



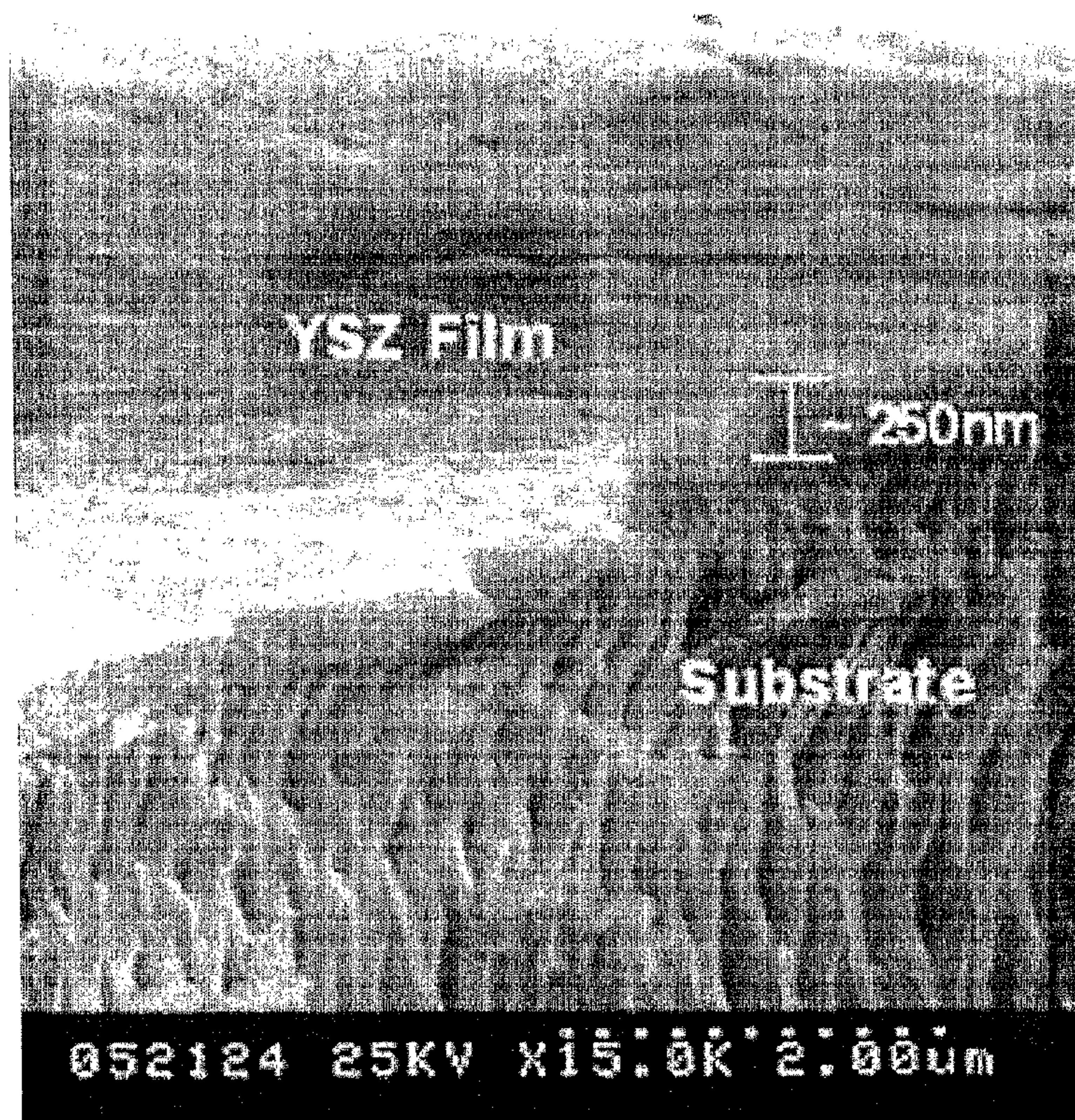
(86) Date de dépôt PCT/PCT Filing Date: 2003/05/29
(87) Date publication PCT/PCT Publication Date: 2003/12/11
(45) Date de délivrance/Issue Date: 2011/11/01
(85) Entrée phase nationale/National Entry: 2004/11/29
(86) N° demande PCT/PCT Application No.: US 2003/017202
(87) N° publication PCT/PCT Publication No.: 2003/101629
(30) Priorités/Priorities: 2002/05/29 (US60/384,380);
2002/05/29 (US60/384,378)

(51) Cl.Int./Int.Cl. *B01D 69/12* (2006.01),
B01D 67/00 (2006.01), *B05D 5/12* (2006.01),
C04B 35/48 (2006.01), *C04B 35/486* (2006.01),
C04B 35/50 (2006.01), *C04B 41/00* (2006.01),
C23C 14/35 (2006.01), *H01M 8/10* (2006.01),
H01M 8/12 (2006.01)

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(54) Titre : FILM MINCE ELECTROLYTIQUE SUBMICRONIQUE FORME SUR UN SUBSTRAT NANOPOREUX PAR
OXYDATION D'UN FILM METALLIQUE
(54) Title: SUB-MICRON ELECTROLYTE THIN FILM ON NANO-POROUS SUBSTRATE BY OXIDATION OF METAL
FILM



(57) Abrégé/Abstract:

A fluid impermeable thin film (20) is fabricated on a porous substrate (10) by depositing a material (4) having a certain spatial oxidation expansion. After depositing the material (4) is oxidized (5) whereby the deposited material (4) expands and forms a void free film (5) on top of the porous substrate (10). Grain boundaries (6) of the void free film (5) may recombine to form a continuous film (20) on the porous substrate (20).

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(12) INTERNATIONAL APPLICATION PUBLISHED UNDER THE PATENT COOPERATION TREATY (PCT)

(19) World Intellectual Property Organization
International Bureau



(43) International Publication Date
11 December 2003 (11.12.2003)

PCT

(10) International Publication Number
WO 03/101629 A1

(51) International Patent Classification⁷: **B05D 5/12**

(21) International Application Number: PCT/US03/17202

(22) International Filing Date: 29 May 2003 (29.05.2003)

(25) Filing Language: English

(26) Publication Language: English

(30) Priority Data:
60/384,378 29 May 2002 (29.05.2002) US
60/384,380 29 May 2002 (29.05.2002) US

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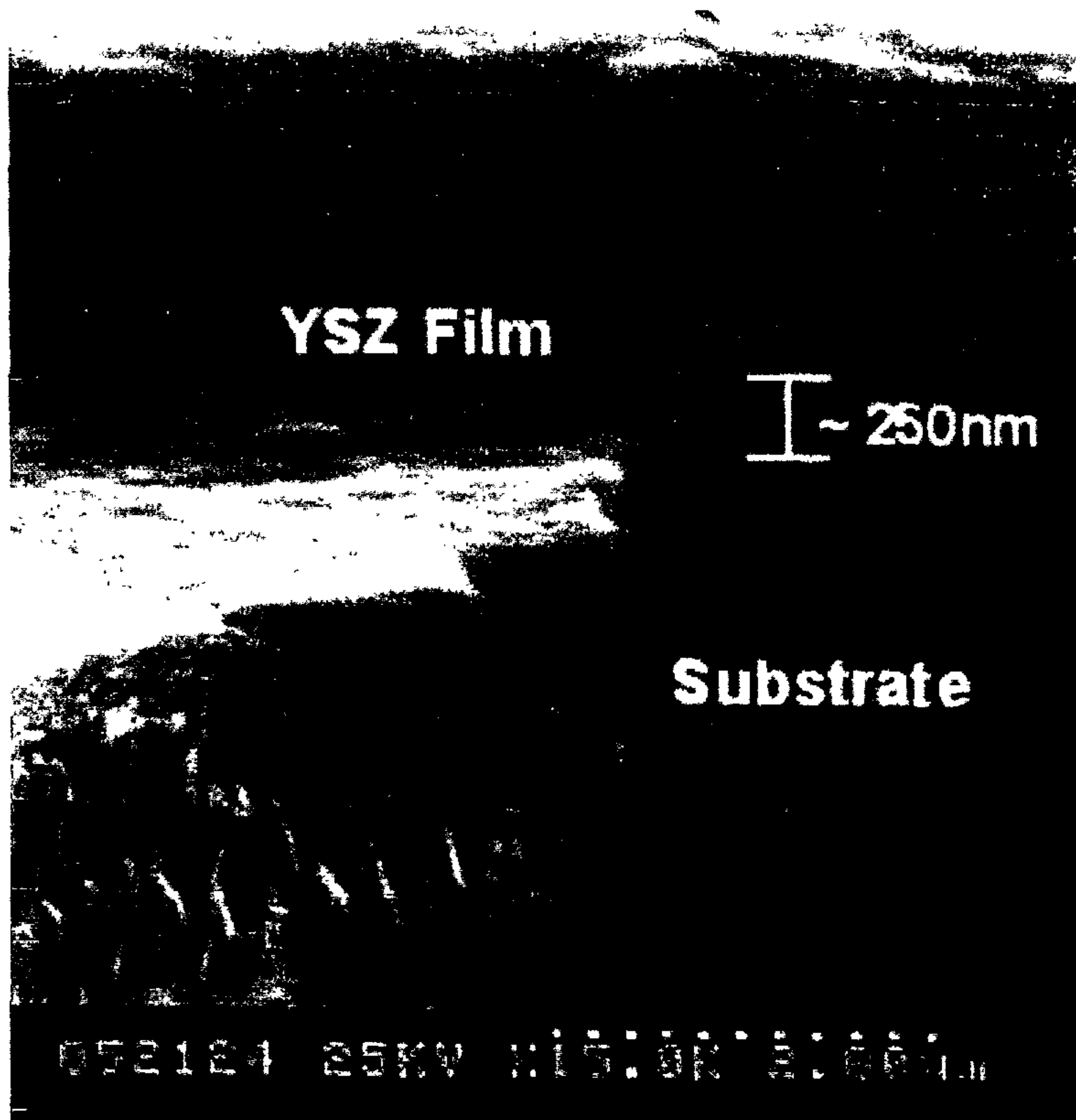
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(81) Designated States (*national*): AE, AL, AM, AT, AU, AZ, BA, BB, BG, BR, BY, CA, CH, CN, CO, CR, CU, CZ, DE, DK, DM, DZ, EC, EE, ES, FI, GB, GD, GE, GH, GM, HR, HU, ID, IL, IN, IS, JP, KE, KG, KP, KR, KZ, LC, LK, LR, LS, LT, LU, LV, MD, MG, MK, MN, MW, MX, NO, NZ, OM, PH, PL, PT, RO, RU, SC, SD, SE, SG, SK, SL, TJ, TM, TN, TT, TZ, UA, UG, US, UZ, VC, VN, YU, ZA, ZM, ZW.

(84) Designated States (*regional*): ARIPO patent (GH, GM, KE, LS, MW, MZ, SD, SL, SZ, TZ, UG, ZM, ZW), Eurasian patent (AM, AZ, BY, KG, KZ, MD, RU, TJ, TM), European patent (AT, BE, BG, CH, CY, CZ, DE, DK, EE, ES, FI, FR, GB, GR, HU, IE, IT, LU, MC, NL, PT, RO,

[Continued on next page]

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SE, SI, SK, TR), OAPI patent (BF, BJ, CF, CG, CI, CM, GA, GN, GQ, GW, ML, MR, NE, SN, TD, TG).

For two-letter codes and other abbreviations, refer to the "Guidance Notes on Codes and Abbreviations" appearing at the beginning of each regular issue of the PCT Gazette.

Published:

— *with international search report*

**Sub-micron Electrolyte Thin Film on Nano-Porous Substrate
by Oxidation of Metal Film**

FIELD OF INVENTION

The present invention relates generally to electrochemical devices and methods. More particularly, the present invention relates to solid oxide fuel cells [SOFC].

BACKGROUND

A fuel cell is an electrochemical device that produces electrical current from chemical reactions. The fundamental device includes an ion-conducting electrolyte between two electrodes, backed by fuel and oxidant flow distributors. A catalyst on one electrode promotes separation of ions and electrons at the oxidant side. Only the ions conduct through the electrolyte, and recombine with electrons at the fuel side. The electrons are conducted through an external circuit, thus supplying electrical power. Solid oxide fuel cells have ionic conducting metal oxide membranes as their electrolyte layer. The oxygen molecules are split into their respective electrons and oxygen ions at the airside. The oxygen ions propagate through the electrolyte membrane and combine with their electrons and hydrogen molecules into water.

Fuel cell operation is increasingly efficient where the well-known electron conductivity of the

electrolyte is brought to a minimum and the well-known ionic conductivity of the electrolyte is brought to a maximum. At the same time it is desirable to keep the fuel cell's driving temperature as low as possible such that the well-known thermodynamic efficiency and the well-known electric load responsiveness of the fuel cell remains at high levels.

As is well-known in the art, ohmic resistance and ionic conductivity of conventional electrolyte materials vary with electrolyte temperature. In that context, **Table 1** lists exemplary values of ionic conductivity σ [S/cm], ohmic resistance of electrolyte film at 1cm² of area and at given film thickness R [Ω] for various yttria stabilized zirconium [YSZ] film, thickness of various YSZ film h [μ m, nm] and temperatures t.

Table 1.

t [°C]	σ [S/cm]	R [Ω] for h=5 μ m	R [Ω] for h=500nm	R [Ω] for h=200nm	R [Ω] for h=20nm
503	0.001	0.5	0.005	0.02	0.002
422	1.58e-4	3.1	0.31	0.12	0.01
360	3.00e-5	16.6	1.66	0.66	0.07
322	8.00e-6	62.5	6.25	2.5	0.25
250	5.11e-7	978	97.8	39.1	3.9
200	4.54e-8	11010	1101	440	44
100	4.88e-11	10245900	1024590	409836	40983

The bold and italic entries of **Table 1** indicate ohmic values relevant for efficient fuel cell operation. Hence, by reducing the electrolyte height, the driving temperature, the ohmic resistance and the ionic conductance are reduced as well.

At the time the present invention was made, extensive effort has been dedicated to reduce electrolyte thickness. In one well-known tape casting technique for example, slurry of submicrometer-sized powders is used to produce highly dense and mechanically strong electrolyte films with good electrical properties. Unfortunately, the electrolyte films are fabricated only with relatively large heights in the range of hundreds of micrometers. This reduces the ionic conductance across the electrolytes' heights to relatively low levels that may be only partially compensated by high working temperatures of 800 - 1000 °C.

In another approach, electrolyte films are fabricated as anode-supported thin oxide films with electrolyte heights between 5 - 20 μm resulting in a working temperature range of 500 - 1000 °C. Even though a significant reduction of the lower working temperature limit was accomplished, 500 °C require still significant constructive effort and limit the feasibility of such a fuel cell for practical applications.

At the time of the invention, high-tech processing such as well-known PVD, CVD, PLD and sol-gel deposition has been tested to further reduce electrolyte thickness and increase film density. Since the electrolyte operates also as a membrane that physically separates reactant fluids within the fuel cell, film density becomes more important with

decreasing electrolyte thickness to provide gas impermeability at sufficiently high levels. Unfortunately, the high-tech deposition technologies have been developed mainly for highly flattened substrates, especially silicon wafers. To the contrary, a substrate qualifying for deposition of an electrolyte membrane must be highly gas permeable to provide for fluid conductance through the substrate and for a direct fluid contact with the electrolyte at the electrolyte side adjacent the substrate. A qualifying substrate needs to be highly porous, which results in a relatively rough, discontinuous and inhomogeneous surface on a scale similar to that of the electrolyte's film height.

A successful deposition of continuous submicron YSZ thin films on anodized nanoporous alumina has been demonstrated in the prior art by a sol-gel deposition technique using viscous alkoxide-derived solution applied on top of the porous substrate. However, sol-gel derived thin films exhibit a large shrinkage during heat-treatment and low density from inherent high organic content that may cause local defects. Consequently, continuous films that are substantially fluid impermeable may not be fabricated with sol-gel deposition techniques.

In summary, prior art oxide film deposition techniques are not suitable in combination with a qualifying substrate. Therefore, there exists a need for a fabrication technique for making an operational electrolyte membrane on top of a porous substrate. The present invention addresses this need.

In general, an electrolyte layer of a solid oxide fuel cell should have the following properties:

- high ionic conductivity, which means also low ionic resistance;
- low electron conductivity, preferably by reduction of thickness;
- high density and impermeability to prevent electric potential drop between both sides of the electrolyte layer due to reactant fluid mixture;
- sufficient mechanical strength at operational fluid pressures; and
- good adhesion to electrode layers to reduce resistance between electrolyte and electrode for increased fuel cell efficiency and to prevent second phase formation.

The present invention addresses these general needs.

In particular, an electrolyte layer is needed that may be fabricated in an inexpensive fashion with a configuration that provides for an efficient fuel cell operation at working temperatures of 500 °C and substantially less. The present invention addresses also these needs.

