrised—also known as seamed or chamfered—, ground, smooth-ground, or polished edges, also strongly influences strength, as it affects the size and distribution of surface flaws. It is generally accepted that the strength of annealed glass edges increases with the degree of optical finishing, although recent studies have called this assumption into question [116].

- (6) λ_A and λ_l are the areal and edge-length size effect factors (cf. Section 3.3), which may be taken to be unity unless the area of the pane exceeds 18 m² or the length of an edge exceeds 6 m.

Table 5. Characteristic bending strength of annealed and prestressed soda–lime–silicate glass [3].

Float Glass	Characteristic Bending Strength
Annealed glass [74]	$f_{g,k} = 45 \text{ N/mm}^2$
Heat-strengthened glass [99]	$f_{b,k} = 70 \text{ N/mm}^2$
Thermally toughened safety glass [102] and heat-soaked thermally toughened safety glass [104]	$f_{b,k} = 120 \text{ N/mm}^2$
Chemically strengthened glass [106]	$f_{b,k} = 150 \text{ N/mm}^2$

Table 6. Modification factor k_{mod} [3]. For the characteristic wind load of EN 1991-1-4 [117], a reference duration of 3 s may be adopted ($k_{\text{mod}} = 1$). Other k_{mod} values for dynamic loading (e.g., blast loading) may be derived from a transparent, reproducible, and EN 1990-compliant assessment [109].

Action Classification	Duration	Action	k_{mod}
Permanent	Permanent	Self-weight	0.29
Variable	Intermediate	Snow (3 to 4 weeks) ⁽¹⁾	0.43
		Imposed vertical action (1 week)	0.45
		Temperature change (8 h) ⁽²⁾	0.58
	Short	Maintenance load (30 min)	0.69
		Wind (10 min) ⁽³⁾	0.74
		Barrier personnel loads—crowd (5 min)	0.77
		Barrier personnel loads—normal duty (30 s)	0.89
	Very short	Wind (3 s)	1.00
	Dynamic	Impact (100 ms)	1.20

⁽¹⁾ $k_{\text{mod}} = 0.43$ can be considered representative for snow loads lasting between 5 days ($k_{\text{mod}} = 0.49$) and 3 months ($k_{\text{mod}} = 0.41$); ⁽²⁾ $k_{\text{mod}} = 0.58$ can be considered representative for climatic loads lasting between 6 h ($k_{\text{mod}} = 0.59$) and 12 h ($k_{\text{mod}} = 0.57$); ⁽³⁾ $k_{\text{mod}} = 0.74$ is based on a cumulative equivalent duration of 10 min, considered representative of the effect of a storm, which may last several hours.

It is worth noting that Eurocode 0 explicitly provides for the use of a “conversion factor” when test results must be converted into values representative of in situ behaviour, accounting for, inter alia, scale effects, differences between test and service conditions, and the duration of the effects of actions [109].

The meanings of the additional symbols in the second term on the right-hand side of Equation (2) are as follows:

- (1) $f_{b,k}$ is the characteristic (5% fractile) bending strength of prestressed glass, corresponding to quasi-static loading over a very short time (Table 5).
- (2) γ_P is a partial safety factor for surface prestress, dependent on the design situation and on the consequence class [118]; it accounts for a different set of uncertainties than γ_M .
- (3) k_i is an interference factor that accounts for the (beneficial) statistical interference between the distributions of annealed glass strengths and surface prestress [119,120].
- (4) $k_{e,p}$ is an edge or hole prestressing factor, used for verifications near edges or holes under tension.
- (5) k_P is a prestressing process factor, dependent on the type of prestressing treatment (thermal tempering with horizontal process, thermal tempering with vertical process, or chemical strengthening—cf. Sections 5 and 6); $k_P = 0$ for annealed glass.

When combining actions of different durations, the value of k_{mod} corresponding to the action with the shortest duration should be used. Bedon and Amadio [121] found that, when benchmarked against the theoretical approach of Franco and Royer-Carfagni [122], this simplified rule provides reasonably accurate results for heat-strengthened and thermally toughened safety glass, but can lead to markedly non-conservative predictions for annealed glass. Due to the influence of load duration on strength, combinations of actions not resulting in the maximum tensile stresses may be relevant.

It should be noted that Franco and Royer-Carfagni [122] argue that it would be more transparent to treat prestressing as an external action rather than as an increase in material strength, because static fatigue—accounted for by the modification factor k_{mod} in the first term on the right-hand side of Equation (2)—occurs only once the compressive surface stress induced by prestressing has been overcome.

8. Environmental Aspects

The production of flat soda–lime–silicate glass is a high-temperature and energy-intensive process. Its principal environmental impact arises from atmospheric emissions of CO₂ and other pollutants during melting and fining [55] (recall from Section 4.2 that almost all furnaces in this sector are cross-fired regenerative furnaces fuelled by natural gas—now the predominant energy carrier—and fuel oil, in some cases with an electrical boost of up to 10% of the energy input):

- (1) The combustion of fossil fuels and the decomposition of carbonate raw materials releases CO₂, of which approximately 75% originates from fuel combustion (direct combustion emissions) and the remaining 25% from decarbonation reactions (process emissions).
- (2) The oxidation of sulphur in the fuels and the decomposition/oxidation of sulphates in the batch materials produce sulphur oxides (SO_x), which contribute to acidification.
- (3) Nitrogen oxides (NO_x), formed primarily through high-temperature oxidation of nitrogen in the combustion chamber and, in some cases, from the decomposition of nitrogen compounds in the batch materials, also contribute to acidification and smog formation.

- (4) Furnace emissions additionally contain dust (from the volatilisation and subsequent condensation of volatile batch materials) and traces of chlorides, fluorides, and metals present as impurities in the batch materials and fuels.

