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SPACE FILLING MODELS OF AMORPHOUS STRUCTURES

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Résumé - La structure des corps amorphes est déterminée par les liaisons chimiques entre les atomes et par les contraintes topologiques de remplissage de l'espace. Les amorphes métalliques présentent un ordre local icosaédrique incompatible avec la périodicité cristalline, mais qui peut exister dans des espaces tridimensionnels de courbure non nulle. D'où il a été suggéré de construire des modèles d'amorphes métalliques par pavage icosaédrique d'espace courbe qui sont ultérieurement décourbés en préservant au maximum l'ordre local. Différentes méthodes de décourbure utilisent les disclinaisons (ou dislocations de rotation). À l'opposé, une autre méthode consiste à paver un espace, en moyenne euclidien, mais de courbure variant aléatoirement : les régions de courbure positive sont les centres de nucléation d'icosaèdres. Nous donnons des règles de somme sur les pavages qui reliant la distribution du nombre de premiers voisins (définis par la construction de Voronoi) à la courbure de l'espace. Nous discutons aussi la structure des "quasi-cristaux" dans lesquels une symétrie d'ordre 5 (ou 10) est observée. Un ordre d'orientation est encore présent, mais l'ordre de translation a disparu. Ces "quasi-cristaux" sont à mettre en relation avec les structures non périodiques de R. Penrose, qui présentent un caractère self-similaire. Les structures des covalents tétraédriques amorphes et des éléments des groupes V et VI sont aussi envisagées.

Abstract - The structure of amorphous systems is determined by the chemical bond between atoms and the topological constraints of the space filling requirement. Amorphous metallic systems show an icosahedral local order incompatible with crystalline periodicity, but such a local order may exist in 3 dimensional spaces with non vanishing curvature. Hence it has been suggested to build up models of amorphous metallic structure by tiling curved spaces with icosahedra: the structure is subsequently decurved by methods that preserve the local order. The concept of disclination (or rotation-dislocation) plays a central role in the decurving procedure. At the opposite, another method consists of tiling a randomly corrugated space which remains euclidean on average: regions of positive curvature are the nucleation centers of icosahedral environments. We define sum rules that relate the distribution of coordination numbers (defined by the Voronoi construction) to the curvature of the space. We discuss also the structure of "quasi-crystals" in which a fivefold (or tenfold) symmetry has been observed. An orientational order is still present but the translational order has disappeared. These "quasi-crystals" have to be related to the non-periodic structures invented by R. Penrose: both show a self-similar character. Endly, the structures of covalent fourfold coordinated amorphous systems are discussed as well as the structure of group V and VI elements.
I - INTRODUCTION

Amorphous and liquid structures are usually characterized by negative definitions such as the absence of long range order (LRO), the absence of crystalline symmetry, etc., by contrast to the crystal structures that represent the perfection. The aim of this paper is to show that some order exists in amorphous structures, even if it is less obvious to grasp than in the crystalline materials.

In any condensed matter systems, the structure is defined by the chemical bond between the constituent atoms leading to a short range order (SRO) and by the requirement to fill in space homogeneously. These two constraints (local and global) can fit together (e.g. sp$^2$ bonded atoms in the diamond or wurztite structures) or they are incompatible and a compromise has to be found between the SRO and the space filling requirement (e.g. Lennard Jones bonded atoms, tends to form tetrahedra but regular tetrahedra cannot fill in the space). The compromise results in a crystalline (in our example octahedral holes are periodically distributed in the structure), quasicrystalline or amorphous structure, depending on the preparation conditions. The kinetics (e.g. the cooling rate when prepared from the melt) plays a central role in determining which type of structure is obtained. The amorphous structures are known to depend on the preparation conditions and subsequent heat or mechanical treatments: by rapid quenching from the melt or cold substrate deposition or irradiation, the system is trapped in a local minimum of energy (metastable state) where it remains a "long" time.

The equilibrium thermodynamics cannot be applied to amorphous systems: the system samples a limited region of the configuration space in a finite time: the ergodicity is broken. Roughly speaking, below the recrystallisation temperature, the system cannot find its way to the crystalline structure (assumed here to be the most stable structure) within the finite time of the experiment.

When the system is cooled down to the amorphous structure, it satisfies as well as possible the constraints of the chemical bond at short distances but it cannot rearrange itself so as to minimize its structural energy as a whole. The system attends to the most pressing thing first.