SUMMARY

The present invention provides for new functional solid oxide fuel cell assemblies. The present invention is based on a different set of materials and processes for manufacturing functional solid oxide fuel cell assemblies compared to previous devices or methods. In summary, the

present invention provides a direct metal film deposition to obtain ultra-thin oxide film by oxidation. Furthermore, the present invention utilizes a porous substrate (e.g. a nanoporous anodized alumina substrate) as a supporting material for electrolyte thin film. A preferred embodiment of the present invention is based on:

1. Selecting a deposition material having a certain spatial oxidation expansion. The selected material may include a metal alloy and/or individual metals prepared as a complex metal sputtering target. Such a complex metal sputtering target may be, for example yttrium pieces allocated on Zr target for direct deposition of metal Yttrium-Zirconium layer.
2. Depositing the deposition material in unoxidized condition directly on the porous substrate. To utilize increased ionic conductivity along grain boundaries without degrading parasitic capacitances it is desirable to deposit the material with single grain height such that ions may propagate along continuous grain boundaries between opposing sides of the final continuous film. The sputtering parameters may be adjusted accordingly and in a well-known fashion. An exemplary deposition of the Yttrium-Zirconium layer may be accomplished by DC-magnetron sputtering on a porous gamma alumina substrate in Ar atmosphere. Even though the deposited material may optically appear a solid surface, microscopic grain boundaries and voids make the deposited layer highly fluid permeable.
3. Oxidizing the deposited material in any suitable fashion. For example, the Yttrium-Zirconium layer may be oxidized in an oxidizing atmosphere at a

temperature between 300°C ~ 1000°C. The spatial expansion that the deposited material undergoes while it is oxidized causes the formation of a void-free film.

4. Heat treatment of the void-free for controlled grain modification including grain growth. Grain growth during heat treatment is not associated with a spatial expansion like during oxidation. Grain growth during heat treatment results from recombination of individual grains whereby eventual residual voids are eliminated.

In fuel cells where it is desired to have the electrolyte film as thin as possible, the simple and inexpensive fabrication of ultra thin fluid impermeable electrolyte films is highly advantageous for providing a highly efficient fuel cell that may be operated at a low working temperature well below 500 °C.

Some of the advantages of the present invention over existing devices and methods include:

1. A decrease of electrolyte resistance that is enabled by the use of the above and below described metal-oxidized electrolyte thin film processes in combination with porous substrates such as nano-porous anodized alumina substrate or other well-known porous metallic and/or ceramic substrates such as porous silicon.

2. A simple and clean fabrication that is enabled by the use of well-known thin film deposition techniques such as DC-magnetron sputtering processes in Ar gas.
3. Flexible fabrication costs, which are provided by a broader choice of deposited materials such as relatively inexpensive Yttrium and Zirconium alloy or complex target compared to high-cost yttria stabilized zirconia target.

BRIEF DESCRIPTION OF THE FIGURES

- Fig. 1** illustrates the inventive utilization of spatial expansion during oxidation with a block diagram and associated schematic. The schematic on the left side of Fig. 1 depicts simplified material grains in a frontal cut view.
- Fig. 2** schematically depicts the steps for fabrication of a continuous film on a porous substrate with associated block diagram.
- Fig. 3a, 3b, 3c** show schematic cut view portions of exemplary interfaces between continuous film and porous substrate.
- Fig. 4** is an enlarged picture of the top of a 2 00nm-type gamma alumina nanoporous substrate.

- Fig. 5** is an enlarged picture of the top of a 20nm-type gamma alumina nanoporous substrate.
- Fig. 6** is an enlarged picture of a deposited unoxidized alumina on top of a 200nm-type gamma alumina nanoporous substrate. The continuous film and the substrate are depicted along a fracture site.
- Fig. 7** is an enlarged picture of a continuous film of oxidized Yttrium stabilized Zirconium layer on top of a 200nm-type gamma alumina nanoporous substrate. The continuous film and the substrate are depicted along a fracture site.
- Fig. 8** is an enlarged picture of a continuous film top of oxidized Yttrium stabilized Zirconium layer.
- Fig. 9** is an enlarged picture of a continuous film of oxidized Yttrium stabilized Zirconium layer on top of a 20nm-type gamma alumina nanoporous substrate. The continuous film and the substrate are depicted along a fracture site.
- Fig. 10** is an enlarged picture of a continuous film bottom of oxidized Yttrium stabilized Zirconium layer. The substrate is etched away.
- Fig. 11** shows a bar graph with hydrogen permeances for various substrate conditions and various continuous films.

- Fig. 12** depicts a schematic test setup for measuring ionic resistance of a continuous film fabricated on a porous substrate.
- Fig. 13** illustrates measurement results obtained with the test setup of **Fig. 12**.
- Fig. 14** shows an enlarged portion of the graph of **Fig. 13**.
- Fig. 15** depicts X-ray normalized reflectometric measurement plots at various stages during fabrication of the continuous film in accordance with a preferred embodiment.
- Fig. 16** shows a base line compensated X-ray reflectometric measurement plot of a final continuous film in accordance with a preferred embodiment.
- Fig. 17** shows ionic conductivity in dependence of working temperature for continuous films with varying thickness made of various deposited materials.
- Fig. 18** graphically illustrates the bending of Arrhenius line by dislocation paths having a given activation energy and having varying densities.
- Fig. 19** shows a device having a continuous film on a porous substrate in accordance with the present invention.

DETAILED DESCRIPTION

Two factors mainly influence the operational configuration of a continuous ultra thin film. A first factor is fabrication feasibility. A second factor is operational structure considerations. In the case where the continuous ultra thin film is utilized as a membrane, fluid impermeability becomes a third influencing factor. In case, where the continuous ultra thin film is utilized as an electrolyte, ionic conductivity becomes a fourth influencing factor.

Referring to **Fig. 1** and in compliance with the first factor, the present invention takes advantage of a spatial oxidation expansion, which a deposited material **4** may undergo during an oxidation process. The spatial oxidation expansion may be broken down into three stages. During a first stage of sole oxygen diffusion, the grains **2** expand substantially symmetric during oxidation of the deposited material **4**. Symmetric spatial expansion takes place as long as the expansion remains physically uninhibited. Oxidized material **5** begins to progress from the grain boundary **6** towards the grain center continuously replacing the deposited material **4**.

Once adjacent grains **2** come into contact, a second stage of oxidation, including a mass transfer, is initiated. During the second stage, an asymmetric spatial expansion takes place while the oxidation of the deposited material **4** continues towards the grain centers. Adjacent grains **2** contact each other at the grain interface **7**. The asymmetric spatial expansion stems from the restricted expansion in the grain interface **7** between the adjacent grains **2**. During

the second stage, eventual voids and pores between adjacent grains **2** are filled with the expanding oxidized material. Voids and pores are likely resulting from initial deposition of the grains **2**.

Once all deposited material **4** is oxidized, during the third stage, the spatial oxidation expansion discontinues. Due to high temperature, mass transfer may still continue and voids and pores between adjacent grains **2** are filled with oxidized material **5**. Excess material is urged along the grain boundaries **6** resulting also in an increase of height **8**. Asymmetric expansion continues, blocking the pores. At the end of the three stages oxidation process, a void free film is formed.

A spatial oxidation expansion ratio between the unoxidized material and the oxidized material may be readily predetermined in a well-known fashion. In the exemplary case of unoxidized Aluminum (FCC, $a=4.049\text{\AA}$) as the deposited material **4** and well-known Alpha Alumina (HCP, $a=4.758\text{\AA}$, $c=12.99\text{\AA}$) as the oxidized material **5**, the spatial oxidation expansion ratio is 1.279. In the exemplary case of an 85/15 Zirconium/Yttrium alloy (HCP, $a=3.232\text{\AA}$, $c=5.147\text{\AA}$) as the deposited material **4** and cubic yttria stabilized zirconia (FCC, $a=5.139\text{\AA}$) as the oxidized material **5**, the spatial oxidation expansion ratio is 1.45.

Important for obtaining a void free film is the oxidization expansion of the deposited grains within the deposition plane. The in plane portion of the spatial oxidation expansion ratio may be calculated as the second power of the third root of the spatial oxidation expansion ratio.

Hence, the in-plane oxidation expansion ratio for unoxidized Aluminum to oxidized Alpha Alumina is about 1.178. The in-plane oxidation expansion ratio for unoxidized 85/15 Zirconium/Yttrium alloy to yttria stabilized zirconia is about 1.131. As may be well appreciated by anyone skilled in the art and in order to obtain a void free film, an in-plane deposition density of the deposited material may be at least the inverse of the in-plane oxidation expansion ratio. In that context, in-plane deposition density refers to the average ratio between voids and deposited grains **2** across a deposition height of the deposited grains **2**. Accordingly, the deposition density for unoxidized aluminum may be at least about 0.849 and for unoxidized 85/15 Zirconium/Yttrium alloy may be at least about 0.78.

The three-stage oxidization is preferably performed at temperatures well above an operational temperature of the finally fabricated continuous film. An oxidization temperature is selected such that the deposited material and/or the oxidized material is/are in a condition of increased plasticity. The increased plasticity in turn is important for the mass transfer. YSZ is oxidized preferably at temperatures between 300° ~ 400°C. **Fig. 2** illustrates the application of the three-stage oxidization in fabrication of a continuous film **20** on top of a porous substrate **10**. The schematic on the left side of **Fig. 2** depicts a simplified frontal cut view through the porous substrate **10** with three stage oxidation and a final heat treatment step.

Initially, the material **4** is deposited in a substantially unoxidized condition on top of the solid portion of the porous substrate **10**. A preferred deposition material may be a metal and a preferred deposition technique may be a well-known DC-magnetron sputtering. The porous substrate **10** may be, for example a 200nm-type Gamma Alumina substrate with vertically protruding pores having a diameter between 150 ~ 200nm (see **Fig. 4**). The 200nm-type porous substrate may have a void/solid ratio of about 9/1, which means that about 10% of the porous substrate's **10** top are solid and the remaining about 90% of the porous substrate's **10** top are open.

The porous substrate **10** may also be, for example a 20nm-type Gamma Alumina substrate with vertically protruding pores having a diameter between 80 ~ 200nm (see **Fig. 5**). The 20nm-type porous substrate may have a void/solid ratio of about 1/1, which means that about 50% of the porous substrate's **10** top are solid and the remaining about 50% of the porous substrate's **10** top are open. Porous substrate **10** made from Gamma Alumina may be commercially available under the tradename Anodisc®.

To fabricate the final continuous film **20** with minimum thickness, the initial deposition step is adjusted to provide the required deposition density while keeping the deposition height to a minimum. As the void/solid ratio of the porous substrate **10** increases, deposition density becomes more challenging to accomplish for a targeted deposition thickness. For example, to deposit Aluminum on a 200nm type porous substrate **10** with required deposition density, about 11.8% of the deposited material **4** is

directly supported by the solid portion of the porous substrate **10**. This example is solely presented for the purpose of general understanding of the structure considerations that exist for optimized deposition of material **4** on top of the porous substrate **10**.

In the preferred case of a DC-magnetron sputtering, deposition thickness may be minimized by adjustment of deposition parameters, such as for example, sputtering angle and/or kinetic energies of the sputtered material.

Following the deposition of the unoxidized material **4**, the three stage oxidation is performed as described under **Fig. 1** resulting in a void free film. In a final fabrication step, the void free film is heat treated for recombining the oxidized grains along the grain boundaries **6**. In addition, the heat treatment process may be utilized to smoothen the top surface **23** of the continuous film **20**. Heat may be applied as directional radiation heat and/or as convective heat. Heat treatment is preferably accomplished in a furnace.

The heat treatment step may be performed at the same and/or different temperature or atmospheric settings than applied during oxidation. Nevertheless, oxidation and heat treatment may overlap, since grain recombination may take place as soon as adjacent grain boundaries **6** expand into contact.

Referring to **Figs. 3a, 3b** operational structure considerations for minimizing the height **21** of the continuous film are described in more detail. In context

with the present invention, operational structure considerations relate to the minimum continuous film structure required for a given porous substrate **10** and mechanical operation conditions to which the continuous film **20** may be exposed. The simplest case is depicted in **Fig. 3a** in which the continuous thin film **21** is a substantially equally thick layer. According to **Fig. 3a**, a minimum for the film thickness **21** may be calculated by the following equations.

$$\sigma_{\max} = 3/8 \cdot \{(1+\nu) \cdot (\Delta P \cdot R^2)\} / t^2 \quad [1]$$

$$t = [3\{(1+\nu) \Delta P R^2\} / (8 \sigma_{\max})]^{1/2} \quad [2]$$

where, $\sigma_{\max}$: Ultimate tensile strength

ΔP : Pressure difference

ν : Poisson's ratio

R : Pore radius

t : Film thickness

The minimum film thickness for sample A (pore size: 20nm, ΔP : 0.1MPa) estimated with the reported mechanical properties of commercial 99.9% bulk alumina ceramics is 0.15nm. However, there are too many differences between bulk ceramics and thin films to directly apply the values reported for bulk alumina to this estimation. Considering the difference in grain size, density, impurity, and defects, especially pore diameter and distribution, which

critically affect fracture strength, the reasonable minimum thickness might be larger than the estimation. Nevertheless, the estimated thickness limit is key reference for practical optimization of the continuous film **20**.

In a similar fashion as the calculations presented for the simplified case of **Fig. 3a**, theoretical thickness minima may be calculated for thin film configurations schematically depicted in **Figs. 3b, 3c**. In **Fig. 3b**, the bottom portion **22** of the continuous film **20** may have a concave curvature eventually resulting from a shrinkage during cooling of the continuous film **20** after heat treatment. In addition, the film may attach laterally at the sidewalls of the substrate's **10** pores. Lateral attachment may be result from the deposition process and/or from the oxidation expansion.

According to **Fig. 3c**, the bottom portion **24** may have a convex curvature eventually resulting from an excess vertical expansion. Such excess vertical expansion may result from a deposition density significantly above the required deposition density where voids and/or pores are filled significantly before the oxidation expansion is completed. The enlarged picture of **Fig. 10** shows such a case.

In the following, two exemplary continuous films **20** experimentally fabricated on porous substrates **20** are described in detail. For porous substrates **10**, commercial inorganic alumina filters (Anodisc®, Whatman Inc.) with a pore size of 20nm and 200nm were used. In context with the present invention, the terms "20nm type" and "200nm type"

relates to terminology used by the producer with respect to the filtering ability of their respective product. The observed actual pore diameters by SEM were 80nm ~ 200nm for 20nm-type filters and 150nm ~ 300nm for 200nm-type filters. The filter diameter (without support ring) was 40mm, filter thickness was 60 μ m and maximum working pressure was about 0.52MPa.

An Aluminum target with 99.999% purity and a Y-Zr complex target consisting of three or four 5mm $\times$ 5mm $\times$ 1mm-sized Y pellets with 99.9% purity on a Zr target with 99.7% purity were used for metal film deposition using DC-magnetron sputtering. The Ar gas flow rate was 10sccm and Ar pressure was 1.2Pa at 50W for aluminum deposition and 10sccm ~ 30sccm, 1.2Pa ~ 2.8Pa at 30W for Y-Zr deposition.

After deposition of metal film, subsequent oxidation was performed at 700°C for 2hours in air. Two Alumina film samples with metal thickness of 30nm and 200nm were prepared (**Table 2**), and five YSZ samples were prepared with three compositions of Y/Zr(at%) = 4/96, 16/84, 43/57 having different metal thickness from about 40nm~230nm. Thickness of the films was controlled with the deposition rate of aluminum (9.43Å/s) and Y-Zr and was measured by using a Si wafer substrate with the same deposition condition. The composition of Y/Zr was controlled by changing location and number of Y pellet. The Y/Zr composition was determined by X-ray photoemission spectroscopy (SSI S-Probe Monochromatized XPS Spectrometer). All samples were oxidized at 700 °C for two hours. The composition of the Y/Zr samples had a purity of higher than 99.7%. The

microstructure of the thin films was observed with SEM. The phase development of sample A including a separate heat-treating step at 400°C~1300°C for 2hours was observed using XRD as is described in the following.

Table 2

Mat.	Sample	Substrate type	Deposition condition (time-Ar flow[Pa]/ pressure [Pa]- energy[W]	Deposited thickness [nm]	Estimated thickness range of continuous film [nm]	Observed thickness of continuous film [nm]	Composition of continuous film
Al	A	20	32-10/1.2-50	30	32.6-38.6	~35	Al=99.999
Al	B	200	212-10/1.2-50	200	217.2-257.4	~240	Al=99.999
Y-Zr	C	20	400-10/1.2-50	38	43.0-55.1	~50	Y/Zr=3/97
Y-Zr	D	200	400-10/1.2-50	38	43.0-55.1	~50	Y/Zr=3/97
Y-Zr	E	20	2000-10/1.2-50	190	215.1-275.5	~250	Y/Zr=3/97
Y-Zr	F	200	2000-10/1.2-50	190	215.1-275.5	~250	Y/Zr=3/97
Y-Zr	G	20	4000-30/2.5-20	-	-	-	Y/Zr=43/57
Y-Zr	H	200	4000-30/2.5-20	230	283.0-362.0	~320	Y/Zr=16/84

The samples obtained were tested in a test setup illustrated in **Fig. 12**. A sample **1** was fixed with polymer adhesives to a thick silicone tube **125** filled with hydrogen **127**, and then immersed in NaCl-water solution **126** to measure hydrogen permeance through oxidized thin films **20**. The hydrogen pressure (ΔP) used was 0.1MPa. Ionic conductivity of oxidized Y-Zr films was measured from nyquist plots obtained using 0.03N NaCl-water solution **126** as electrodes with Solartron 1260/1287 impedance analyzer **122** at room temperature.