The inclusion of cullet in the batch provides considerable environmental benefits [55,58,123,124]:

- (1) Considering typical commercial soda–lime–silicate glass compositions, it is estimated that 1.2 tons of raw materials are saved per ton of cullet used; of the 200 kg difference between cullet input and raw material savings, approximately 150–160 kg corresponds to CO₂ released during melting from carbonate raw materials.
- (2) Remelting cullet requires less energy than melting a batch of raw materials. As a general rule, each 10% increase in cullet by weight in the batch results in a 2.5–3.0% reduction in total glass-melting energy consumption.
- (3) The use of cullet also reduces the demand for raw materials; for example, the production of soda ash via the Solvay process requires approximately 9–13 GJ/t and, in addition to CO₂ emissions, generates significant amounts of aqueous and solid wastes.

These benefits are particularly relevant, as glass can be recycled indefinitely [125]. However, due to stringent quality requirements, the flat glass sector predominantly utilises internally sourced cullet, which typically constitutes around 20–30% of the batch weight [126]; since virtually all internally sourced cullet is already reused, further substantial increases in cullet share are unlikely in the near future.

Over the years, improvements in furnace design and operation, as well as in refractory materials, have brought about a remarkable increase in the energy efficiency of glass-melting processes—Jatzwauk [59] reports a tenfold increase since the 1880s. However, further gains now appear marginal, suggesting that the intrinsic limits of current technologies are being approached.

The emissions of NO_x and SO_x are a more recent concern. According to Jatzwauk [59], companies such as AGC (Asahi Glass Company) have been able to reduce these emissions by roughly a factor of four over the past two decades through improvements in furnace design and operation, selective catalytic reduction, SO_x scrubbing systems (which produce significant quantities of by-products, mainly sulphates, that are recycled as raw materials whenever feasible), and the careful selection of raw materials [55]. Emissions of dust have followed a similar downward trend.

Promising options for further decarbonisation of flat glass production and reduction in dust and pollutant emissions, currently under investigation or undergoing trials, include (1) hot oxy-combustion, which uses purified oxygen instead of air for combustion and recovers heat from flue gases to pre-heat both the natural gas and oxygen streams before entering the furnace, potentially combined with pelletised batch (with low fractions of cullet) pre-heating; (2) the use of carbon-neutral gaseous energy carriers as substitutes for natural gas; and (3) the adoption of all-electric and hybrid (electric-combustion) melting technologies [58,127,128]. Carbon capture, utilisation, and storage (CCUS) is another potential avenue for research and development [129]. The technological and economic viability of these measures and their cross-media effects require careful assessment.

Oxy-fuel combustion is now a proven technology. Conversion from an air-fuel to an oxy-fuel furnace typically reduces combustion-generated NO_x emissions by 80–90%, depending on oxygen purity, batch and fuel nitrogen levels, and burner design; particulate emissions can also be reduced by up to 30%. In addition, available data indicate that oxy-fuel combustion can have a positive impact on glass quality while eliminating the capital and maintenance costs associated with regenerators [130]. The main cross-media effects of oxy-fuel combustion are related to the electricity required for oxygen purification [55]. A high share of renewable energy in the electricity mix is important for reducing associated

CO₂ emissions. As of 2019, only around ten float glass furnaces had reportedly been converted to oxy-fuel combustion, mainly due to the additional and uncertain cost of oxygen supply [130]. In oxy-fuel-fired glass furnaces, approximately 25% of the fuel energy input is lost as sensible heat in the flue gases [58]. An innovative and promising approach to improving the energy efficiency of oxy-fuel combustion technology is the application of waste heat recovery systems to pre-heat oxygen and natural gas. When purified oxygen and natural gas are pre-heated to around 500 °C, an energy-efficiency gain of approximately 10% has been reported [131]. Waste heat recovery systems, preferably designed to operate with flue gases that are not cooled or only slightly cooled, can also be used to pre-heat the batch [55,132,133]. Another promising option for oxy-fuel processes is the application of thermo-chemical recuperators to convert, in a highly endothermic reaction, natural gas into a high-calorific pre-heated fuel, syngas, with achievable energy savings of 20% or more [133,134].

In order to reach the target of carbon-neutral glass production, the industry will have to switch to alternative energy carriers (biogas, synthetic methane, green hydrogen, or renewable electricity) as substitutes for fossil fuels. Notwithstanding economic and distribution challenges, a switch to biogas is, in principle, technically feasible. Under appropriately adjusted process conditions, no significant negative impacts on combustion behaviour, refractory materials, and product quality have been reported on test rig experiments. However, biogas is a limited resource, and fluctuating gas compositions are undesirable from a glass manufacturer's perspective, as they may lead to quality losses or reduced pull rates. A co-firing approach, in which natural gas is partially substituted by biogas rather than fully replaced, may therefore represent a more appropriate solution [135].

The feasibility of substituting natural gas with green hydrogen (i.e., hydrogen produced by the electrolysis of water using renewable electricity) for firing float glass furnaces is currently the subject of intensive investigation. According to Kobayashi [130], no fundamental technical barriers have been identified that would preclude the use of hydrogen for glass melting, provided that furnace and burner designs, as well as batch composition, are appropriately adapted; oxygen-firing is likely to be favoured over air-firing. Beyond combustion-related issues, the large-scale use of hydrogen requires substantial modifications to plant infrastructure, including safety systems, in compliance with relevant standards and regulations. One potential advantage of hydrogen, at least in Europe, is the anticipated development of a dedicated hydrogen supply infrastructure as demand increases [58].