The determination of the order in crystalline materials is in principle straightforward: a single diffraction experiment leads unequivocally to the crystal structure (forgetting the phase problem), essentially because the number of parameters required for describing a crystal structure is finite (and small). By contrast, in non-crystalline structures, the number of parameters (the coordinates of the atoms) is roughly infinite while the diffraction experiments give less informations (halos instead of Bragg's peaks) compared to crystals. The usual way to approach the structure is by modelling it at the microscopic level. The models are requested to fit the experimental pair correlation function but also the electronic vibration spectra, etc. and important macroscopic parameters as well like the density, the cohesive energy, etc. Indirect structure related quantities should in principle be fitted too like e.g. the plasticity.

 Needless to say that an amorphous structure is never unequivocally determined. Consequently a structural model has to be based on a sound description of the chemical bond between the atoms, responsible for the SRO.

In this paper, we concentrate only on the amorphous structure of pure elements. Although most amorphous systems are alloys, some pure (or nearly pure) elements (even metals) can be amorphized. In the last years substantial progress have been made in the field of understanding the topology of the structure of pure amorphous substances which is the unavoidable first step for understanding the structure of amorphous alloys. We show the relevance of the concepts of disclination and space curvature in the description of amorphous (or liquid) structures. On the other hand, the chemical SRO is reviewed by F. Gautier /1/ and J. Hafner et al. /2/. The interplay of topological SRO and compositional SRO require still more investigations.

II - A POOR MAN'S APPROACH OF COHESION

In this section we show, using simple arguments, that the electronic structure of the atom largely determines the bonding properties in condensed matter, in particular the directionality of the bond. Roughly the elements of columns I, II (and III to a certain extent) have a cohesion due to the s electrons. Because of their spherical
symmetry, the most stable structure corresponds to the largest number of interatomic bonds (around twelve in 3D). Practically, the amorphous structure is obtained by packing spheres as density as possible (Bernal's \(/3/\) construction).

At the opposite, the elements of columns \(V, VI\) and \(VII\) are linked by \(p\) electrons. As the \(p\) orbitals point at right angles in the three directions of the space, the structures are related to simple cubic lattice (e.g. structure of Po). However, a subsequent distortion of the sixfold coordinated lattice (Peierls instability) shortens some bonds and elongates the remaining ones. The group \(V\) (resp.\(VI\)) elements have a coordination \(3(+3)\) (resp. \(2(+4)\)) \(\sqrt{2}\) except the lightest which, with the group \(VII\) elements, form molecular crystals of diatomic molecules. We recover the \(Z\)\\(=8-N\) Grimm-Sommerfeld rule where \(Z\) is the coordination number and \(N\) the number of \(s\) and \(p\) electrons. Such rule is also followed by the amorphous and liquid phases at not too high temperatures \(\sqrt{2}\). Indeed, the temperature tends to destroy the Peierls instability so that at high temperature the liquids gets close to sixfold coordination (e.g. \(Te\)). The group \(VIII\) atoms (rare gases) do not form covalent bond but instead they are linked by non directional dispersion (Van der Waals) forces and so form compact structures.

The group \(IV\) elements are at borderline of different behaviours. Their bonding properties are due to \(s\) and \(p\) electrons. \(C\) is threefold coordinated in graphitic planes or crystallizes in a metastable diamond structure. \(Si, Ge\) and \(Sn\) are fourfold coordinated in the diamond structure. When temperature and/or pressure increases, they form a sixfold coordinated structure (\(B\)-\(Sn\) structure of \(Sn\), liquid \(Si\) and \(Ge\) with coordination number \(6.5\) \(\sqrt{6}\)) where the \(p\) electrons play a central role in the bonding mechanism. The heaviest element (molten \(Sn, Pb\)) shows compact metallic structures.

The transition metals have a cohesion dominated by their unfilled \(d\)-band \(\sqrt{2}\). The fivefold degeneracy of the \(d\) electrons gives lobes that point in many directions of the space, so that, the resulting anisotropy of the bond is no longer observable. Consequently one finds compact (\(FCC, HCP\)) or semicompact structures (\(BCC\)) as if the orbitals were spherically symmetric.

Summarizing briefly, three different types of structures are observed

- Groups I, II, III\(\text{+\text{VIII+}}\) transition metals : non-directional bonding. Referred to as "metallic bonding", in the following.
- Groups \(IV\) : directed covalent bonds (four bonds at \(109^\circ\)). [For liquid \(Si\) or \(Ge\), the rules of the following series apply]
- Groups \(V, VI\) : covalent bonds related to distorted simple cubic structures.