In **Figs. 4** and **5**, SEM images of the Anodisc® substrates having of 200nm-type and 20nm-type are shown. The pores are columnar and penetrate through the porous substrates **10**. **Fig. 6** shows the surface and fractured edge after deposition of unoxidized aluminum on the porous substrate **10**. Even though the deposited film thickness is only about 200nm, it already covers the whole surface showing smooth and grain-like aggregated metal islands. There are also flake-shaped particles on the metal surface suggesting that impurities existed and worked as nuclei for aggregation of metal atoms during sputtering. A continuous film **20** of oxidized Aluminum was obtained from the deposited film of **Fig. 6** after heat-treated at 700°C for 2 hours. The continuous film **20** displayed a homogeneous surface covering pores of substrates without cracks or pinholes. The oxidized Aluminum film **20** may be utilized as a membrane where ionic conductivity is irrelevant. The Aluminum sample where primarily fabricated and tested for fluid impermeability and fabrication feasibility.

Fluid impermeability was tested in a hydrogen permeation test in which the hydrogen permeance was measured at 0.1 MPa. The results are depicted in **Fig. 11a**. The permeance for sample B (column **(a)**) is $6.40 \times 10^{-6} \text{ mol/m}^2 \cdot \text{s} \cdot \text{Pa}$ and slightly decreased to $1.97 \times 10^{-6} \text{ mol/m}^2 \cdot \text{s} \cdot \text{Pa}$ by heat-treatment at 700°C for 2 hours (column **(b)**). This may be due to change of the dimension of columns by shrinkage during heat-treatment. The permeance for sample B (with a thin film, column **(c)**) drastically decreased to $2.01 \times 10^{-9} \text{ mol/m}^2 \cdot \text{s} \cdot \text{Pa}$.

The permeances measured for the sample A are reflected in columns **(d)**, **(e)** and **(f)**. They show smaller values than

those of (a), (b) and (c) due to their smaller pore size. The gas permeance for sample A (column(f)) could not be detected with the present measurement settings and thus the measurable minimum permeance of $7.44 \times 10^{-11} \text{mol/m}^2 \cdot \text{s} \cdot \text{Pa}$ is assigned for column(f).

Fabrication feasibility of continuous films 20 is strongly influenced by the deposited material's 4 oxidation behavior and phase development. To observe phase development during oxidation of the thin films of samples A, B, XRD patterns at various oxidation stages and during the heat-treatment were obtained. No distinct crystalline peak was observed until the temperature reaches 900°C. However, the color of thin film changed from metal silver to glassy transparent at 500°C. At 900°C, theta-alumina phase was detected and the peak intensity increased with increasing temperature. At 1300°C well-developed theta phase was observed. However, several peaks of alpha phase were also detected in the pattern of 1300°C. Since the film thickness was just 30nm, the collected patterns were strongly affected by the substrate 10 and show both phases of thin film 20 and substrate 10. To determine whether or not this alpha phase was from the thin film, the XRD pattern was compared with that of a bare substrate 10 heated at 1300°C for 2hours. The comparison clearly showed that only the pattern from thin film showed alpha phase even though the peak intensity was small compared to that of bare substrate. The relative peak intensity estimated from the fitted alpha phase peaks matched well with the relative intensity from reported data for bulk polycrystalline alpha-alumina

In summary, Aluminum **20** deposited with a height of about 30nm oxidizes around 500°C and transforms to alpha phase between 1000°C ~ 1300°C. Otherwise, the substrate **10** has amorphous phase up to 700°C and then transforms to theta-form at 800°C. The protective oxide layer formed on the surface of aluminum and low oxygen diffusivity may be responsible for the high crystallization temperature observed. It was found that aluminum film with a thickness of 200nm was not fully oxidized after 10 hours of heat-treatment at 700°C, while clear single YSZ phase was easily achieved from oxidized Y-Zr alloy deposited with a height of about 200nm and heat-treated at 700°C for 2hour.

For the preferred embodiment of an electrolyte membrane in a fuel cell, continuous films **20** made of deposited Y/Zr alloy are now described in detail. A Y/Zr alloy deposited with a thickness of about 50nm displayed in an SEM photography a silver-colored smooth surface without pores or voids being detected. The actual deposition density of the samples C - H was above the required deposition density. Consequently and as depicted in **Fig. 10**, the bottom of the derived continuous film **20** appeared similar to the bottom of **Fig. 3c**.

As shown in **Fig. 7** and after oxidation and heat treatment at 700°C for 2h, sample E displayed a continuous film **20** without any cracks and pinholes appeared on the porous substrate **10**. Sample H is shown in **Figs. 8** and **9**. **Fig. 8** shows the film top **23** and **Fig. 9** shows the fractured edge of Y/Zr=16/84 film oxidized at 700°C for 2h. The surface was not smooth and many large island-shaped grains were found. Although sample H showed good fluid-separation, the

rough surface may limit achievable minimum film thickness. The inhomogeneous surface could be due to the high Ar pressure to retain plasma on the complex target surface with the high contact resistance between Y-pellet and Zr target.

Fig. 10 shows the backside-view of the continuous film **20** of Sample E that was in contact with the 20nm-type substrate **10**. To have unobstructed access to the bottom said of the continuous film **20**, the substrate **10** was etched out with 1N-NaOH solution. The bottom side **24** was different from the top surface **23** of the same sample E. The bottom side displayed circle-shaped islands with diameter ranging from 50nm to 120nm on the whole surface, suggesting that the thin film covering the top of the substrate's **10** pores was swallowed into the pores.

Fluid impermeability was tested in a hydrogen permeation test in which the hydrogen permeance was measured for the oxidized Y-Zr thin films **20** at 0.1MPa hydrogen pressure. The results are illustrated in **Fig. 11b**. The permeance of the substrate **10** without heating (column(a) for 200nm-type substrate **10** and column (e) for 20nm-type substrate **10**) and with heating at 700°C for 2h (column(b), for 200nm-type substrate **10** and column (f) for 20nm-type substrate **10**) are again shown for comparison. The permeance for sample D (column(c)) slightly decreased to $7.49 \times 10^{-7} \text{mol/m}^2 \cdot \text{s} \cdot \text{Pa}$. The continuous thin film **20** of sample D did not demonstrate sufficient fluid-separation compared to the low permeance value in the 35nm-thick Alumina thin film on 200nm-type substrate **10**. However, the permeance for the sample F (column(d)) dropped to $7.44 \times 10^{-11} \text{mol/m}^2 \cdot \text{s} \cdot \text{Pa}$, which was below

the measuring limit. This implies that the 250nm-thick film **20** on 20nm-pore-sized substrate **10** is enough to work as a gas separation layer. That is, the thickness of 50nm deposited with DC-magnetron sputtering in a substantially perpendicular deposition direction was insufficient to form a continuous film **20** across the pores of the 200nm-type porous substrate **10**.

The samples C, G, and E fabricated on top of the 20nm-type substrate **10**, displayed small fluid permeance values as indicated with columns(**g**), (**h**) and (**i**). Fluid permeance for sample E and G (columns(**h**) and (**i**)) was below measurement resolution and could not be detected (illustrated in **Fig. 11b** as $7.44 \times 10^{-11} \text{mol/m}^2 \cdot \text{s} \cdot \text{Pa}$, which was the lower measuring limit). Since the measured minimal hydrogen permeance for initial driving of an SOFC was about $4.5 \times 10^{-9} \text{mol/m}^2 \cdot \text{s} \cdot \text{Pa}$, the obtained results strongly support the high possibility for direct application in low-temperature SOFCs.

The minimum film thickness for continuous films **20** of YSZ on 20nm-type porous substrate **10** is estimated with 6.61nm by use of **Eq[1,2]** for $\Delta P = 0.1 \text{MPa}$ at room temperature. The difference between the reported mechanical data for 8YSZ sheet and thin films may be reasonably applied to adjust the minimum thickness estimation.

Oxidation and Phase Development of YSZ Thin Films **20** was observed using again XRD patterns obtained with sample C. The measurement results are illustrated in **Fig 15**. There was no distinct crystalline peak in the pattern at room

temperature (graph(c)) and at 500°C (graph(d)) for sample E. However, the color of thin film changed from metal silver to glassy transparent during oxidation at 500°C for 10h. At 700°C, a clear crystalline phase was detected. As mentioned above, the crystalline phase for Y-Zr film was easily achieved at a lower temperature than Alumina thin film due to its inherent high oxygen diffusivity. The observed crystalline peaks matched a mixed phase of cubic and monoclinic (graph (e)) that had been reported as a crystalline phase for a solid solution of Y_2O_3 and ZrO_2 having a composition in the range of about 3/97 ~ 15.6/83.4 of Y/Zr composition. By increasing the Yttrium concentration from 3at%(sample E, graph(e)) to 16at%(sample H, graph(f)) and 43at%(sample G, graph(g)), the monoclinic phase in oxidized thin film **20** disappeared and clear cubic single phase was achieved. The relative intensity displayed in **Fig. 16** was estimated from the fitted peaks, matched well with the relative peak intensity from reported data for bulk polycrystalline $Y_{0.15}Zr_{0.85}O_{1.93}$ indicating that the film had no preferred orientation.

The crystalline YSZ could be easily obtained by oxidation at 500°C ~ 700°C, a quite lower temperature than Alumina thin film. Oxidation of the deposited Y/Zr alloy material **4** occurs at very low temperatures in a relatively short period compared to deposited Alumina material **4**. As a favorable result, the inventive process has also significance as a low-temperature processing for oxide thin films, besides its preferred application for fabricating ultra-thin film with nano-size conductivity effects as described in more detail in the below.

Conductivity of YSZ thin Films was measured with Nyquist plots shown in **Figs. 13** and **14**. Measurements were taken for sample E ((a) ○, Y/Zr=3/97, thickness: 250nm), sample C ((b) ■, Y/Zr=3/97, thickness: 50nm) and sample H ((c) △, Y/Zr=16/84, thickness: 320nm). The obtained obvious semicircles represent resistance of the YSZ thin films **20** on porous substrates **10**. A small resistance component was associated to the electrode-solution **126**. About 6000Ω resistance at the left ends of the semicircles matched well with the measured electrode resistance of 0.03N-NaCl. As the thickness was reduced from 250nm(sample E) to 50nm(sample C), the resistance decreases from $3.67 \times 10^8 \Omega$ (Fig. 13, 14(a)) to $2.50 \times 10^7 \Omega$ (**Fig. 13, 14(b)**). A large decrease of resistance to $1.79 \times 10^5 \Omega$ (sample H, **Fig. 13, 14(c)**) was observed by changing composition from Y/Zr=3/97 to Y/Zr=16/84 in spite of the film thickness being thicker (~320nm) than the other two samples.

The ionic conductivity of samples E and C, which had substantially the same composition (Y/Zr=3/97) and different thickness (50nm and 250nm), calculated from the ionic resistances obtained from the nyquist plots (**Fig. 13, 14(b)** and **(c)**) were $4.07 \times 10^{-12} \text{S/cm}$ and $1.11 \times 10^{-12} \text{S/cm}$ respectively. These results showed almost the same value. However, the conductivity for sample H (**Fig. 13, 14(a)**) with a composition of increased yttrium concentration (Y/Zr=16/84) showed a much higher value of $3.42 \times 10^{-9} \text{S/cm}$. All the conductivities obtained for the thin films **20** are indicated in **Fig. 17**.

As can be seen in **Fig. 17**, the conductivity values for thin films **20** are not on the extrapolated line of the reported Arrhenius plot of 8YSZ (**Fig.17(e)**). All the measured conductivity values are largely shifted to the estimated Arrhenius plot (**Fig.17(a)**) with low activation energy of only 0.18eV. **Fig. 17(a)** is a reference graph suggested in the prior art based on the fast oxygen diffusion through grain boundaries and its large contribution to lower the activation energy(ΔH) for oxygen ionic conduction. Since the ionic conductivity term (σ) is proportional to $\exp(-\Delta H/kT)$, the contribution of lowered ΔH to σ exponentially increases with a decrease of temperature. As all the conductivity of the samples was measured just at room temperature, it is difficult to define activation energy. However, it is reasonable to say that ΔH for sample H (**Fig. 15(b)**) possibly is in the range of 0.18eV to 0.47eV, which is much smaller than 1.02eV (**Fig. 15(e)**). That is, if the estimated ΔH value of 0.18eV is exact and the activation energy for the defect-free YSZ is exactly the same as that of the grains in the obtained thin films **20**, defect density should be the only variable to change the absolute conductivity.

The bending Arrhenius plot at low temperature and change of the absolute conductivity can be simulated with a model with edge dislocations. The ratio of the conductivity of YSZ single crystal films having penetrating edge dislocation paths with different densities having an activation energy $\Delta H_d = \Delta H_0/2$ to the conductivity of bulk YSZ single crystal having an activation energy ΔH_0 is calculated by the following equations:

$$\varphi = \pi\phi^2\delta \quad [5]$$

$$D_d = D_o \cdot (1-\varphi) \cdot \exp(-\Delta H_o/kT) + D_o \cdot \varphi \exp(-\Delta H_d/kT) \quad [6]$$

$$D_b = D_o \cdot \exp(-\Delta H_o/kT) \quad [7]$$

therefore,

$$D_d/D_b = D_o \cdot [(1-\varphi) \cdot \exp(-\Delta H_o/kT) + \varphi \exp(-\Delta H_d/kT)] / [D_o \cdot \exp(-\Delta H_o/kT)] \quad [8]$$

$$= (1 - \varphi) + \varphi \exp[(-\Delta H_d + \Delta H_o)/kT] \quad [9]$$

where, φ : volume fraction of dislocations

ϕ : radius of dislocation

δ : dislocation density

D_b : diffusivity of 8YSZ single crystal

D_d : diffusivity of 8YSZ with dislocations

k : Boltzmann's constant

T : temperature

Several assumption are used in the estimation:

- (i) all the dislocations are assumed as edge-dislocations, and as arranged in parallel in the current-flow direction penetrating thin films;
- (ii) the shape of dislocation is assumed as pipe with circular crosssection, and the diameter of dislocation as $3\sqrt{2}/4$ times of lattice parameter(a) because the dislocation diameter (2ϕ) is supposed to be in the range of $\sqrt{2} a/2 < 2\phi < \sqrt{2} a$ for the $a/\sqrt{2}\langle 110 \rangle\{100\}$ dislocation system;
- (iii) the diffusivity(D_0) is assumed to have the same value in bulk and in dislocations because it depends on several constants that are independent with activation energy. The D_{od} , that is D_0 of 8YSZ with dislocations, is expressed as $Ar^2vN_v \cdot \exp[(\Delta S_o + \Delta S_d)/k]$, where all the variables are constants at the given temperature, that is, a proportional constant(A), atomic oscillation frequency(v), vacancy density(N_v), and entropy(ΔS). Although the absolute value of the conductivity will be slightly changed by applying D_{od} into the estimation for **Eq. [8]**, that is, $D_d/D_b = D_o \cdot [(1-\phi) \cdot \exp(-\Delta H_o/kT) + \phi \exp(-\Delta H_d/kT)] / [D_o \cdot \exp(-\Delta H_o/kT)]$. However, the slope of the conductivity plot, $-\Delta H/k$, in the lower temperature region is not affected.

The Arrhenius plot of YSZ with dislocations estimated with **Eq. [9]** and extrapolated Arrhenius plot of a reported 8YSZ single crystal are bent as shown in **Fig. 17(a) - (c)**, indicating a great increase of conductivity at temperatures below 500°C. With increasing dislocation density, the

temperature to deviate from **Fig. 17(d)** increases, accompanying a significant increase of absolute conductivity. Since the grain boundary **6** itself may be theoretically expressed as bundles of dislocations, the estimation performed for a dislocation effect on the conductivity at low temperature can be one of reasonable explanations for the extraordinary increase of ionic conductivity at low temperature, especially in thin films **20** with nanometer-sized grains **2**, which contain a large volume fraction of grain boundaries **6**. Parasitic capacitances may build up where ions eventually have to bridge across adjacent grain boundaries **6** perpendicularly oriented to the ions' path from one thin film surface to the opposing surface. To minimize the degrading effect of parasitic capacitances, it is desirable to deposit the material **4** in a fashion that grains **2** predominantly form in a single plane. Consequently, grain boundaries **6** extend substantially continuously from surface-to-surface and provide highly conductive pathways substantially free of parasitic capacitances. The experimental results suggest that heat treatment following oxidation may be adjusted to tune recombination of grain boundaries **6** for an optimized balance between low temperature ion conductivity and fluid impermeability.

It will be clear to a person of average skill in the art that the preferred embodiments described in the above may be altered in many ways without departing from the scope of the invention. Particularly, other porous substrates **10** such as porous silicon may be used with deposition materials **4** that exhibit oxidation expansion. Also, other well-known ceramic material with suitable ionic

conductivity may be fabricated in a fashion similar as described in the above for YSZ. Such ceramic may be for example, gadolinium doped ceria.