All-electric melting represents another pathway towards decarbonisation and offers several advantages over combustion-based melting [55]. Electric furnaces can typically achieve higher melt rates per unit furnace area and thermal efficiencies that are two to three times higher than those of fossil-fuel-fired furnaces, when assessed on the basis of energy delivered to the furnace rather than the primary energy required to generate the electricity. In addition, all-electric melting produces no direct combustion-related CO₂, thermal NO_x, or SO_x emissions. It is generally associated with reduced volatilisation of raw materials, primarily due to the reduced gas flow, and this simplifies exhaust gas treatment. Regenerators and high-temperature furnace crowns are not required. However, the prevailing high cost of electricity relative to fossil fuels currently limits the economic viability of all-electric melting to small-scale applications, up to about 250 t/day, which is well below the output of an average float line. Whether upscaling of this technology without extensive changes to the production processes is feasible remains a subject of debate [58,136]. It must also be recognised that currently established all-electric melting systems offer limited operational flexibility with respect to the integration of electricity from renewable sources, since a highly stable heat input is essential. The comparatively

shorter service life of electric melting tanks is also a concern. To mitigate some of these disadvantages, hybrid melting combines a predominantly electrically heated furnace with the use of oxy-fuel combustion as a support, increasing operational flexibility and resilience to electricity price fluctuations and potential supply constraints.

In addition to the environmental improvements already achieved in the manufacture of flat glass products and those expected in the foreseeable future, it is important to consider the benefits associated with their use. The reduction in CO₂ emissions resulting from the use of energy-efficient glass products over their operational lifetime far outweighs the emissions associated with their manufacture [55,127]. Indeed, in applications such as building glazing, the flat glass industry can already be regarded as a net carbon avoidance activity. Flat glass products are also essential in supporting the transition to clean mobility and in contributing to the efficient generation of renewable solar energy [124]. Moreover, in line with the principles of circularity, secondary structural glass elements can be designed for easy removal at the end of a building's life and potential reuse, supported by appropriate inspection and traceability systems to ensure reliability and quality compliance. As of 2023, only around 5% of end-of-life construction and demolition glass from buildings was effectively recycled [124]. Closed-loop recycling of flat glass diverts waste from landfill, conserves raw materials, and is one of the most effective means of reducing energy consumption and CO₂ emissions from glass manufacturing.

9. Conclusions

Eurocode 10, a forthcoming addition to the Eurocode family, provides the first harmonised framework for the design of glass structures across Europe and is expected to promote the adoption of glass as a structural material, with potential influence beyond the European context. It aligns with the overarching principles of EN 1990, while considering the particular mechanical and physical behaviour of glass, a brittle and flaw-sensitive material that can experience delayed fracture under sustained load in the presence of water. By introducing novel limit states—the fracture and post-fracture limit states—and consistent verification procedures, Eurocode 10 establishes a design philosophy that is fundamentally rooted in material behaviour rather than past experience alone. This review has sought to support this shift by bringing together key elements of glass science and technological processes with practical design considerations, underscoring their fundamental connection.

Within the semi-probabilistic approach to structural reliability defined in EN 1990, the Eurocode-based determination of the glass design bending strength is of central importance for practising engineers and was therefore given particular attention in this review. The various factors affecting the characteristic values $f_{g,k}$ and $f_{b,k}$ were explicitly linked to underlying fracture-mechanical phenomena, manufacturing and processing events, environmental effects, and statistical considerations. This approach promotes a deeper and more objective appreciation of the nature and purpose of these factors, while enabling a critical perspective on some of the assumptions and simplifications on which they are based.

Environmental considerations, including ongoing advances in energy efficiency, decarbonisation, and reduction in pollutant emissions, were also examined. This synthesis supports structural engineers in making more informed and responsible decisions in the design, detailing, and specification of glass structures and contributes to the continued development of glass as a reliable, innovative, and sustainable structural material.