This constitutes the set of rules for constructing hand or computer built models. The case of compounds involving covalent elements is more complex : the atoms may be linked according to the \(8-N\) rule like in their pure elements (e.g. \(SiO\),) or the behave as if they were pure elements with an average number of electrons (e.g. \(GaAs,Ge\) or \(GaAs\) like \(As\)). Let us stress that the bonding properties at short distances are independent of the periodicity.

### III - COMMENSURABILITY - CURVED SPACES

Packing spheres (or disks in 2D) gives qualitatively different results in 2D and 3D and the curvature of the space is an important parameter. In 2D, the SRO generates exactly the LRO but it does not in 3D. Indeed packing disks of the same size in an euclidean 2D space leads to the triangular lattice (Fig.1a): six equilateral triangles with a common vertex tile a portion of the euclidean plane without any void because \(60^\circ\) is a submultiple of \(360^\circ\).

By contrast, the tiling properties are quite different in 3D : the regular tetrahedra do not tile exactly the space. The dihedral angle of the tetrahedron \(2\ \arcsin(1/\sqrt{3}) = 70.53^\circ\) is not commensurate with \(360^\circ\), so that a little bit more than \(5\) \((5.104)\) tetrahedra can be put around a common edge. If one puts twelve atoms around a central atom in the most symmetric structure, one gets an icosahedron in which the surface atoms are not in contact. Their ratio distance \(1\) to diameter \(D\) \((1/D=1.05)\) is a measure of the incommensurability.

This situation is quite similar to the 2D hyperbolic (negatively curved) space shown in Fig.1b. By analogy, it is possible to tile exactly a positively curved 3D space with regular tetrahedra. The hypersphere \(S^3\) with the equation \(x^2+y^2+z^2+t^2=R^2\) and \(R/D=(1+\sqrt{5})/2\) (the golden mean) is paved with 120 atoms (called the polytope \(\{3,3,5\}\)), each atom being the center of a regular icosahedron (Fig.2). As it is admitted that
atoms in amorphous metals have our approximate icosahedral symmetry, the polytope \( \{3,3,5\} \) is good starting point for building amorphous models as suggested by Sadoc in his pioneering contributions to this field /9,10,11/.

The pair correlation function of the \( \{3,3,5\} \) polytope gives a strikingly good agreement with the experimental pair correlation function obtained on a-Fe, up to 3.5 D. On the polytope the distances are measured along geodesic lines. Let us remark that the distances are quite different from those measured on a FCC structure i.e. \( d = \sqrt{2} D \) (i=1...13 without disruption of the series). The \( \{3,3,5\} \) polytope has a shell structure: first shell of 12 nearest neighbours at a distance \( D \), second shell of 20 atoms at a distance 1.69 D and a third shell of 12 atoms at 2.00 D. The last two shells constitute the split second peak of \( g(r) \) which is recognized as a (quasi) universal feature of amorphous metals. At this stage we are left with a model that reproduces correctly the \( g(r) \) but which is as regular as a crystal (a rotation group of 14400 elements plays the role of the translation and symmetry group of the crystal) and it is embedded in a (unphysical) curved space.

The question is now to decrease the curvature to zero while preserving as much as possible the icosahedral SRO of the curved space model. Several methods may be used to reduce or cancel the curvature

a) The \textit{decoration}. An atom is replaced by a group of atoms. The radius of curvature is increased but the SRO can be considerably altered. In fact this procedure is used to transform a structure into another (see section VI).

b) The \textit{star projection} (Fig.3). It is the 3D analogous of peeling an orange. Lobes of skin are cut and applied on the 3D space with a limited deformation. The shorter...
the lobes, the smaller the deformation. It is clear that the SRO is largely preserved as well as the distances but large surfaces are created (or lines in the 2D example of Fig.3).

c) The method of disclinations. A more systematic procedure for changing the curvature is based on the concept of disclination (or rotation dislocation). In the continuum approximation, a disclination can be generated by the Volterra method: a cut is made in the plane (2D) (Fig.4) or in the space (3D); matter is added or removed and the lips of the cut are reglued. In 2D (resp. 3D) a disclination is a point (resp. line) defect defined by its amplitude $\alpha$ and its sign.