Ionic conductivity of the continuous film **20** may be further improved by fabricating continuous surface-to-surface dislocations into the continuous film **20**. For a more detailed description it is referred to the concurrently filed and cross referenced application titled "Solid oxide electrolyte with ion conductivity enhancement by dislocation".

Referring to **Fig. 19** and the case, where the continuous film **20** operates as an electrolyte membrane in a device **100** such as a fuel cell or a gas sensor, the porous substrate **10** may be made electrically conductive in a well-known fashion like, for example through electroplating and/or doping. The porous substrate **10** may also be a metallic substrate.

Accordingly, the scope of the invention described in the specification above is set forth by the following claims and their legal equivalent:

What is claimed is:

1. A method for fabricating a continuous film on a porous substrate, said method comprising:

a. selecting a deposition material having a oxidation expansion ratio, wherein an in-plane portion of the oxidation expansion ratio is equal to the second power of the third root of the oxidation expansion ratio;

b. depositing said deposition material with a deposition height and deposition density on top of said porous substrate, said deposition material being in an unoxidized condition;

c. oxidizing said deposition material on top of said substrate whereby said deposition material is converted into an oxidized material being spatially expanded with respect to said deposition material;

wherein said deposition density is greater than or equal to the inverse of said in-plane portion of the oxidation expansion ratio;

wherein said deposition height and said deposition density are selected in conjunction with said spatial oxidation expansion such that during said oxidizing a void-free film is formed on top of said substrate; and

- d. heat treating said void-free film to recombine said oxidized material into said continuous film.
2. The method of claim 1, wherein said deposition material is selected such that following said oxidizing said oxidized material is an ion conducting electrolyte and such that said continuous film is fluid impermeable.
 3. The method of claim 2, wherein said deposition material is an yttria stabilized zirconia.
 4. The method of any one of claims 1 to 3, wherein said porous substrate is made of an oxide.
 5. The method of claim 4, wherein said porous substrate is made of anodized alumina.
 6. The method of any one of claims 1 to 5, wherein on top of said porous substrate are pores through the full depth of the substrate with a pore diameter of less than 200 nm, said continuous film having a thickness of less than 1 μm and having a hydrogen permeability of less than $10\text{e-}10$ mol/m²sPa at room temperature.
 7. The method of any one of claims 1 to 6, wherein said continuous film has a hydrogen permeability of less than $10\text{e-}10$ mol/m²sPa at room temperature and an ionic area resistance of less than 200 ohm at a temperature of 250° C. for a thickness of said continuous film of about 1 μm and for said continuous film being made of yttria stabilized zirconia.

8. The method of any one of claims 1 to 7, wherein said continuous film is an electrolyte membrane for a fuel cell.
9. The method of any one of claims 1 to 8, wherein said depositing is performed with DC magnetron sputtering.
10. The method of any one of claims 1 to 9, wherein said porous substrate includes pores having a diameter between 80 nm and 200 nm prior to said heat treating.

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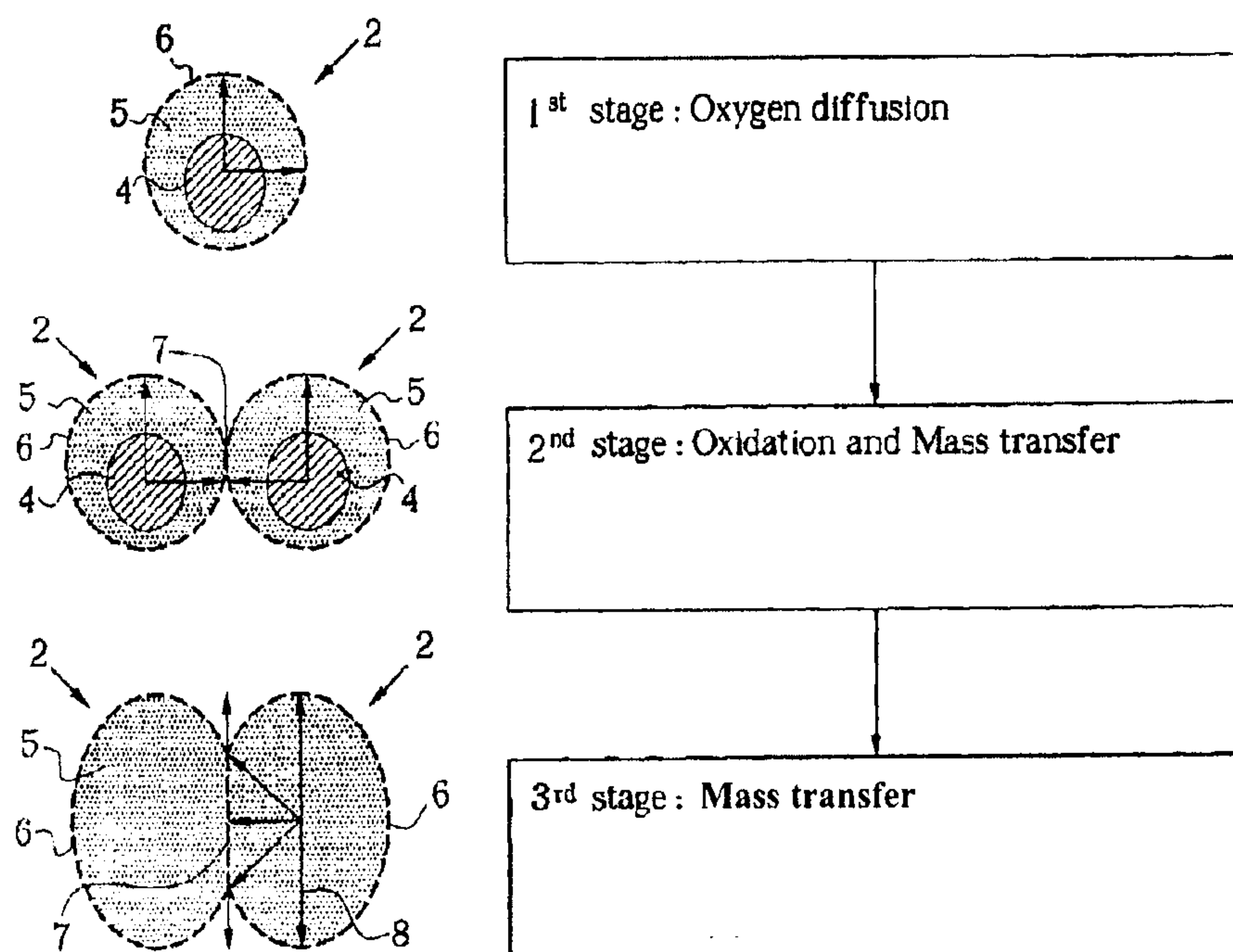


Fig. 1

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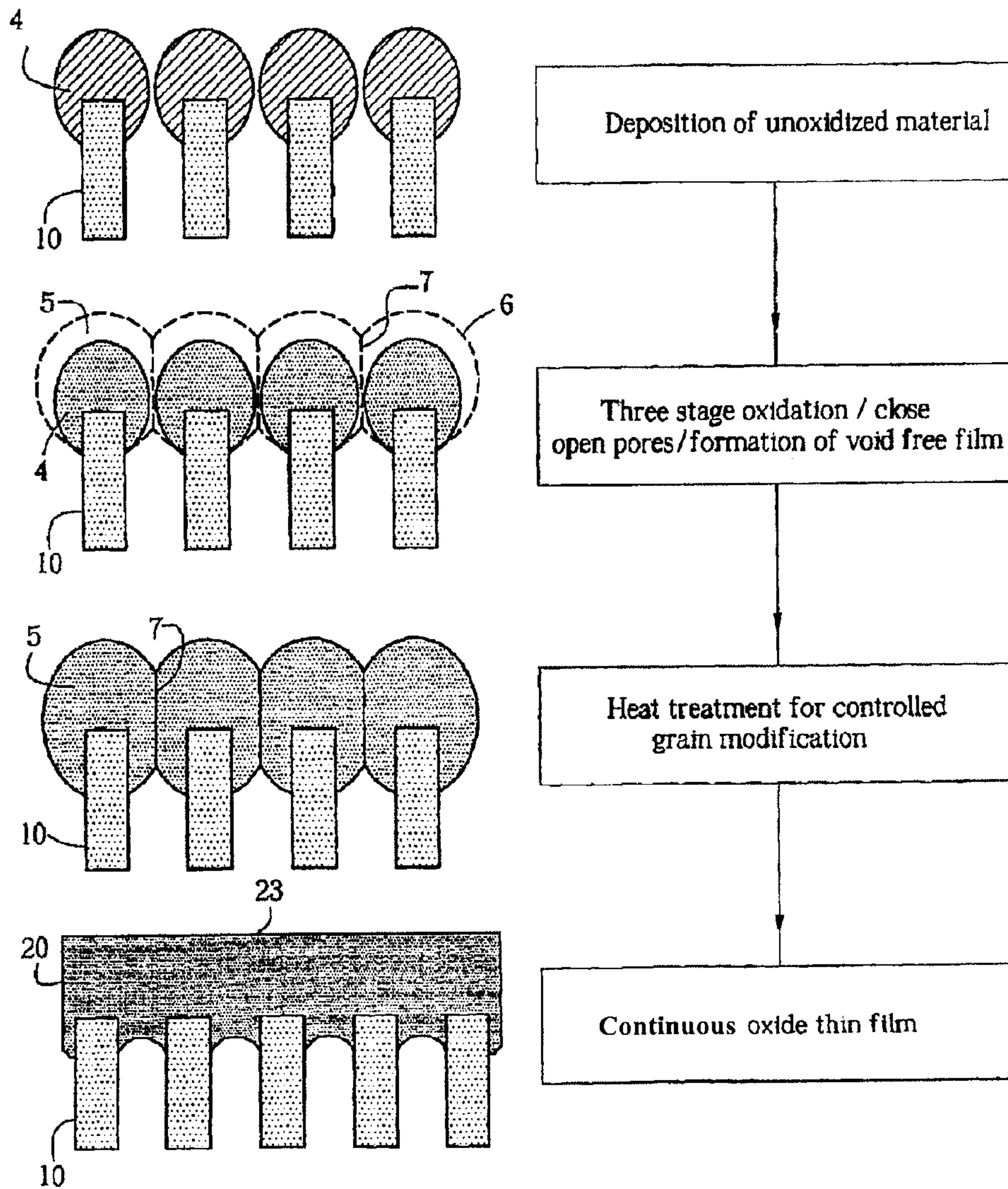


Fig. 2

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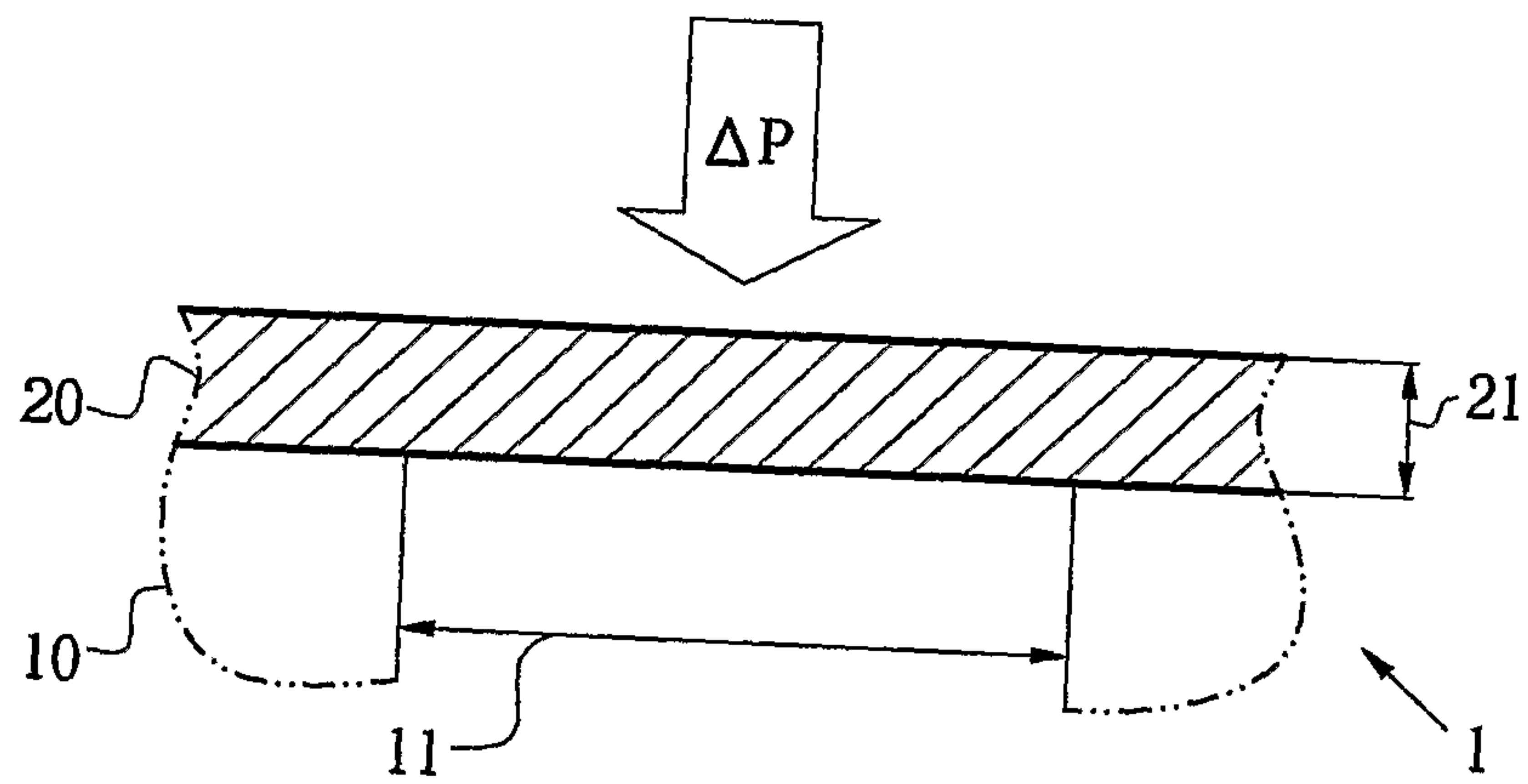