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References

1. Rasmussen, S.C. *How Glass Changed the World—The History and Chemistry of Glass*; Springer: Heidelberg, Germany, 2012.
2. Feldmann, M.; Kasper, R.; Abeln, B.; Cruz, P.; Belis, J.; Beyer, J.; Colvin, J.; Ensslen, F.; Eliasova, M.; Galuppi, L.; et al. *Guidance for European Structural Design of Glass Components*; Publications Office of the European Union: Luxembourg, 2014.
3. *prEN 19100-1:2023*; Eurocode 10—Design of Glass Structures—Part 1: General Rules. European Committee for Standardization: Brussels, Belgium, 2023.
4. *prEN 19100-2:2023*; Eurocode 10—Design of Glass Structures—Part 2: Out-of-Plane Loaded Glass Components. European Committee for Standardization: Brussels, Belgium, 2023.
5. *prEN 19100-3:2023*; Eurocode 10—Design of Glass Structures—Part 3: In-Plane Loaded Glass Components. European Committee for Standardization: Brussels, Belgium, 2023.
6. *FprCEN/TS 19100-4:2023*; Design of Glass Structures—Part 4: Glass Selection Relating to the Risk of Human Injury—Guidance for Specification. European Committee for Standardization: Brussels, Belgium, 2023.
7. Feldmann, M.; Laurs, M.; Belis, J.; Buljan, N.; Criaud, A.; Dupont, E.; Eliasova, M.; Galuppi, L.; Hassinen, P.; Kasper, R.; et al. The new CEN/TS 19100: Design of glass structures. *Glass Struct. Eng.* **2023**, *8*, 317–337. [\[CrossRef\]](#)
8. Regulation (EU) 2024/3110 of the European Parliament and of the Council of 27 November 2024 Laying Down Harmonised Rules for the Marketing of Construction Products and Repealing Regulation (EU) No 305/2011, Official Journal of the European Union, L Series. 2024. Available online: <http://data.europa.eu/eli/reg/2024/3110/oj> (accessed on 10 January 2026).
9. *ASTM C162-23*; Standard Terminology of Glass and Glass Products. ASTM International: West Conshohocken, PA, USA, 2023.
10. Conradt, R. Thermodynamics and kinetics of glass. In *Springer Handbook of Glass*; Musgraves, J.D., Hu, J., Calvez, L., Eds.; Springer: Cham, Switzerland, 2019; pp. 51–77.
11. Varshneya, A.K.; Mauro, J.C. *Fundamentals of Inorganic Glasses*, 3rd ed.; Elsevier: Amsterdam, The Netherlands, 2019.
12. Ediger, M.D.; Angell, C.A.; Nagel, S.R. Supercooled liquids and glasses. *J. Phys. Chem.* **1996**, *100*, 13200–13212. [\[CrossRef\]](#)
13. Scholze, H. *Glass—Nature, Structure, and Properties*; Springer: New York, NY, USA, 1991.
14. Shelby, J.E. *Introduction to Glass Science and Technology*, 2nd ed.; The Royal Society of Chemistry: Cambridge, UK, 2005.
15. International Union of Pure and Applied Chemistry. Glass-transition temperature. In *IUPAC Compendium of Chemical Terminology*, 5th ed.; International Union of Pure and Applied Chemistry: Raleigh, NC, USA, 2025; Online version 5.0.0. [\[CrossRef\]](#)
16. Hand, R.J. Soda-lime-silica glasses. In *Encyclopedia of Materials—Technical Ceramics and Glasses*; Pomeroy, M., Ed.; Elsevier: Amsterdam, The Netherlands, 2021; Volume 2, pp. 483–495.
17. Zallen, R. *The Physics of Amorphous Solids*; Wiley: Weinheim, Germany, 2004.
18. Napolitano, A.; Hawking, E.G. Viscosity of a standard soda-lime-silica glass. *J. Res. Natl. Bur. Stand. A Phys. Chem.* **1964**, *68*, 439–448. [\[CrossRef\]](#)
19. *ISO 7884-2:1987*; Glass—Viscosity and Viscometric Fixed Points—Part 2: Determination of Viscosity by Rotation Viscometers. International Organization for Standardization: Geneva, Switzerland, 1987.
20. *ISO 7884-5:1987*; Glass—Viscosity and Viscometric Fixed Points—Part 5: Determination of Working Point by Sinking Bar Viscometer. International Organization for Standardization: Geneva, Switzerland, 1987.
21. *ISO 7884-6:1987*; Glass—Viscosity and Viscometric Fixed Points—Part 6: Determination of Softening Point. International Organization for Standardization: Geneva, Switzerland, 1987.
22. *ASTM C338-24*; Standard Test Method for Softening Point of Glass. ASTM International: West Conshohocken, PA, USA, 2024.
23. *ISO 7884-7:1987*; Glass—Viscosity and Viscometric Fixed Points—Part 7: Determination of Annealing Point and Strain Point by Beam Bending. International Organization for Standardization: Geneva, Switzerland, 1987.
24. *ASTM C598-24*; Standard Test Method for Annealing Point and Strain Point of Glass by Beam Bending. ASTM International: West Conshohocken, PA, USA, 2024.
25. Conradt, R. Glass production: An overview. In *Encyclopedia of Glass Science, Technology, History, and Culture*; Richet, P., Ed.; Wiley: Hoboken, NJ, USA, 2021; Volume 1, pp. 25–37.
26. Doremus, R.H. *Glass Science*, 2nd ed.; Wiley: New York, NY, USA, 1994.
27. Deng, W.; Wakelin, E.; Kilinc, E.; Bingham, P.A. A survey of commercial soda-lime-silica glass compositions: Trends and opportunities I—Compositions, properties and theoretical energy requirements. *Int. J. Appl. Glass Sci.* **2025**, *16*, e16691. [\[CrossRef\]](#)
28. Letz, M. Linear optical properties. In *Springer Handbook of Glass*; Musgraves, J.D., Hu, J., Calvez, L., Eds.; Springer: Cham, Switzerland, 2019; pp. 169–191.