![Disclinations of the plane](image)

**Fig.4**: Disclinations of the plane. A finite angle of matter is removed a) or added c).

At the disclination center (resp. line) curvature appears, negative if matter is added or positive if it is removed. Consequently, the disclination appears as a tool for changing the curvature of the space, in particular for flattening the polytopes constructed on curved spaces.

If a network is put on the underlying space, the amplitude $\alpha$ of the disclination is quantized (e.g. a multiple of $60^\circ$ for the honeycomb or triangular lattice, $90^\circ$ for the square lattice). If the disclination center is inside a polygon, it changes the number of edges (Fig.5): if it coincides with a vertex, it changes its coordination number.

![Disclinated honeycomb lattices](image)

**Fig.5**: Disclinated honeycomb lattices. a) $60^\circ$ of matter is removed b) or added. c) The dodecahedron can be considered as a disclinated honeycomb lattice in which every hexagon is a disclination center.

In 3D, two types of disclinations have to be considered like in dislocations, the wedge (Fig.6.a) and screw (Fig.6.b) disclinations.

![The two types of disclinations in 3D](image)

**Fig.6**: The two types of disclinations in 3D. Wedge $(90^\circ)$ (a) and screw $(90^\circ)$ (b) disclinations.
The latter have been used -almost exclusively up to now- for flattening curved space models. In 2D, in the continuum model where the disclination amplitude is not quantized, it is possible to consider a continuous distribution of disclinations and to define a disclination density \( s(\mathbf{r}) \). The reference state (with a vanishing disclination density) is the euclidean plane. Then it is possible to show \( \text{12} \) that the \( s(\mathbf{r}) \) is nothing but the Gauss curvature of the space

\[
s(\mathbf{r}) = K(\mathbf{r})
\]  

This relation shows again the role played by the disclinations for varying the space curvature. An integral form of (1) can be obtained for 2D non periodic lattices of constant coordination number \( c \) (e.g. covalent lattices without dangling bonds). In that case, considering disclinations that are inside the polygons and consequently change the number of edges, leaving the coordination constant, it is easy to demonstrate that \( \text{13} \)

\[
\sum_{p=3}^{\infty} (2c-cp+2p) F_p = \frac{c}{\pi} \oint K d\sigma
\]

Where \( F_p \) is the number of polygon with \( p \) edges and \( K \) is the Gauss curvature and is the \( \mathcal{D} \) domain of integration.

For 3 coordinated systems, formula (2) writes

\[
2F_4 + F_5 - F_7 - 2F_8 \ldots = \frac{3}{\pi} \oint K d\sigma
\]

We observe that the number of hexagons is not restricted: the hexagons act as neutral charges, while pentagons are +1 charges, heptagons are -1 charges, etc. Endly formula (3) is a charge conservation rule. If a pentagon is created, a heptagon has to be created simultaneously, if the average curvature given by the right member is kept constant. Practical realizations are a) the honeycomb lattice \( (K=0, \mathcal{D}=\text{plane}) \) b) the regular dodecahedron \( (K=1/R^2, \mathcal{D}=\text{sphere}) \) for which \( F_5=12 \); c) the soccer ball for which \( F_5=12 \) and \( F_7=20 \). In formula (3) the hexagon \( \delta \) is a flat tile (for 3 coordinated systems) while a pentagon (resp. heptagon) generates positive (resp. negative) curvature (cf. Fig.5). Formula (3) is useful in hard discs packings as the Voronoi construction \( \text{14} \) which determines the coordination numbers of the atoms is a 3 coordinated lattice. In that case \( F_p \) is the number of \( p \) coordinated atoms. Such formula is difficult to generalize in 3 dimensions but the general mechanism remains the same: subtracting a wedge of matter increases the positive curvature and vice versa.

Consequently, one sees that disclinations can be used to cancel the curvature of a model of amorphous structure created in a curved space like the \( (3,3,5) \) polytope. Adding disclinations to a curved space show some similarities with the second method of flattening the space (peeling the orange, Fig.3) but instead of creating surfaces or internal voids, the structure is homogeneous by construction. The local order is preserved, except near the disclination line (in 3D), so that the good agreement (Fig.2) between experiment and the theoretical model remains. Practically, the addition of disclinations one by one in a 3D curved space to make it flat is very difficult because the disclination algebra is non commutative \( \text{15} \). Disclination lines cannot end in the structure but ramification modes may exist. Progress are currently made. Two types of structures have been obtained by a complete flattening of the \( (3,3,5) \) polytope: crystalline structures and hierarchical structures.