Fig. 3a

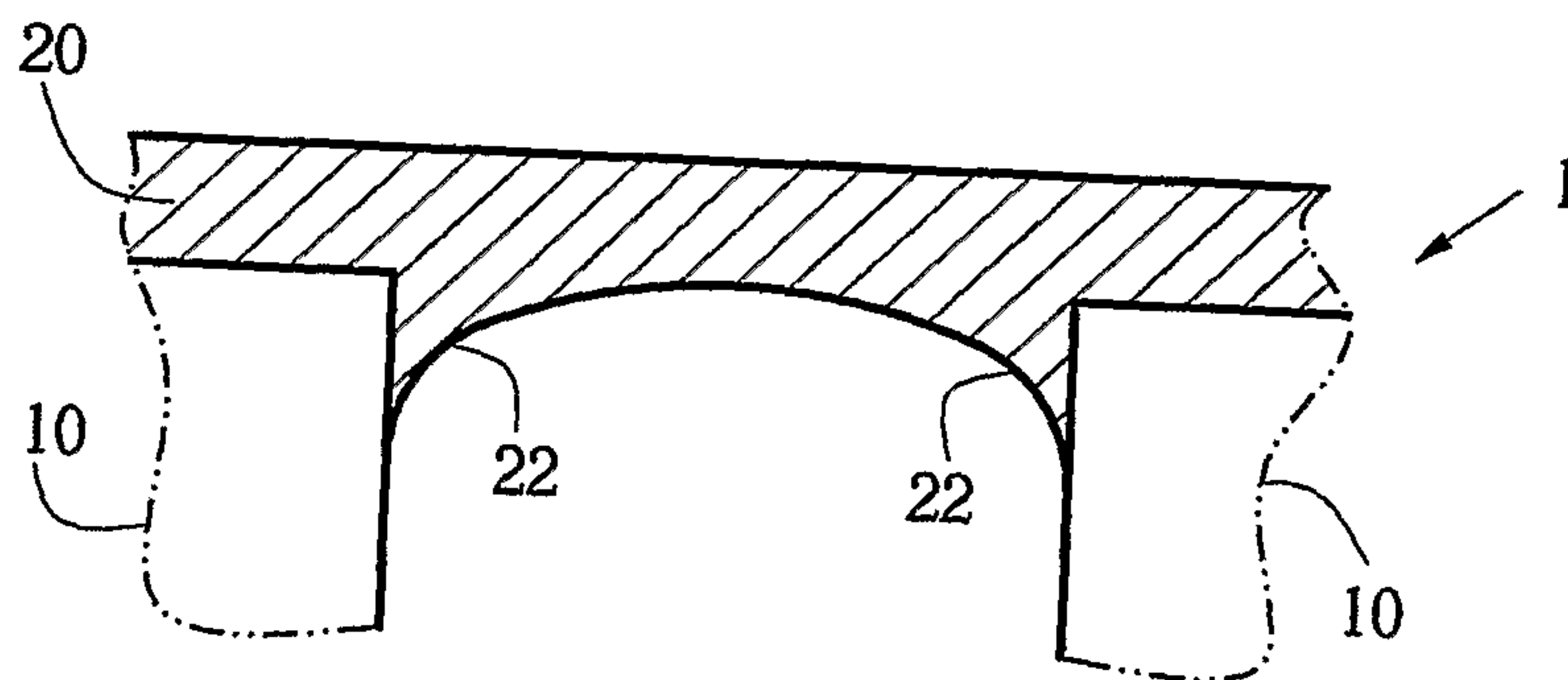


Fig. 3b

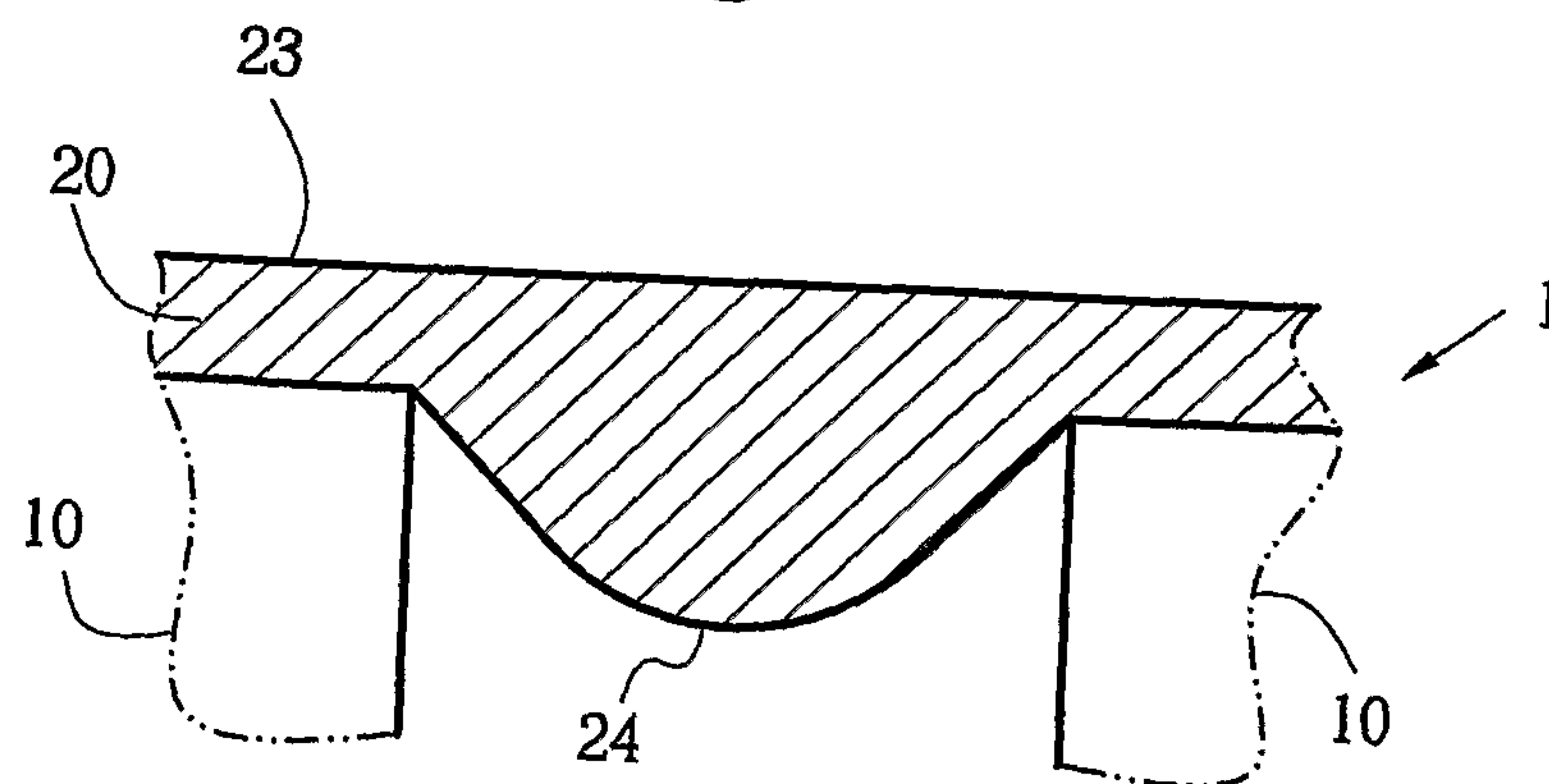


Fig. 3c

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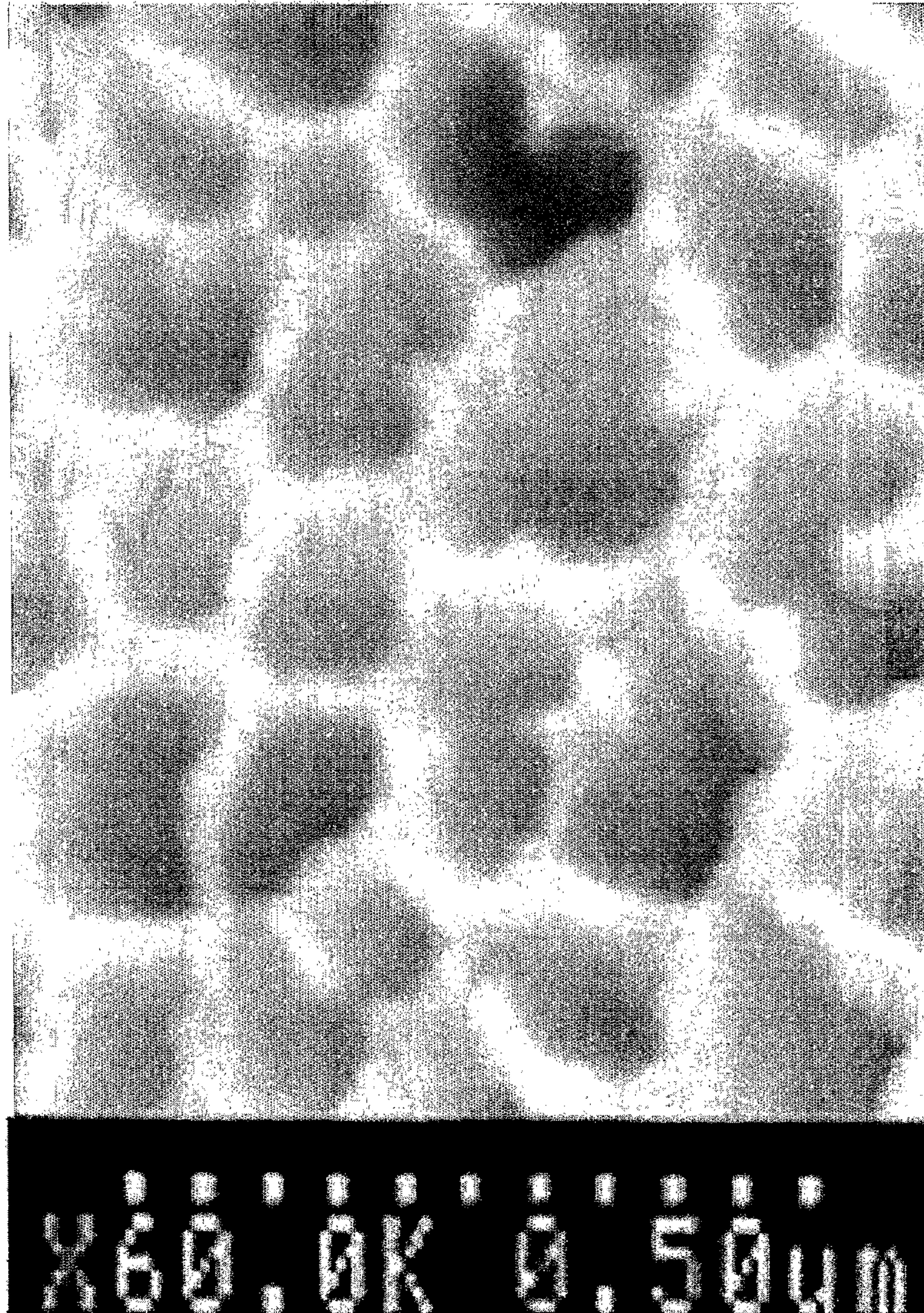


Fig. 4

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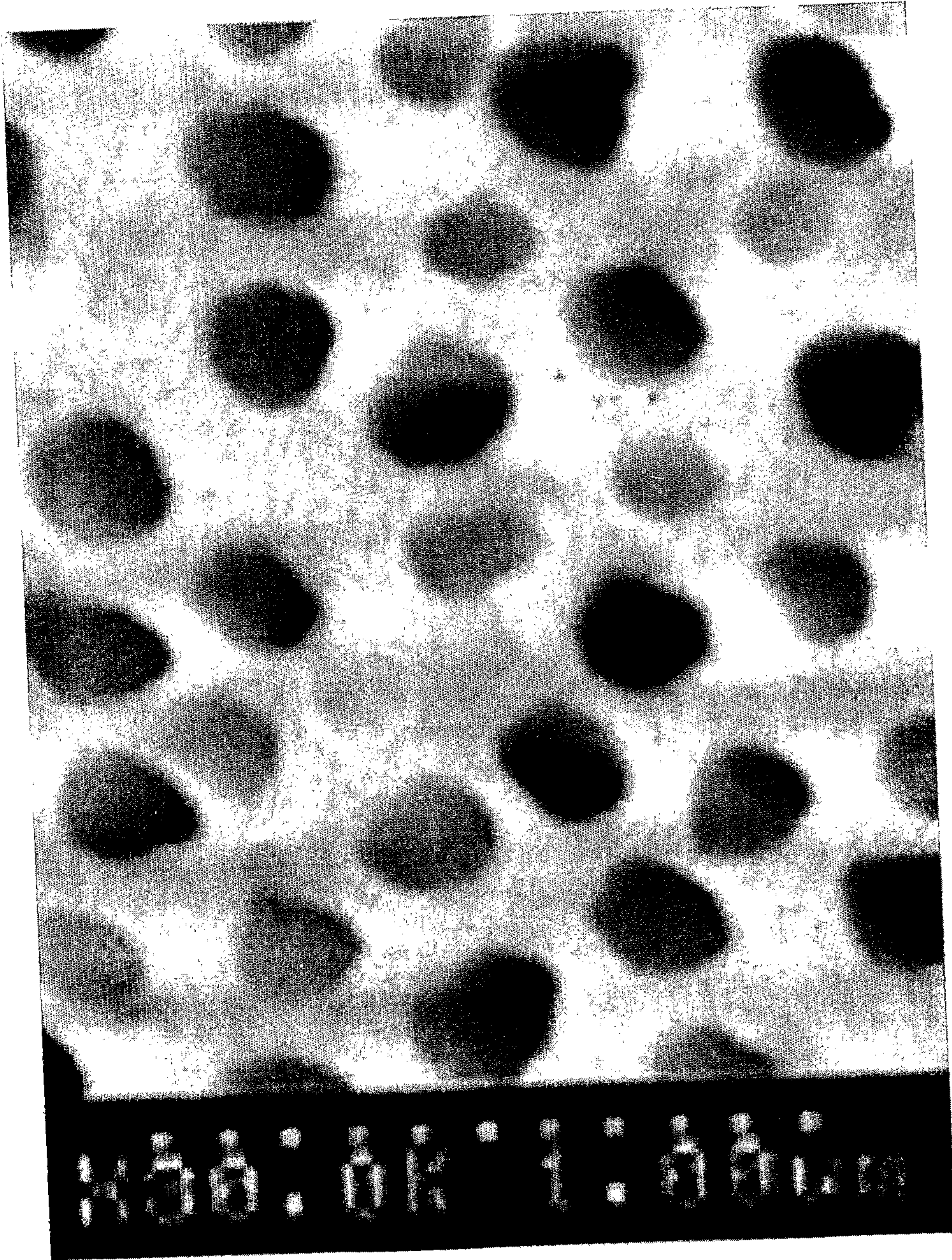


Fig. 5

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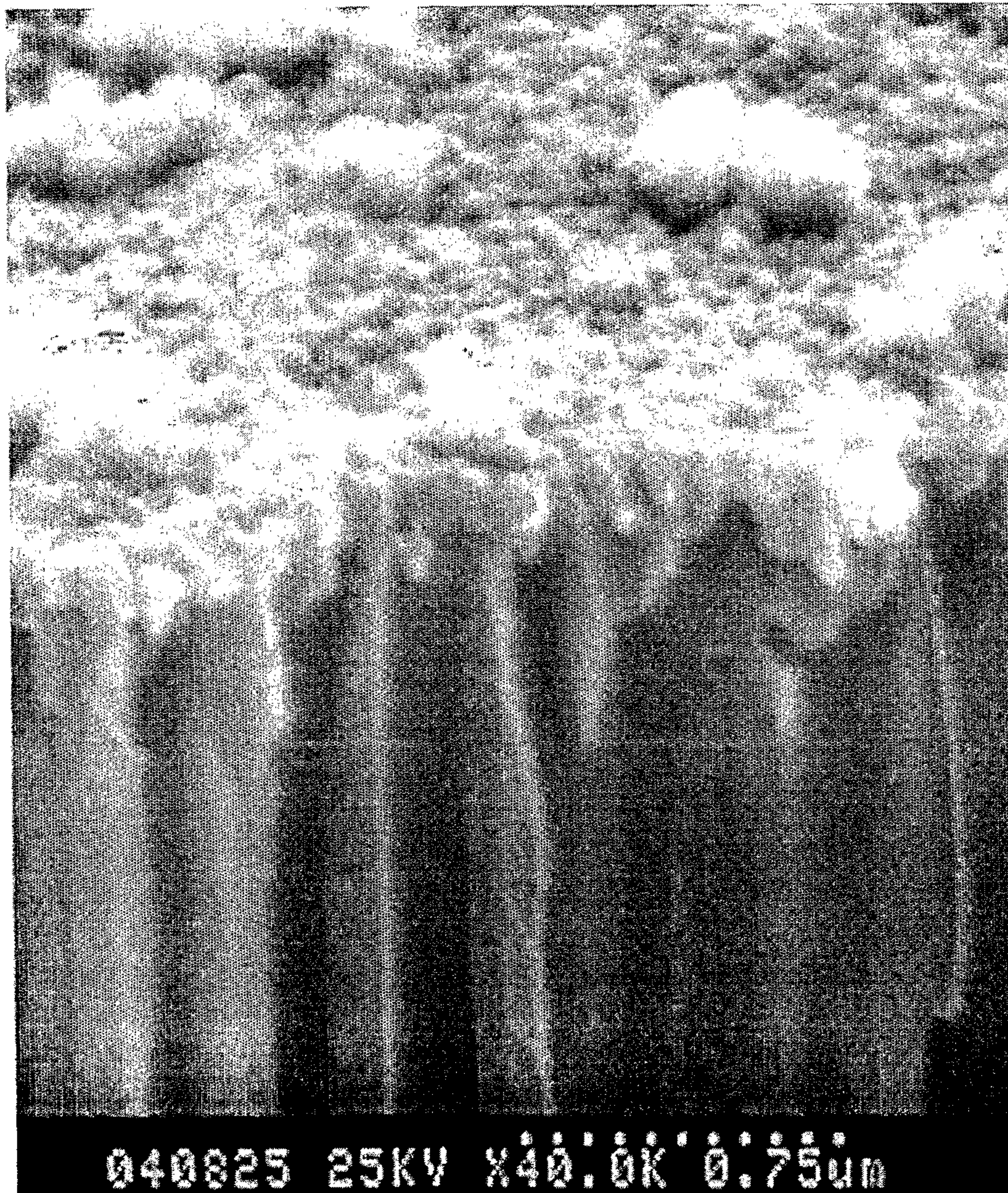