29. EN 572-1:2012+A1:2016; Glass in Building—Basic Soda-Lime Silicate Glass Products—Part 1: Definitions and General Physical and Mechanical Properties. European Committee for Standardization: Brussels, Belgium, 2016.
30. Brückner-Foit, A.; Fett, T.; Schirmer, K.-S.; Munz, D. Discrimination of multiaxiality criteria using brittle fracture loci. *J. Eur. Ceram. Soc.* **1996**, *16*, 1201–1207. [[CrossRef](#)]
31. Haldimann, M. Fracture Strength of Structural Glass Elements—Analytical and Numerical Modelling, Testing and Design. Ph.D. Thesis, École Polytechnique Fédérale de Lausanne, Lausanne, Switzerland, 2006.
32. Badalassi, M.; Biolzi, L.; Royer-Carfagni, G.; Salvatore, W. Safety factors for the structural design of glass. *Constr. Build. Mater.* **2014**, *55*, 114–127. [[CrossRef](#)]
33. Pisano, G.; Bonati, A.; Royer-Carfagni, G. The effect of size and stress state on the strength of architectural glass. *Experiments versus theory. Constr. Build. Mater.* **2021**, *283*, 122635. [[CrossRef](#)]
34. Lawn, B. *Fracture of Brittle Solids*, 2nd ed.; Cambridge University Press: Cambridge, UK, 1993.
35. Gy, R. Stress corrosion of silicate glass: A review. *J. Non-Cryst. Solids* **2003**, *316*, 1–11. [[CrossRef](#)]
36. Ciccotti, M. Stress-corrosion mechanisms in silicate glasses. *J. Phys. D Appl. Phys.* **2009**, *42*, 214006. [[CrossRef](#)]
37. Freiman, S.W.; Wiederhorn, S.M.; Mecholsky, J.J. Environmentally enhanced fracture of glass: A historical perspective. *J. Am. Ceram. Soc.* **2009**, *92*, 1371–1382. [[CrossRef](#)]
38. Guin, J.-P.; Gueguen, Y. Mechanical properties of glass. In *Springer Handbook of Glass*; Musgraves, J.D., Hu, J., Calvez, L., Eds.; Springer: Cham, Switzerland, 2019; pp. 227–271.
39. Wiederhorn, S.M.; Johnson, H.; Diness, A.M.; Heuer, A.H. Fracture of glass in vacuum. *J. Am. Ceram. Soc.* **1974**, *57*, 336–341. [[CrossRef](#)]
40. Michalske, T.A.; Freiman, S.W. A molecular mechanism for stress corrosion in vitreous silica. *J. Am. Ceram. Soc.* **1983**, *70*, 780–784. [[CrossRef](#)]
41. Michalske, T.A.; Bunker, B.C. Slow fracture model based on strained silicate structures. *J. Appl. Phys.* **1984**, *56*, 2686–2693. [[CrossRef](#)]
42. Evans, A.G.; Wiederhorn, S.M. Proof testing of ceramic materials—An analytical basis for failure prediction. *Int. J. Fract.* **1974**, *10*, 379–392. [[CrossRef](#)]
43. Gehrke, E.; Ullner, C.; Hähner, M. Fatigue limit and crack arrest in alkali-containing silicate glasses. *J. Mater. Sci.* **1991**, *26*, 5445–5455. [[CrossRef](#)]
44. Wiederhorn, S.M.; Dretzke, A.; Rödel, J. Crack growth in soda-lime-silicate glass near the static fatigue limit. *J. Am. Ceram. Soc.* **2002**, *85*, 2287–2292. [[CrossRef](#)]
45. Guin, J.-P.; Wiederhorn, S.M. Crack growth threshold in soda lime silicate glass: Role of hold-time. *J. Non-Cryst. Solids* **2003**, *316*, 12–20. [[CrossRef](#)]
46. Wiederhorn, S.M. Influence of water vapor on crack propagation in soda-lime glass. *J. Am. Ceram. Soc.* **1967**, *50*, 407–414. [[CrossRef](#)]
47. Wiederhorn, S.M.; Bolz, L.H. Stress corrosion and static fatigue of glass. *J. Am. Ceram. Soc.* **1970**, *53*, 543–548. [[CrossRef](#)]
48. Wan, K.-T.; Lathabai, S.; Lawn, B.R. Crack velocity functions and thresholds in brittle solids. *J. Eur. Ceram. Soc.* **1990**, *6*, 259–268. [[CrossRef](#)]
49. Wiederhorn, S.M.; Johnson, H. Effect of electrolyte pH on crack propagation in glass. *J. Am. Ceram. Soc.* **1973**, *56*, 192–197. [[CrossRef](#)]
50. Gehrke, E.; Ullner, C.; Hähner, M. Effect of corrosive media on crack growth of model glasses and commercial silicate glasses. *Glastech. Berichte* **1990**, *63*, 255–265.
51. Schula, S. *Charakterisierung der Kratzanfälligkeit von Gläsern im Bauwesen*; Springer: Berlin/Heidelberg, Germany, 2015.
52. Brokmann, C.; Kolling, S.; Schneider, J. Subcritical crack growth parameters in glass as a function of environmental conditions. *Glass Struct. Eng.* **2021**, *6*, 89–101. [[CrossRef](#)]
53. Consiglio Nazionale delle Ricerche. *Guide for the Design, Construction and Control of Buildings with Structural Glass Elements (CNR-DT 210)*; Consiglio Nazionale delle Ricerche: Rome, Italy, 2013.
54. Swarts, E.L. The melting of glass. In *Introduction to Glass Science*; Pye, L.D., Stevens, H.J., LaCourse, W.C., Eds.; Plenum Press: New York, NY, USA, 1972; pp. 273–327.
55. Scalet, B.M.; Garcia Muñoz, M.; Sissa, A.Q.; Roudier, S.; Delgado Sancho, L. *Best Available Techniques (BAT) Reference Document for the Manufacture of Glass—Industrial Emissions Directive 2010/75/EU (Integrated Pollution Prevention and Control)*; Publications Office of the European Union: Luxembourg, 2013.
56. Hubert, M. Industrial glass processing and fabrication. In *Springer Handbook of Glass*; Musgraves, J.D., Hu, J., Calvez, L., Eds.; Springer: Cham, Switzerland, 2019; pp. 1193–1229.
57. Di Pierro, S. Raw materials for glassmaking: Properties and constraints. In *Encyclopedia of Glass Science, Technology, History, and Culture*; Richet, P., Ed.; Wiley: Hoboken, NJ, USA, 2021; Volume 1, pp. 39–51.