IV - CRYSTALLINE AND HIERARCHICAL STRUCTURES AS DEFECTED \( (3,3,5) \) POLYTOPE

Several crystalline structures can be obtained from the \( (3,3,5) \) polytope which is decurved by a periodic array of disclination lines. An analogy in 2D is the transformation from the dodecahedron to the honeycomb lattice: at the center of every pentagon or +60° disclination transforms it into a hexagon.

Similarly, in 3D, one gets a series of periodic structures by disclinating an icosahedral medium (the \( (3,3,5) \) polytope) some of them are very simple, others ones have very large unit cells \( \text{14} \).
The P-W structure (A15 structure of compounds $V_3Si$, $Nb_3Ge$...) is a \{3\,3\,5\} polytope with atoms on the vertices in which disclination lines are along on the metal atom chains of $V_3Si$ (mutually orthogonal but non crossing). (Fig.7). Similarly, the Laves phases A,B are also examples of a structure derived from the \{3\,3\,5\} polytope with disclinations lines lowering the curvature to zero. But in this case four disclination lines meet at a common vertex \$v_3\$, the coordination number of which is 16. The disclination lines form a diamond structure. The A atoms keep a coordination number 12, so that the average coordination is 13.33, very close to the ideal coordination number of 13.39.

These two examples show the relevance of the \{3\,3\,5\} polytope even in the description of crystalline structures. However, let us endly remark that the simplest metallic structures (FCC,HCP) have not been described up to now as the result of a disclination network on the \{3\,3\,5\} polytope.

Fig. 7: The P-W structure can be considered as a disclinated \{3\,3\,5\} polytope. The disclination lines are the full lines.

Fig. 8: Hierarchical flattening.

Non-crystalline structures can also be obtained from curved space models by an iterative flattening method. Because it is very difficult to add disclination lines one by one to the polytope, it has been suggested instead \cite{16} to generate a disclination network with a symmetry group contained in the symmetry group of the polytope. Fig.8 shows the disclination network (heavy lines) used to disclinate the network of dodecahedra sharing edges (the \{5\,3\,3\} polytope containing 600 atoms, which is the dual of the \{3\,3\,5\} polytope). Each primitive dodecahedron is decurved by a set of four disclination lines crossing at 109°, so that four pentagonal faces are transformed into hexagonal faces. At this stage, the flattening is not complete. Fig.8 suggests that the process can be repeated as the network of defects is similar to the original network, at another scale, consequently the process can be iteratively continued. The structure obtained at the end is non crystalline, however it cannot be considered as disordered because of its symmetry properties resulting from the construction.

Some disorder could be introduced if we had two (or more) different iterative procedures which have to be compatible i.e. their associated defect structure share the same symmetry group.

In this section the defects associated to the flattening are intrinsic (or minimal) in the sense that they have to exist to completely flatten the structure. Random structure have an excess of positive and negative disclinations.

V - EXTRINSIC DEFECTS. THE RANDOMLY CORRUGATED STRUCTURE

As it is very difficult to insert disclinations individually in a random way, one can imagine working in a different way. Instead of starting from a curved space idealization, which is a highly symmetric and non euclidean structure one can start from a randomly corrugated space which is euclidean in the average and pack on it hard spheres (or discs in 2D).

In 2D, the regions of positive (resp. negative) curvature are attractor centers of
fivefold (resp. sevenfold) coordinated atoms, provided the integral of the right
and side of (3) is of the order of +1 (resp. -1) on a domain $\mathcal{D}$ of diameter $2D$. Let
us remark that the corrugation is clearly observed in molecular dynamics experiments
of the 2D melting /17/. At zero temperature the triangular crystal corresponds to a
flat space where every atom has six neighbours; when the temperature increases qua-
druplets of neighbouring atoms change their coordination from six to five and seven.
Quadrupole of disclinations (or dipole of dislocations) appear. They correspond to
an incipient corrugation of the underlying space, which amplifies when temperature
goes up.
In 3D, the situation is analogous with the important difference that the most stable
and symmetric state does not correspond to a flat underlying space but to an hyper-
sphere. The requirement of a vanishing mean curvature writes

$$\sum_{i,j} L_+^i \delta_+^i + L_-^j \delta_-^j \approx 0$$

(4)

where $L_+^i$ and $L_-^j$ are the length per unit volume of positive (of strength $\delta_+^i>0$) and
negative ($\delta_-^j<0$) disclinations. The left hand member of (4) is exactly zero if the
disclinations are infinitesimal i.e. if the curvature is spread in the volume. If the
curvature is concentrated on disclination lines, the compensation of positive and
negative curvatures is only approximate /13/.
Let us endly remark that in a liquid, the disclination network evolves with time,
and so does the underlying corrugated space.
If the disclinations are characterized by their address instead of their strength
/18/ it is possible to estimate the maximal entropy per mole associated to the defects
which is

$$S = R \ln 2$$

(5)
in good agreement with experimental data on melting.