Fig. 6

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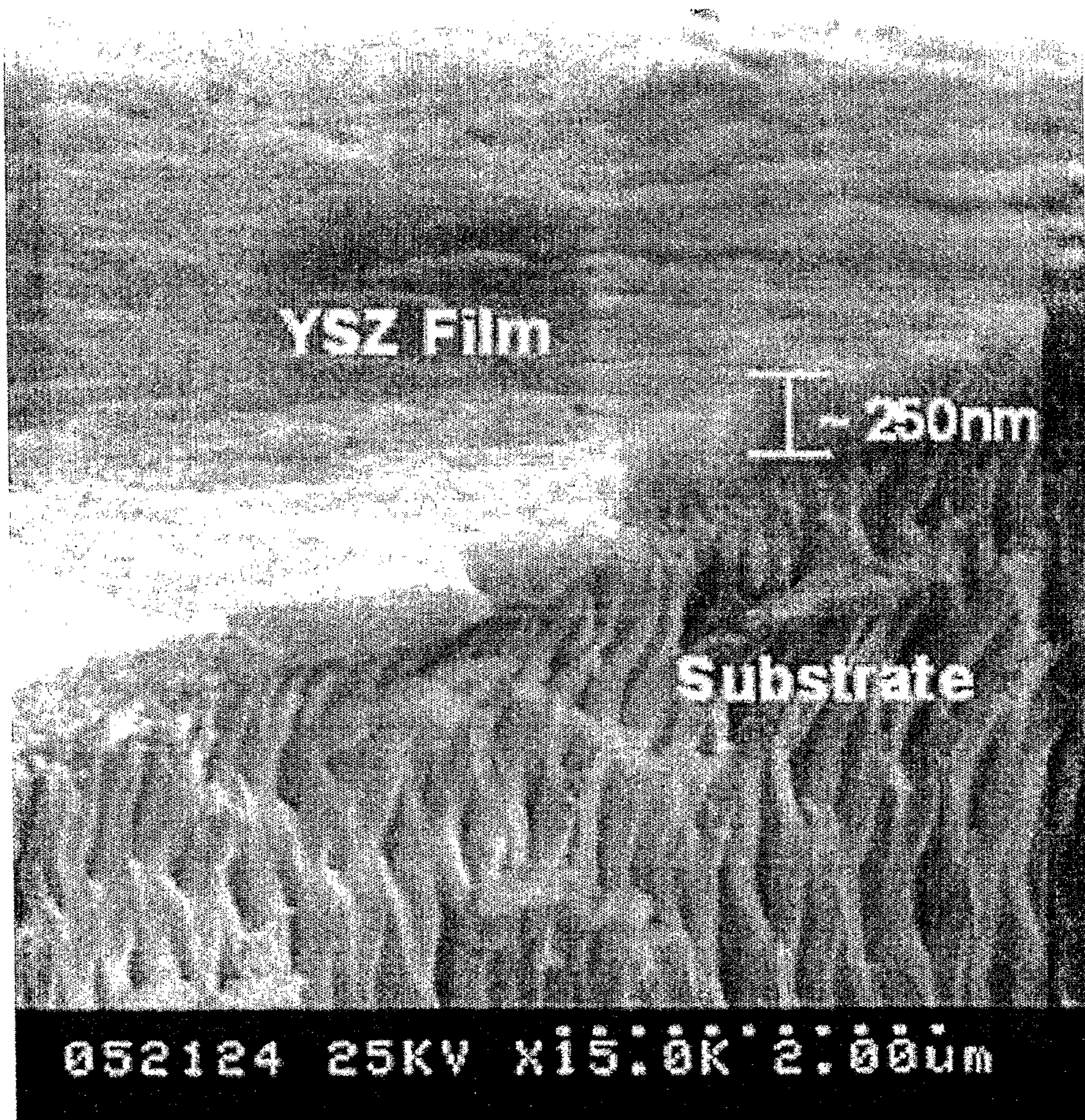


Fig. 7

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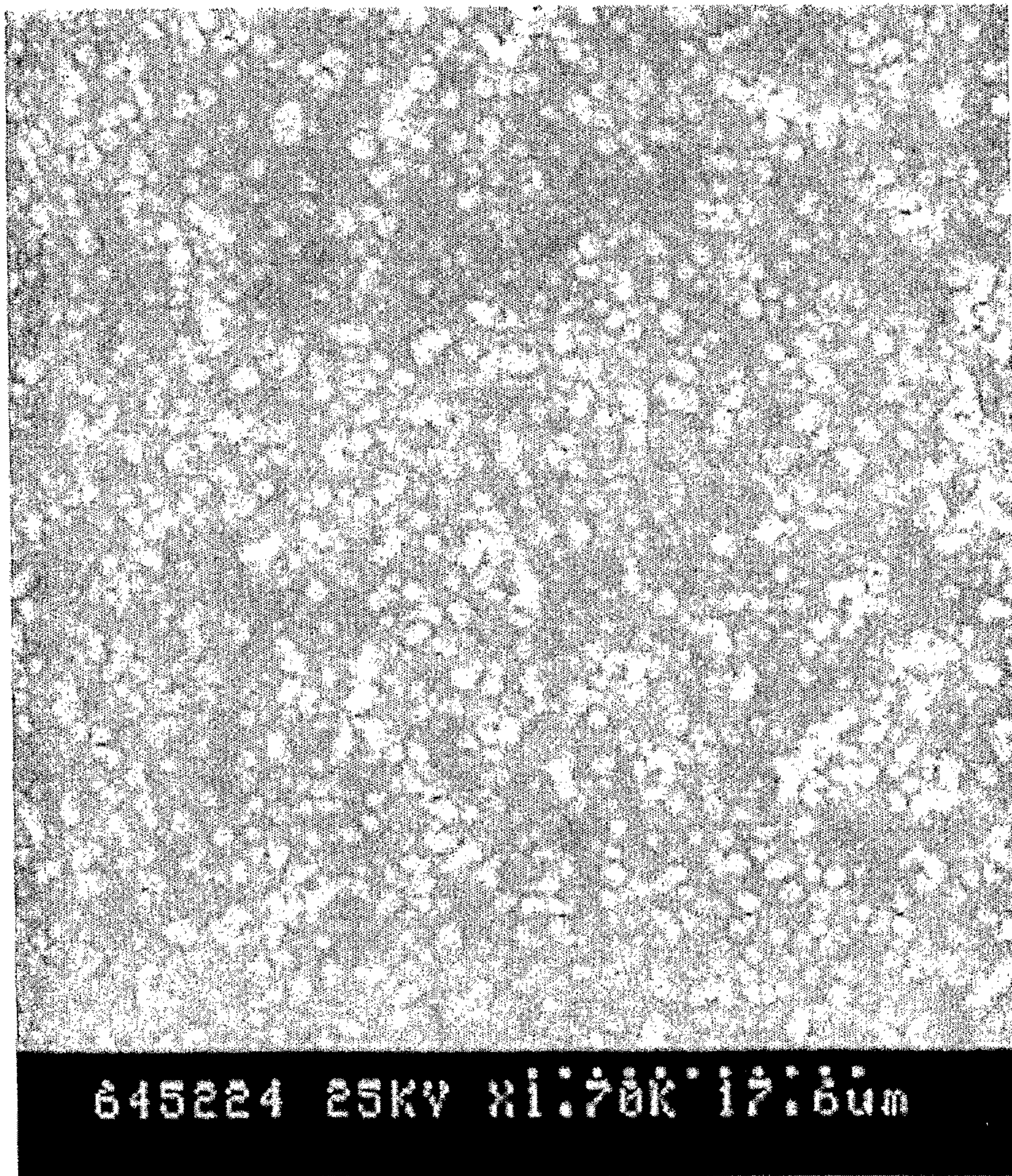


Fig. 8

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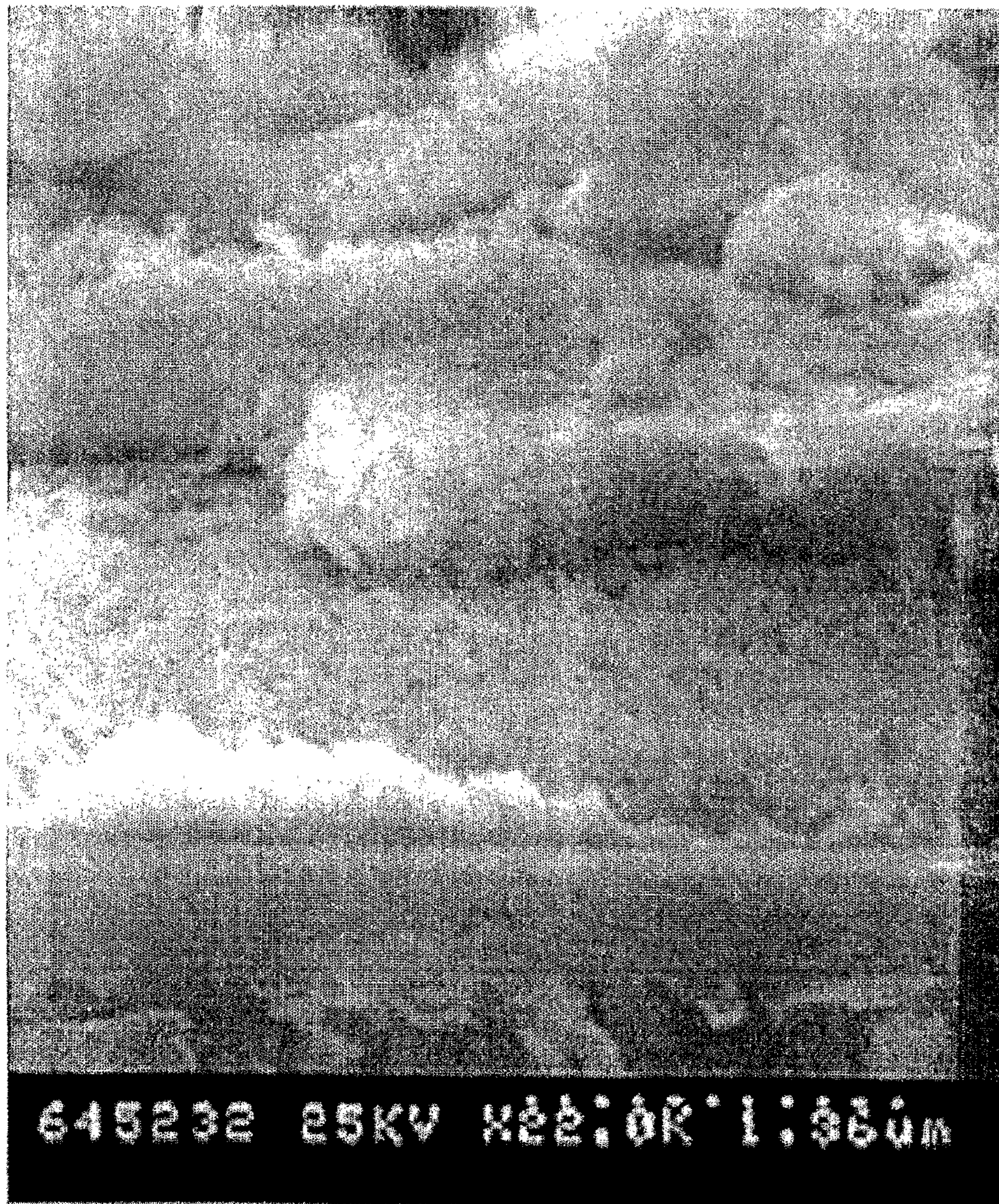


Fig. 9

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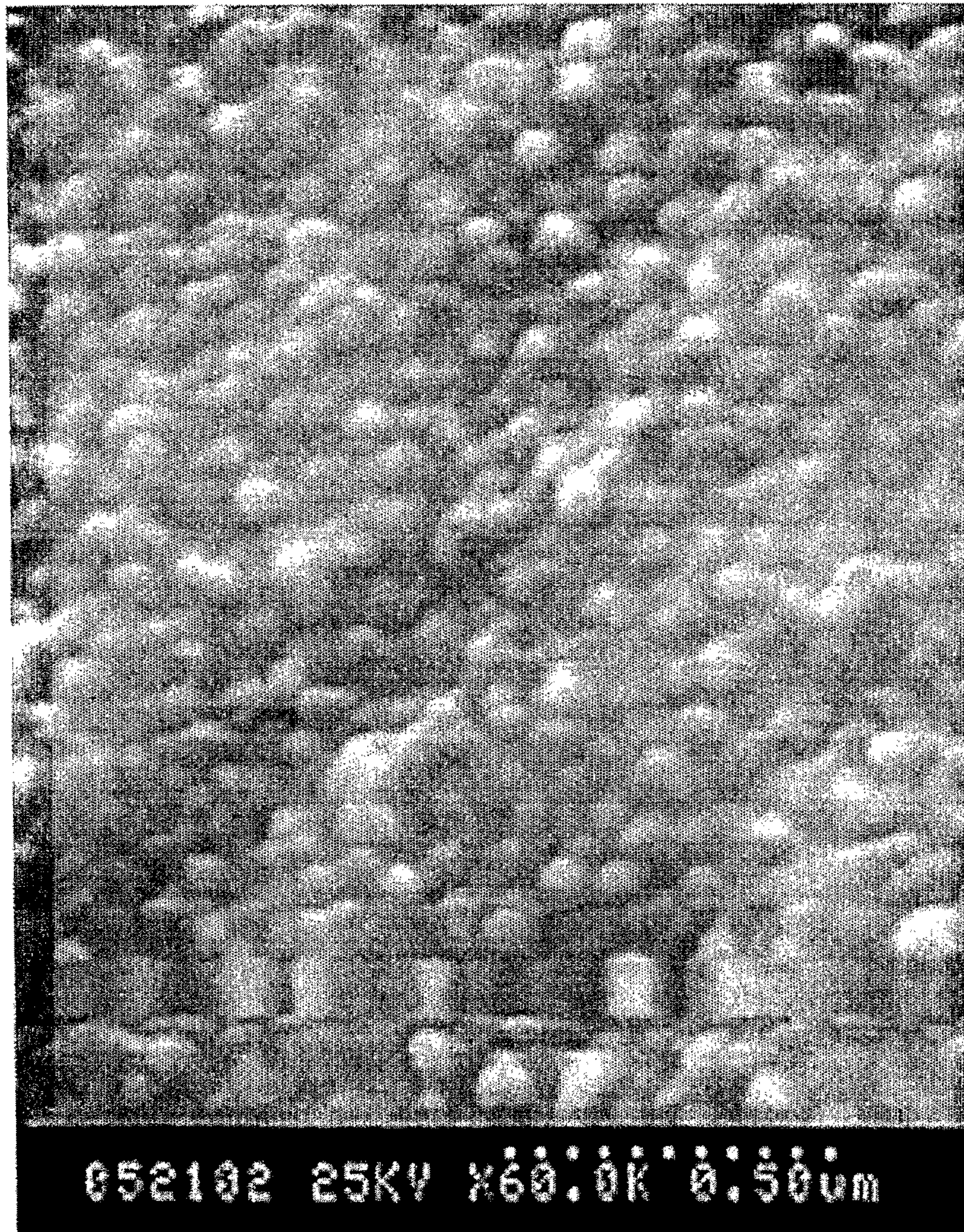


Fig. 10

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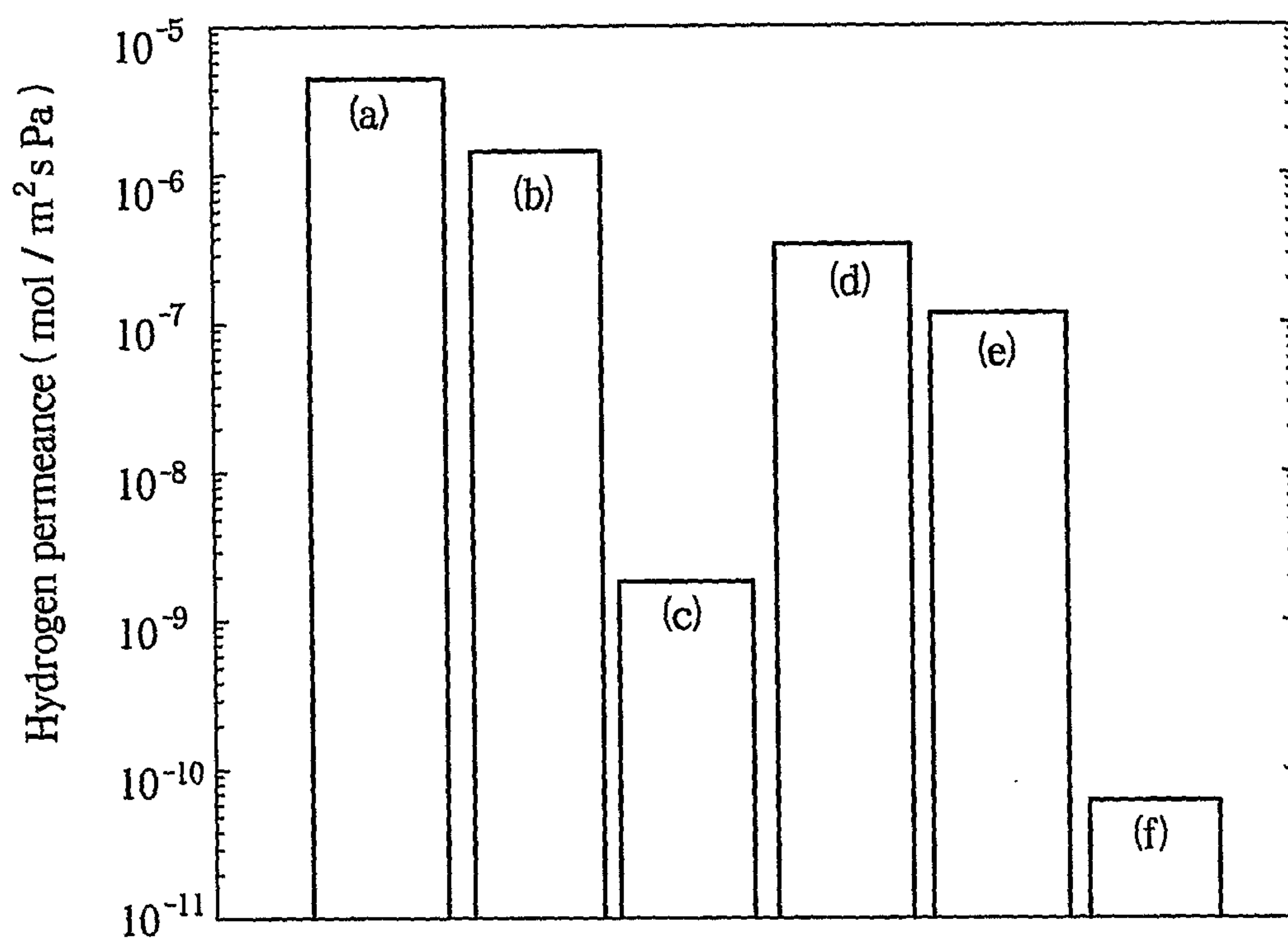