58. Zier, M.; Stenzel, P.; Kotzur, L.; Stolten, D. A review of decarbonisation options for the glass industry. *Energy Convers. Manag.* **2021**, *10*, 100083.
59. Jatzwauk, C. Design and operation of glass furnaces. In *Encyclopedia of Glass Science, Technology, History, and Culture*; Richet, P., Ed.; Wiley: Hoboken, NJ, USA, 2021; Volume 2, pp. 1147–1164.
60. Verheijen, O.S.; Hubert, M. Batch chemistry and reactions. In *Springer Handbook of Glass*; Musgraves, J.D., Hu, J., Calvez, L., Eds.; Springer: Cham, Switzerland, 2019; pp. 1231–1256.
61. Beerkens, R. Fining of glass melts: What we know about fining processes today. In *69th Conference on Glass Problems, Columbus, OH, USA, 4–5 November 2008*; Drummond, C.H., Ed.; Wiley: Hoboken, NJ, USA, 2009; pp. 13–28.
62. Conradt, R. Fusion of glass. In *Encyclopedia of Glass Science, Technology, History, and Culture*; Richet, P., Ed.; Wiley: Hoboken, NJ, USA, 2021; Volume 1, pp. 53–66.
63. Pilkington, L.A.B. The float glass process. *Proc. R. Soc. A* **1969**, *314*, 1–25.
64. Kamihori, T. Primary fabrication of flat glass. In *Encyclopedia of Glass Science, Technology, History, and Culture*; Richet, P., Ed.; Wiley: Hoboken, NJ, USA, 2021; Volume 1, pp. 67–79.
65. Charnock, H. The float glass process. *Phys. Bull.* **1970**, *21*, 153–156. [[CrossRef](#)]
66. Edge, C.K. Flat glass manufacturing processes (update). In *The Handbook of Glass Manufacture*, 3rd ed.; Tooley, F.V., Ed.; Ashlee Publishing Co.: New York, NY, USA, 1984; Volume 2, pp. 714/1–714/21.
67. Hynd, W.C. Flat glass manufacturing processes. In *Glass: Science and Technology*; Uhlmann, D.R., Kreidl, N.J., Eds.; Academic Press: New York, NY, USA, 1984; Volume 2, pp. 45–106.
68. Laniel, R.; Hubert, M.; Miroir, M.; Brient, A. Glass shaping. In *Springer Handbook of Glass*; Musgraves, J.D., Hu, J., Calvez, L., Eds.; Springer: Cham, Switzerland, 2019; pp. 1257–1290.
69. Sieger, J.S. Chemical characteristics of float glass surfaces. *J. Non-Cryst. Solids* **1975**, *19*, 213–220. [[CrossRef](#)]
70. Adams, L.H.; Williamson, E.D. The annealing of glass. *J. Frank. Inst.* **1920**, *190*, 597–631. [[CrossRef](#)]
71. Gardon, R.; Narayanaswamy, O.S. Stress and volume relaxation in annealing flat glass. *J. Am. Ceram. Soc.* **1970**, *53*, 380–385. [[CrossRef](#)]
72. Narayanaswamy, O.S. Annealing of glass. In *Glass: Science and Technology*; Uhlmann, D.R., Kreidl, N.J., Eds.; Academic Press: New York, NY, USA, 1986; Volume 3, pp. 275–318.
73. Pulker, H.K. *Coatings on Glass*, 2nd ed.; Elsevier: Amsterdam, The Netherlands, 1999.
74. EN 572-2:2012; Glass in Building—Basic Soda-Lime Silicate Glass Products—Part 2: Float Glass. European Committee for Standardization: Brussels, Belgium, 2012.
75. Gardon, R. Thermal tempering of glass. In *Glass: Science and Technology*; Uhlmann, D.R., Kreidl, N.J., Eds.; Academic Press: New York, NY, USA, 1980; Volume 5, pp. 145–216.
76. Karlsson, K.S.R.; Wondraczek, L. Strengthening of oxide glasses. In *Encyclopedia of Glass Science, Technology, History, and Culture*; Richet, P., Ed.; Wiley: Hoboken, NJ, USA, 2021; Volume 1, pp. 391–404.
77. Daudeville, L.; Carré, H. Thermal tempering simulation of glass plates: Inner and edge residual stresses. *J. Therm. Stress.* **1998**, *21*, 667–689. [[CrossRef](#)]
78. Laufs, W.; Sedlacek, G. Stress distribution in thermally tempered glass panes near the edges, corners and holes. Part 2: Distribution of thermal stresses. *Glastech. Berichte* **1999**, *72*, 42–48.
79. Pourmoghaddam, N.; Schneider, J. Finite-element analysis of the residual stresses in tempered glass plates with holes or cut-outs. *Glass Struct. Eng.* **2018**, *3*, 17–37. [[CrossRef](#)]
80. Efferz, L.; Schuler, C.; Siebert, G. Photoelastic measurements and evaluation methods for edge stress in architectural tempered glass. *Glass Struct. Eng.* **2024**, *9*, 439–454. [[CrossRef](#)]
81. Narayanaswamy, O.S. Stress and structural relaxation in tempering glass. *J. Am. Ceram. Soc.* **1978**, *61*, 146–152. [[CrossRef](#)]
82. Barsom, J.M. Fracture of tempered glass. *J. Am. Ceram. Soc.* **1968**, *51*, 75–78. [[CrossRef](#)]
83. Belis, J.; Louter, C.; Nielsen, J.H.; Schneider, J. Architectural glass. In *Springer Handbook of Glass*; Musgraves, J.D., Hu, J., Calvez, L., Eds.; Springer: Cham, Switzerland, 2019; pp. 1779–1817.
84. Swain, M.V. Nickel sulphide inclusions in glass: An example of microcracking induced by a volumetric expanding phase change. *J. Mater. Sci.* **1981**, *16*, 151–158. [[CrossRef](#)]
85. Schneider, J.; Hilcken, J.; Aronen, A.; Karvinen, R.; Olesen, J.F.; Nielsen, J. Stress relaxation in tempered glass caused by heat soak testing. *Eng. Struct.* **2016**, *122*, 42–49. [[CrossRef](#)]
86. Bartholomew, R.F.; Garfinkel, H.M. Chemical strengthening of glass. In *Glass: Science and Technology*; Uhlmann, D.R., Kreidl, N.J., Eds.; Academic Press: New York, NY, USA, 1980; Volume 5, pp. 217–270.
87. Ernsberger, F.M. Techniques of strengthening glasses. In *Glass: Science and Technology*; Uhlmann, D.R., Kreidl, N.J., Eds.; Academic Press: New York, NY, USA, 1980; Volume 5, pp. 133–144.
88. Varshneya, A.K. Chemical strengthening of glass products. *Trans. Indian Ceram. Soc.* **2001**, *60*, 1–6. [[CrossRef](#)]
89. Gy, R. Ion exchange for glass strengthening. *Mater. Sci. Eng. B* **2008**, *149*, 159–165. [[CrossRef](#)]