VI - BOND ORIENTATIONAL ORDER AND QUASICRYSTALS

Recently, D. Schechtman et al. /21/ reported the observation of a tenfold symmetry
in patterns of Bragg peaks (Fig.9) observed by electron microscopy in rapidly cooled
Al$_6$Mn alloys. As a multiple twinning has been ruled out in this case, the famous
theorem of crystallography that excludes a fivefold (or tenfold) symmetry axis and
the periodic repetition of the unit cell has to be reconciled with these experiments.
Let us recall that the symmetry of the diffraction pattern does not imply the same
symmetry for every atom in the crystal. This a priori astonishing diffraction result

![Fig. 9](image1)

![Fig. 10](image2)

was in fact the experimental counterpart of a serie of recent remarks of theoretists
studying of the amorphous structures. In 1974, R. Penrose /21/ showed that it was
possible to tile the infinite plane with two different unit cells, showing a fivefold
symmetry (Fig. 10). 3D extensions of the Penrose tiling have been proposed by McKay
/22/, Mosseri and Sadoc /23/ and Kramer /24/. We see in Fig.10 that the angles of the
rhombuses are integral multiples of $36^\circ(=\pi/5)$, so that the direction of the edges are quantized: only ten different directions are observed. A tenfold long range orientational order exists. It has been shown both theoretically and by diffraction with a laser beam, that the Fourier transform of a Penrose tiling is made up of 6 peaks, with a tenfold symmetry. The diffraction pattern has a striking resemblance with the pattern obtained on Al$_2$Mn. The mathematical structure of the pattern is very unusual: it is a dense set of 6 δ-peaks, the amplitudes of which decrease rapidly with the order of diffraction, so that only a finite number of them are observable. The sharpness of the Bragg spots is due to a long range order in the orientation of the bond (Bond orientational order) and to a quasi-translational order. In these apparently disordered structures, a strong order exists but it is hidden. Recently, Duneau and Katz have shown that the Penrose tilings can be obtained by projecting a six dimensional simple hypercubic structure into a 3 dimensional space. The agreement with the diffraction data and more recent high resolution electron microscopy experiments are excellent. The projection method shows that the structure is non random, as a projection of a periodic structure. The number of degree of freedom of these structures is reduced by the self similarity that plays a role analogous to the translational symmetry in crystals (this has to be compared also to the hierarchical structure of defects of section V).

However, up to now, it has not been possible to suggest a microscopic model with atoms of Al and Mn.

Structures with long range order in the orientation but with short range translational order have previously been found in other systems: the nematic liquid crystals phases in which rod-like molecules align themselves parallel to their axis. The hexatic phase appearing in dislocation mediated 2D melting is another example that has some counterpart in 3D fluids. Indeed, Nelson et al. have shown by molecular dynamics simulation of a Lennard-Jones fluid that an icosahedral orientational correlation grows with undercooling.

VII - COVALENT STRUCTURES

Despite the fact that amorphous covalent structures are known for a long time (e.g. the chalcogenide glasses), that they represent an important class of materials (e.g. glasses, amorphous semiconductors) and that their structure has been correctly described half a century ago (Zachariazen's continuous random network) little theoretical work has been successfully devoted to their structure. In principle, the dual structure of an amorphous metal (in the sense of the Voronoi's construction) is an amorphous tetravalent structure. But such cage-like models are not satisfactory because their dihedral angles are vanishingly small in disagreement with experimental evidences. So that more theoretical work is needed to understand the structure of amorphous tetravalent semiconductors. The groups V and VI amorphous (and liquid) semiconductors have not yet been considered along the lines described in this paper. Their bonding properties are due to the p electrons that have been recognized to favor the formation of amorphous structures. As simple cubic lattice can be chosen as the starting structure, in which wedge and screw disclinations should be added.

VIII - CONCLUSION

It is now widely recognized that the structure of amorphous materials can be described by a two step procedure: the definition of an ideal structure which may necessitate an embedding in a non euclidean space and the definition of defects identified as disclinations.

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