Fig. 11a

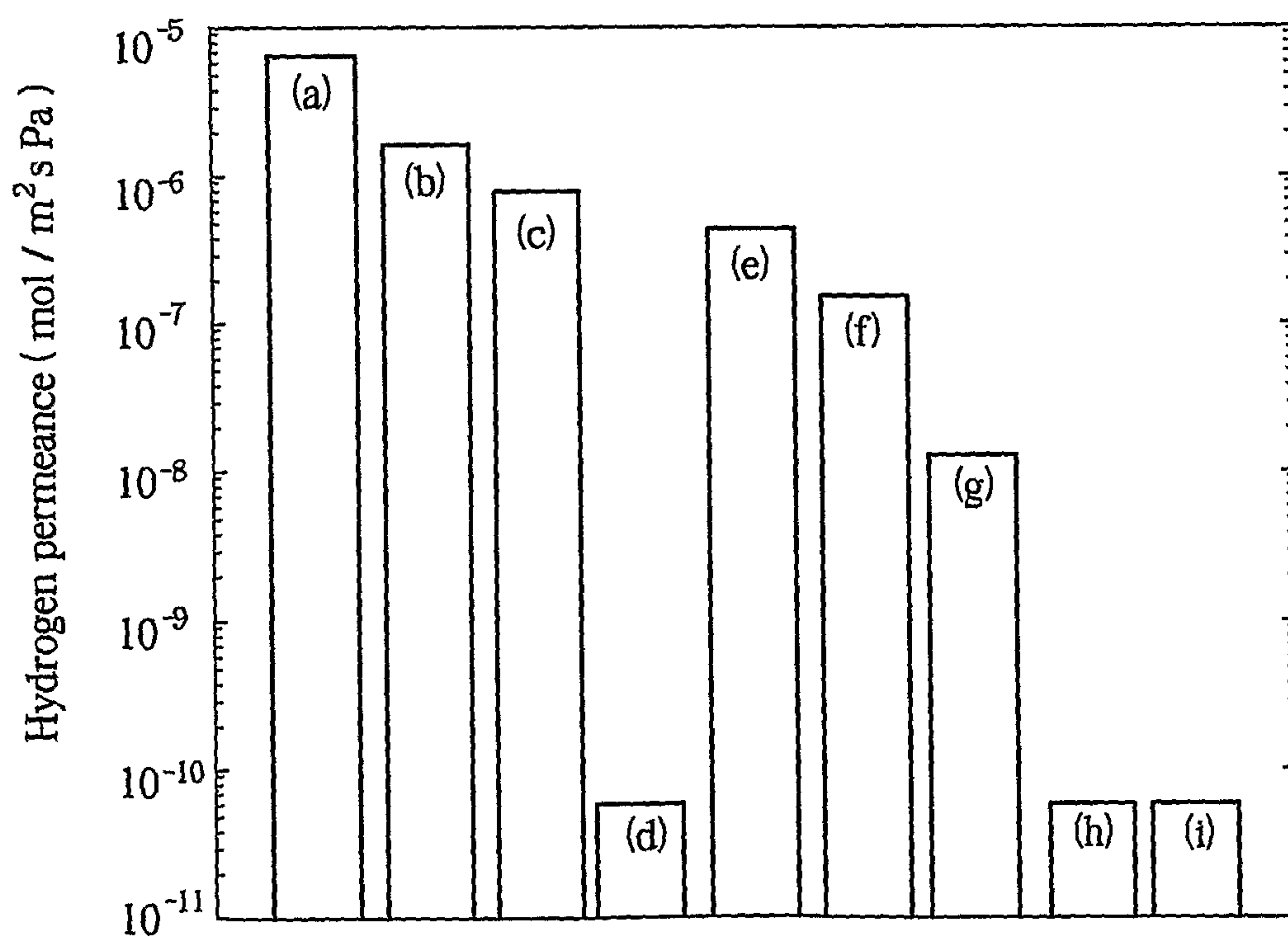


Fig. 11b

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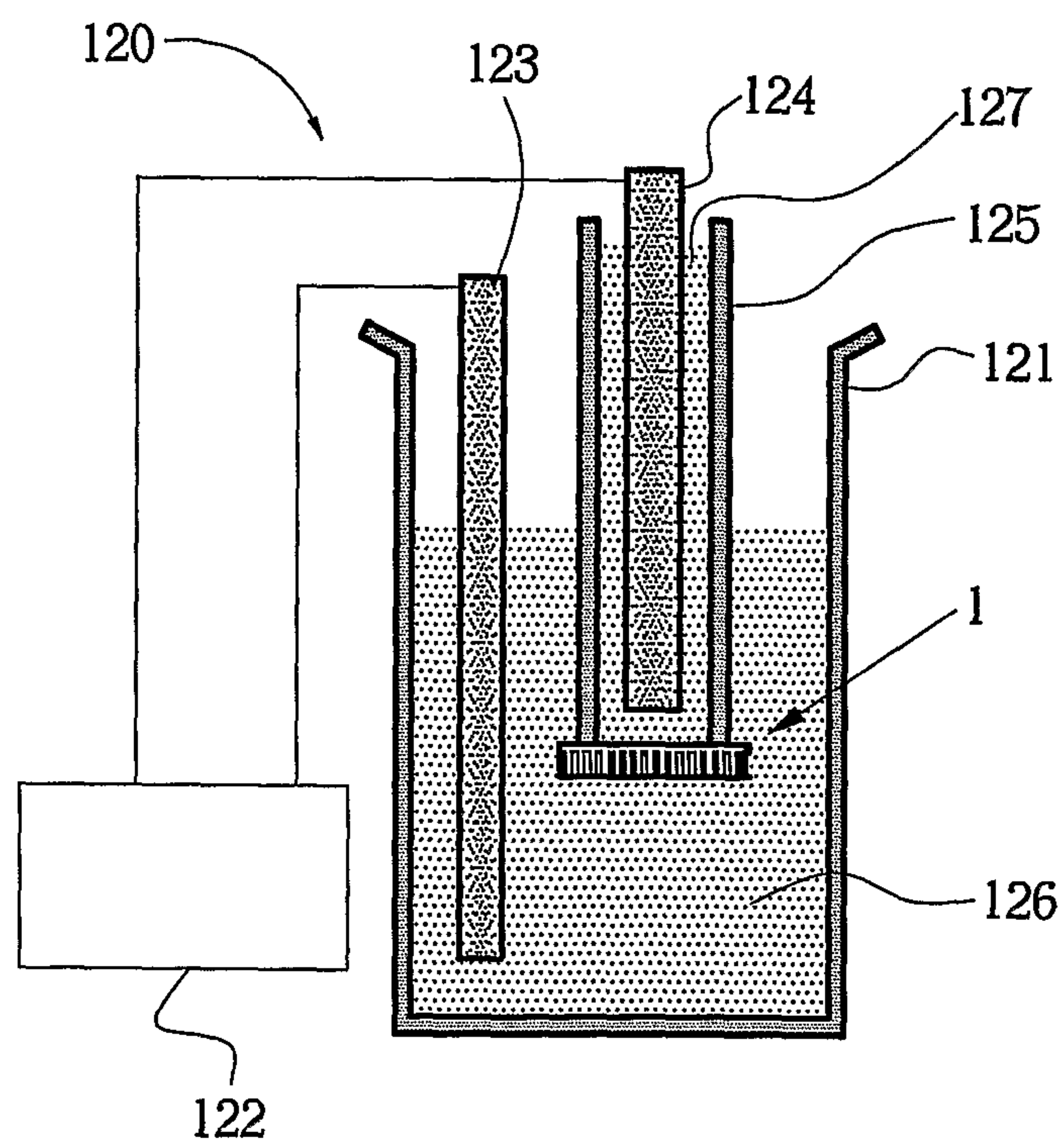


Fig. 12

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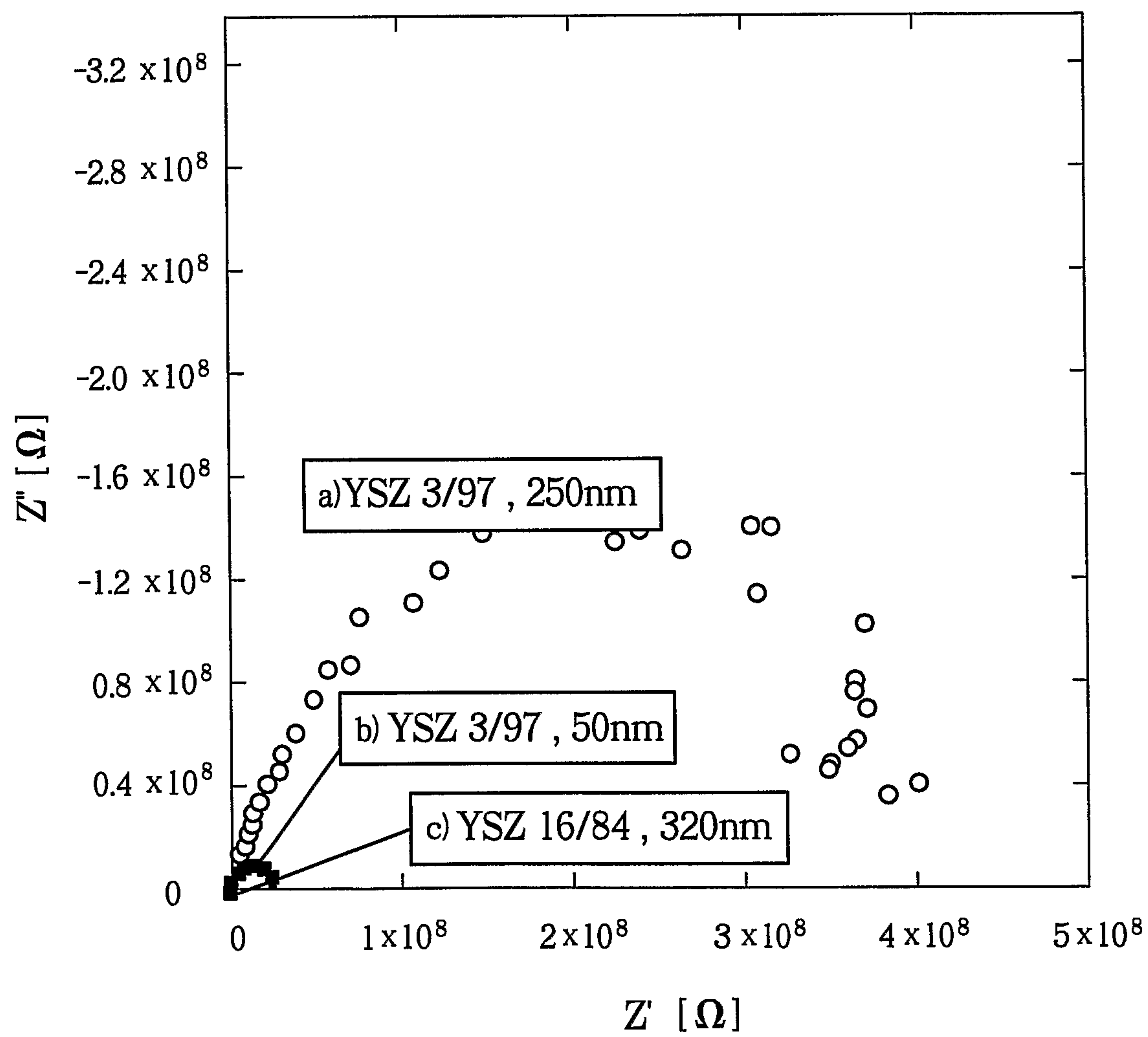


Fig. 13

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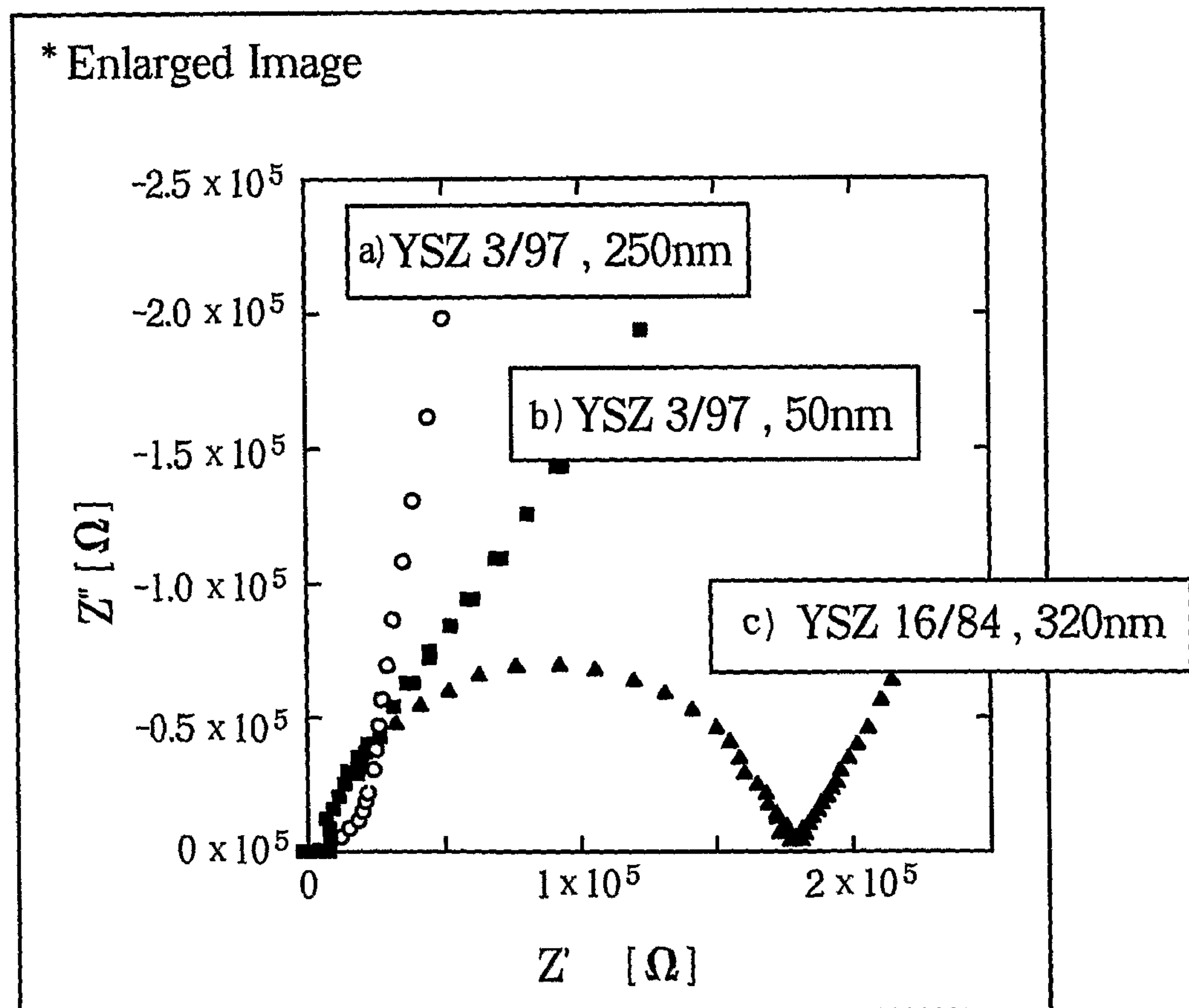