90. Karlsson, S.; Jonson, B.; Stålhandske, C. The technology of chemical glass strengthening—A review. *Glass Technol.-Eur. J. Glass Sci. Technol. A* **2010**, *51*, 41–54.
91. Varshneya, A.K. Chemical strengthening of glass: Lessons learned and yet to be learned. *Int. J. Appl. Glass Sci.* **2010**, *1*, 131–142. [[CrossRef](#)]
92. Mazzoldi, P.; Carturan, S.; Quaranta, A.; Sada, C.; Sglavo, V.M. Ion exchange process: History, evolution and applications. *Riv. Del Nuovo C.* **2013**, *36*, 397–460.
93. Gross, T.M. Chemical strengthening of glass. In *Springer Handbook of Glass*; Musgraves, J.D., Hu, J., Calvez, L., Eds.; Springer: Cham, Switzerland, 2019; pp. 273–296.
94. LaCourse, W.; Kasprzyk, J.; Moulton, B.J.A. Prospects for ion-exchange processing of commercial soda-lime-silica glasses. *Am. Ceram. Soc. Bull.* **2025**, *104*, 22–27.
95. Varshneya, A.K.; Olson, G.A.; Kreski, P.K.; Gupta, P.K. Buildup and relaxation of stress in chemically strengthened glass. *J. Non-Cryst. Solids* **2015**, *427*, 91–97. [[CrossRef](#)]
96. LaCourse, W.C. Design of SLS compositions for accelerated chemical strengthening. In *78th Conference on Glass Problems, Columbus, OH, USA, 6–9 November 2017*; Sundaram, S.K., Ed.; Wiley: Hoboken, NJ, USA, 2018; pp. 181–189.
97. Gomez, S.; Dejneka, M.J.; Ellison, A.J.; Rossington, K.R. A look at the chemical strengthening process: Alkali aluminosilicate glasses vs. soda-lime glass. In *71st Conference on Glass Problems, Columbus, OH, USA, 19–20 October 2010*; Drummond, C.H., Ed.; Wiley: Hoboken, NJ, USA, 2011; pp. 61–66.
98. Aben, H.; Guillet, C. *Photoelasticity of Glass*; Springer: Berlin/Heidelberg, Germany, 1993.
99. EN 1863-1:2011; Glass in Building—Heat Strengthened Soda Lime Silicate Glass—Part 1: Definition and Description. European Committee for Standardization: Brussels, Belgium, 2011.
100. EN 1863-2:2004; Glass in Building—Heat Strengthened Soda Lime Silicate Glass—Part 2: Evaluation of Conformity/Product Standard. European Committee for Standardization: Brussels, Belgium, 2004.
101. ASTM C1048-25; Standard Specification for Heat-Strengthened and Fully Tempered Flat Glass. ASTM International: West Conshohocken, PA, USA, 2025.
102. EN 12150-1:2015+A1:2019; Glass in Building—Thermally Toughened Soda Lime Silicate Safety Glass—Part 1: Definition and Description (EN 12150-1:2015+A1:2019). European Committee for Standardization: Brussels, Belgium, 2019.
103. EN 12150-2:2004; Glass in Building—Thermally Toughened Soda Lime Silicate Safety Glass—Part 2: Evaluation of Conformity/Product Standard. European Committee for Standardization: Brussels, Belgium, 2004.
104. EN 14179-1:2016; Glass in Building—Heat Soaked Thermally Toughened Soda Lime Silicate Safety Glass—Part 1: Definition and Description. European Committee for Standardization: Brussels, Belgium, 2016.
105. EN 14179-2:2005; Glass in Building—Heat Soaked Thermally Toughened Soda Lime Silicate Safety Glass—Part 2: Evaluation of Conformity/Product Standard. European Committee for Standardization: Brussels, Belgium, 2005.
106. EN 12337-1:2000; Glass in Building—Chemically Strengthened Soda Lime Silicate Glass—Part 1: Definition and Description. European Committee for Standardization: Brussels, Belgium, 2000.
107. EN 12337-2:2004; Glass in Building – Chemically Strengthened Soda Lime Silicate Glass—Part 2: Evaluation of Conformity/Product Standard. European Committee for Standardization: Brussels, Belgium, 2004.
108. ASTM C1422/C1422M-20a; Standard Specification for Chemically Strengthened Flat Glass. ASTM International: West Conshohocken, PA, USA, 2020.
109. EN 1990:2023; Eurocode—Basis of Structural and Geotechnical Design. European Committee for Standardization: Brussels, Belgium, 2023.
110. EN 16612:2019; Glass in Building—Determination of the Lateral Load Resistance of Glass Panes by Calculation. European Committee for Standardization: Brussels, Belgium, 2019.
111. ASTM E1300-24; Standard Practice for Determining Load Resistance of Glass in Buildings. ASTM International: West Conshohocken, PA, USA.
112. Knoll, F.; Vogel, T. *Design for Robustness*; International Association for Bridge and Structural Engineering (IABSE): Zürich, Switzerland, 2009.
113. Ballarini, R.; Pisano, G.; Royer-Carfagni, G. New calibration of partial material factors for the structural design of float glass. Comparison of bounded and unbounded statistics for glass strength. *Constr. Build. Mater.* **2016**, *121*, 69–80. [[CrossRef](#)]
114. Vrouwenvelder, T.; Dimova, S.; Sousa, M.L.; Marková, J.; Mancini, G.; Kuhlmann, U.; Taras, A.; Jockwer, R.; Jäger, W.; Schweckendiek, T.; et al. *Reliability Background of the Eurocodes*; Publications Office of the European Union: Luxembourg, 2024.
115. Meyland, M.J.; Nielsen, J.H.; Kocer, C. Tensile behaviour of soda-lime-silica glass and the significance of load duration—A literature review. *J. Build. Eng.* **2021**, *44*, 102966. [[CrossRef](#)]
116. Bukieda, P.; Meyer, K.; Engelmann, M.; Weller, B. Glass edges in the building industry: Examination method and design approaches. *Glass Struct. Eng.* **2024**, *9*, 419–437. [[CrossRef](#)]

117. EN 1991-1-4:2005+A1:2010; Eurocode 1: Actions on Structures—Part 1-4: General Actions—Wind Actions. European Committee for Standardization: Brussels, Belgium, 2010.
118. Pisano, G.; Royer-Carfagni, G.; Schneider, J. Open issues in the calibration of partial safety factors for heat-treated glass. *Struct. Saf.* **2019**, *79*, 1–11. [\[CrossRef\]](#)
119. Pisano, G.; Royer-Carfagni, G. Statistical interference of material strength and surface prestress in heat-treated glass. *J. Am. Ceram. Soc.* **2017**, *100*, 954–967. [\[CrossRef\]](#)
120. Pisano, G.; Bonati, A.; Royer Carfagni, G. Calibration of the statistical-interference factors for the design of tempered structural glass. *J. Eng. Mech.* **2022**, *148*, 04022013. [\[CrossRef\]](#)
121. Bedon, C.; Amadio, C. Assessment of analytical formulations for the ULS resistance verification of structural glass elements accounting for the effects of different load durations. *Structures* **2017**, *11*, 218–228. [\[CrossRef\]](#)
122. Franco, A.; Royer-Carfagni, G. Verification formulae for structural glass under combined variable loads. *Eng. Struct.* **2015**, *83*, 233–242. [\[CrossRef\]](#)
123. Favaro, N.; Ceola, S. Glass cullet: Sources, uses, and environmental benefits. In *Encyclopedia of Glass Science, Technology, History, and Culture*; Richet, P., Ed.; Wiley: Hoboken, NJ, USA, 2021; Volume 2, pp. 1179–1189.
124. Glass for Europe. A Circular Economy Act to Support Flat Glass Closed-Loop Recycling Systems. 2023. Available online: <https://glassforeurope.com/wp-content/uploads/2025/01/Position-Circular-Economy-Act-January-2025.pdf> (accessed on 1 November 2025).
125. Lebullenger, R.; Mear, F.O. Glass recycling. In *Springer Handbook of Glass*; Musgraves, J.D., Hu, J., Calvez, L., Eds.; Springer: Cham, Switzerland, 2019; pp. 1353–1375.
126. Rose, A.; Sack, N.; Nothacker, K.; Grassman, A. *Recycling of Flat Glass in the Building Industry—Analysis of the Current Situation and Derivation of Recommendations for Action*; ift-Research Report; ift Rosenheim: Rosenheim, Germany, 2019.
127. Griffin, P.W.; Hammond, G.P.; McKenna, R.S. Industrial energy use and decarbonisation in the glass sector: A UK perspective. *Adv. Appl. Energy* **2021**, *3*, 100037. [\[CrossRef\]](#)
128. Salman, M.; Flórez-Orrego, D.; Coppitters, D.; Mitraki, R.; Maréchal, F.; Léonard, G. Decarbonising the glass industry: A comprehensive techno-economic assessment of low-emission pathways. *Comput. Chem. Eng.* **2025**, *203*, 109329. [\[CrossRef\]](#)
129. Caudle, B.; Taniguchi, S.; Nguyen, T.T.H.; Kataoka, S. Integrating carbon capture and utilization into the glass industry: Economic analysis of emissions reduction through CO₂ mineralization. *J. Clean. Prod.* **2023**, *416*, 137846. [\[CrossRef\]](#)
130. Kobayashi, H. Future of oxy-fuel glass melting: Oxygen production, energy efficiency, emissions and CO₂ neutral glass melting. In *80th Conference on Glass Problems, Columbus, OH, USA, 28–31 October 2019*; Sundaram, S.K., Ed.; Wiley: Hoboken, NJ, USA, 2021; pp. 3–12.
131. Kim, H.; Kang, T.; Kaiser, K.; Liedel, S.; Jarry, L.; Paubel, X.; Jumaní, Y.; Kaya, L. Heat oxy-combustion: An innovative energy saving solution for glass industry. In *76th Conference on Glass Problems, Columbus, OH, USA, 2–5 November 2015*; Sundaram, S.K., Ed.; Wiley: Hoboken, NJ, USA, 2016; pp. 149–155.
132. Beerkens, R. Energy saving options for glass furnaces and recovery of heat from their flue gases and experiences with batch and cullet pre-heaters applied in the glass industry. In *69th Conference on Glass Problems, Columbus, OH, USA, 4–5 November 2008*; Drummond, C.H., Ed.; Wiley: Hoboken, NJ, USA, 2009; pp. 143–162.
133. Chakravarti, S.; Alexander, J.; Kobayashi, H. Waste heat recovery in oxy-fuel glass furnaces—A path to sustainability and lower CO₂ emissions. In *81st Conference on Glass Problems, Virtual, 26–30 October 2020*; Sundaram, S.K., Ed.; Wiley: Hoboken, NJ, USA, 2021; pp. 125–135.
134. van Limpt, H.; Beerkens, R. Energy recovery from waste heat in the glass industry and thermochemical recuperator. In *73rd Conference on Glass Problems, Cincinnati, OH, USA, 1–3 October 2012*; Sundaram, S.K., Ed.; Wiley: Hoboken, NJ, USA, 2013; pp. 3–15.
135. Fiehl, M.; Leicher, J.; Giese, A.; Görner, K.; Fleischmann, B.; Spielmann, S. Biogas as a co-firing fuel in thermal processing industries: Implementation in a glass melting furnace. *Energy Procedia* **2017**, *120*, 302–308. [\[CrossRef\]](#)
136. Glass for Europe. Flat Glass in Climate-Neutral Europe—Triggering a Virtuous Cycle of Decarbonisation. 2020. Available online: <https://glassforeurope.com/wp-content/uploads/2020/01/flat-glass-climate-neutral-europe.pdf> (accessed on 22 December 2025).

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