Fig. 14

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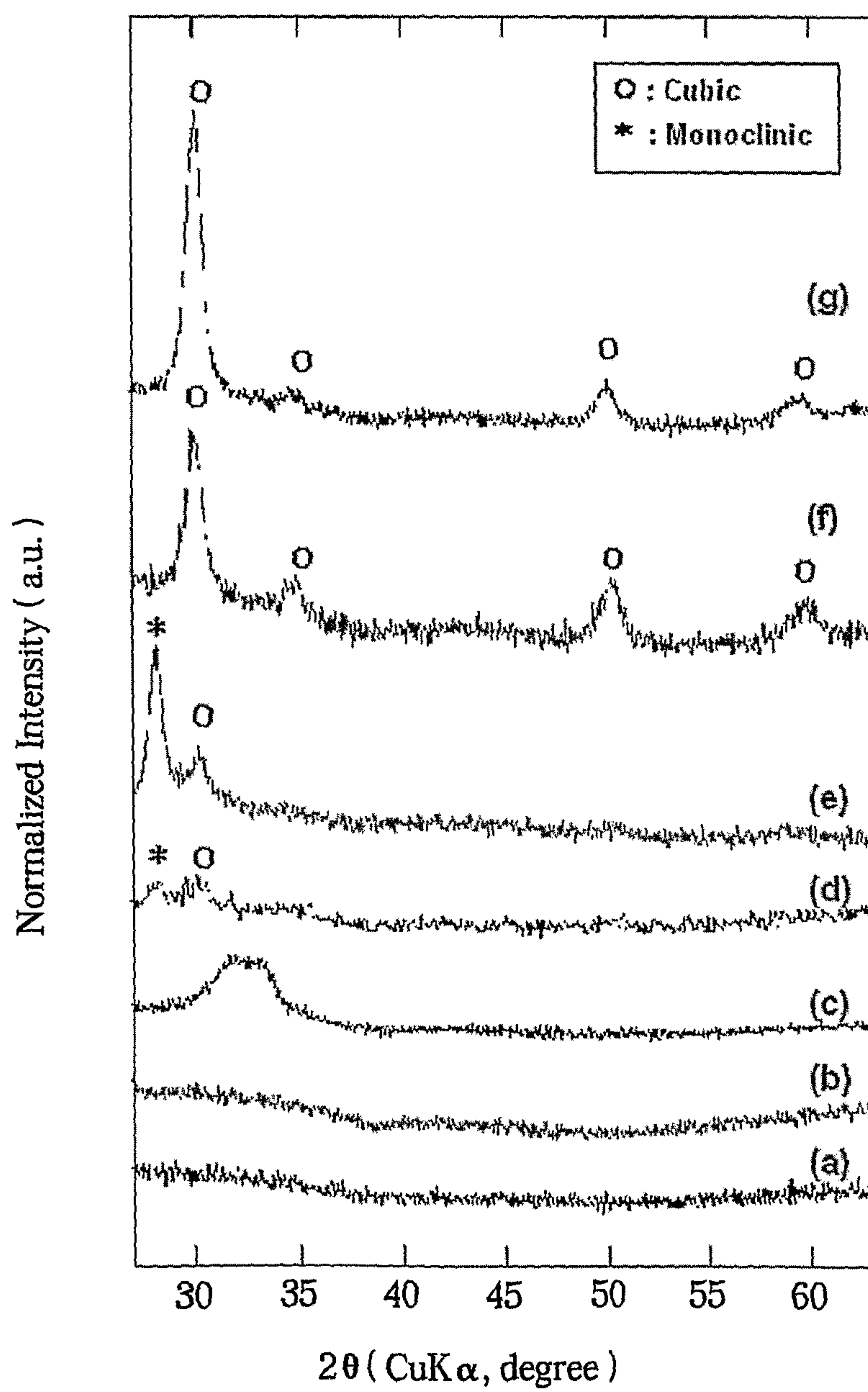


Fig. 15

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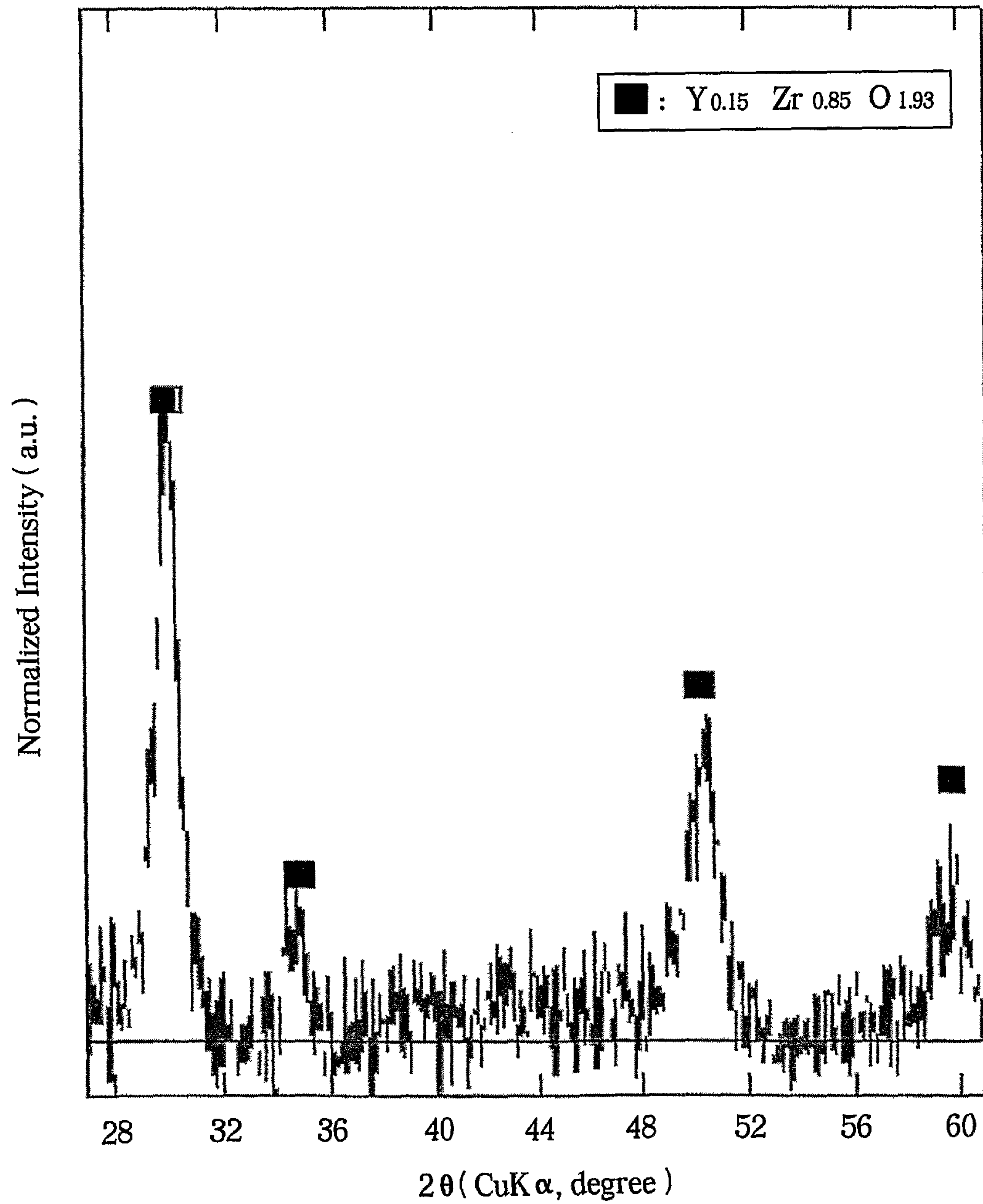


Fig. 16

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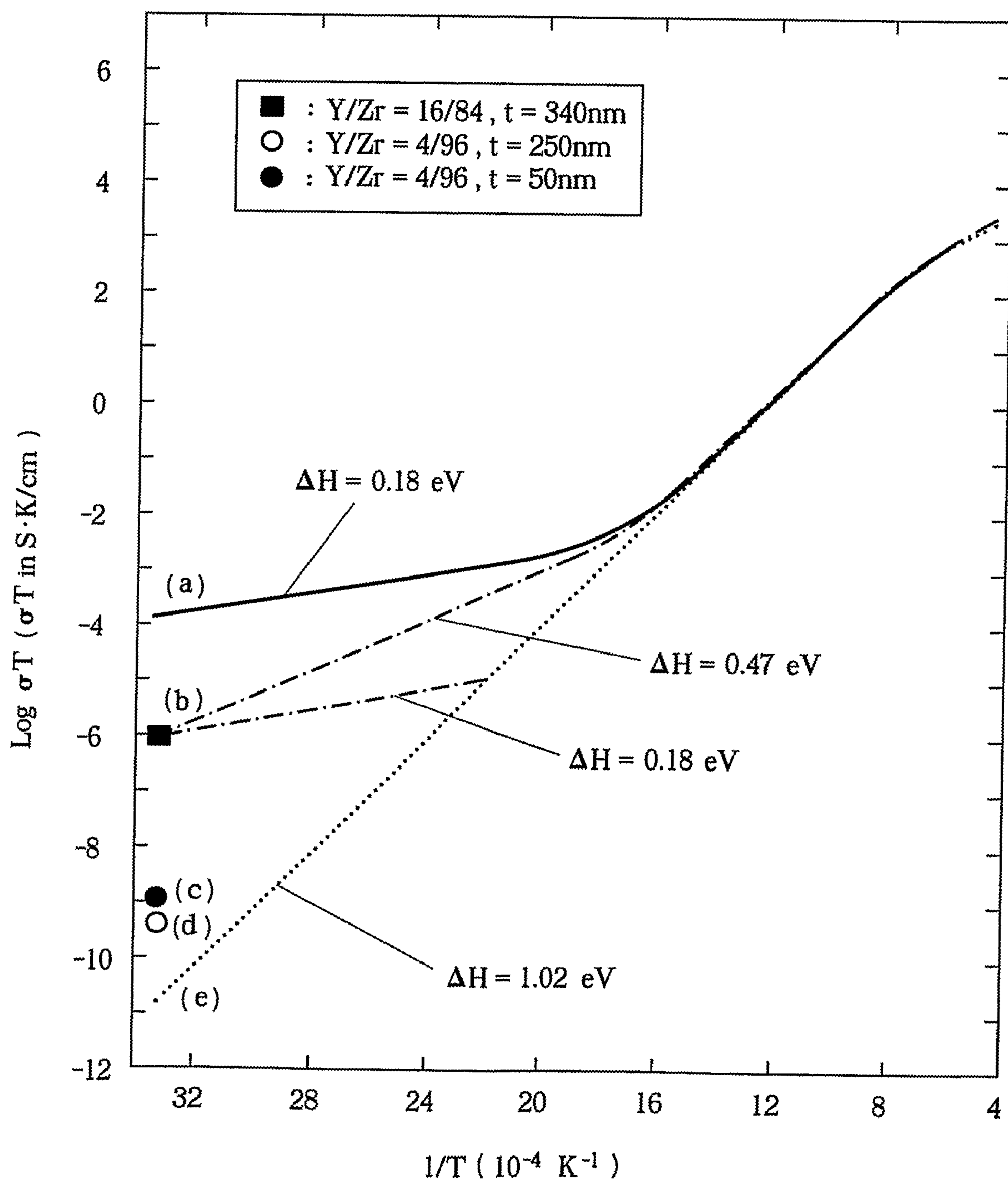


Fig. 17

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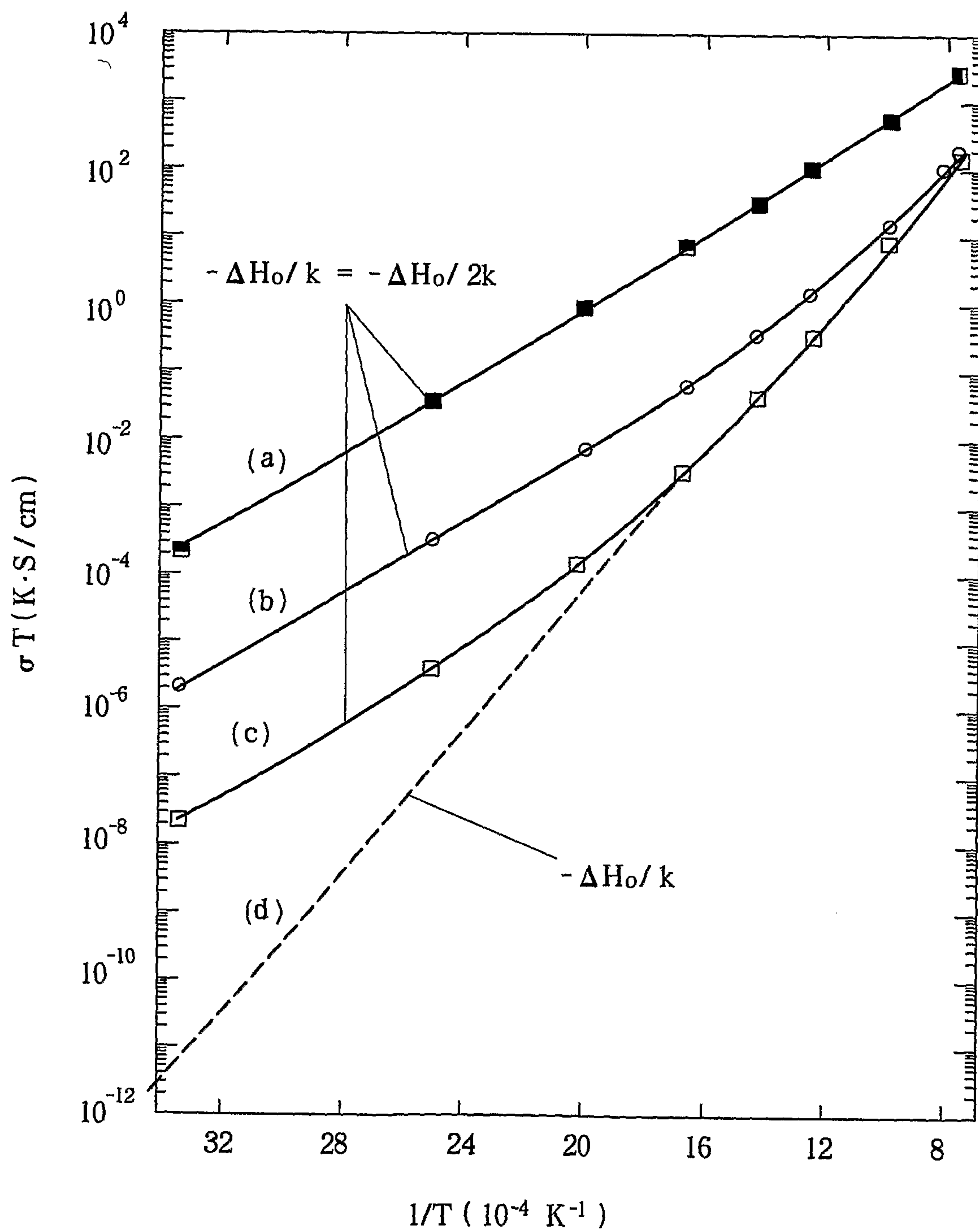


Fig. 18

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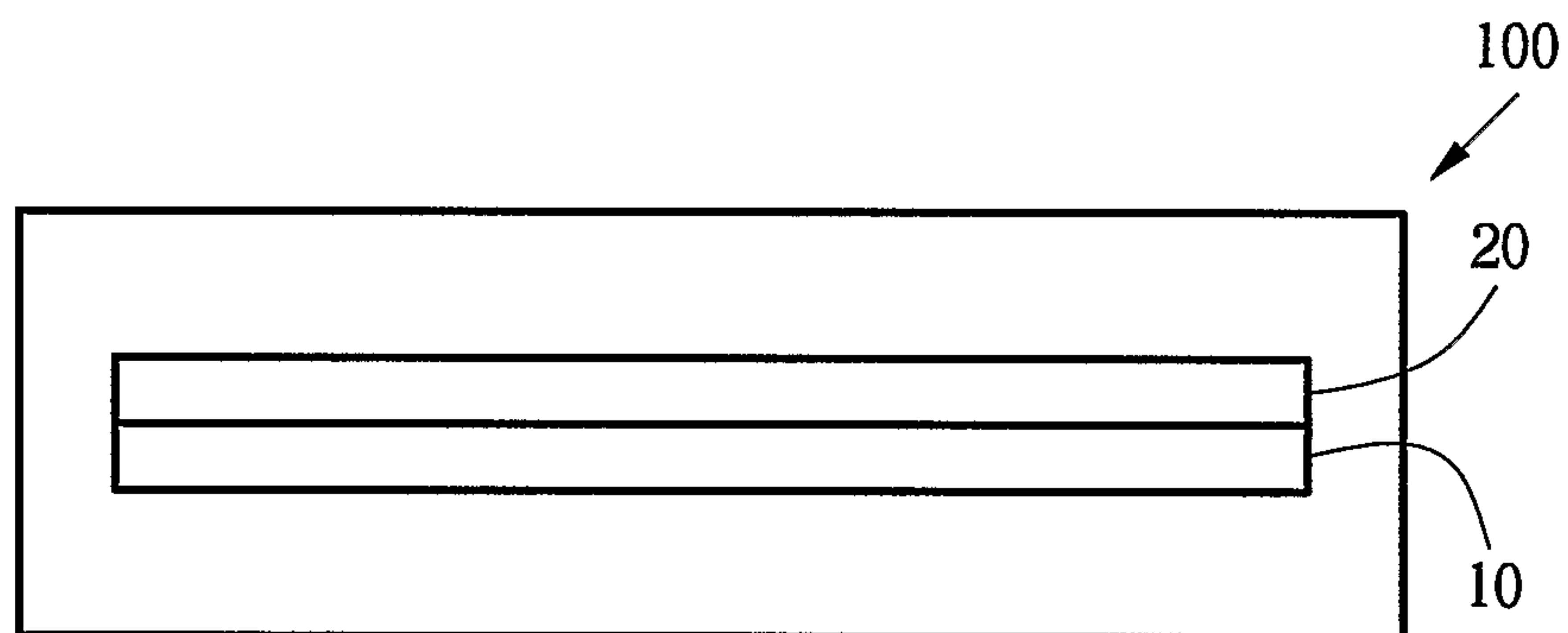


Fig. 19

YSZ Film

I - 250nm

Substrate

052124 25KV X15.0K 2.00um