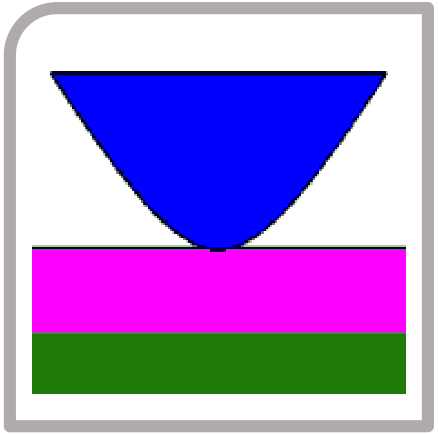
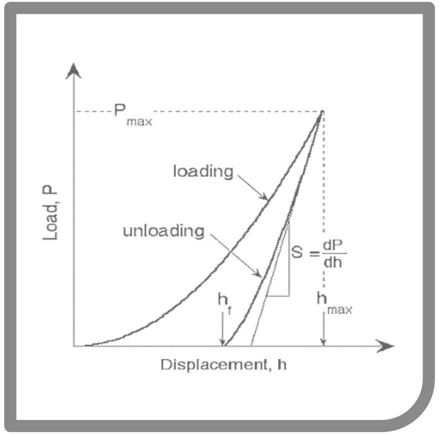




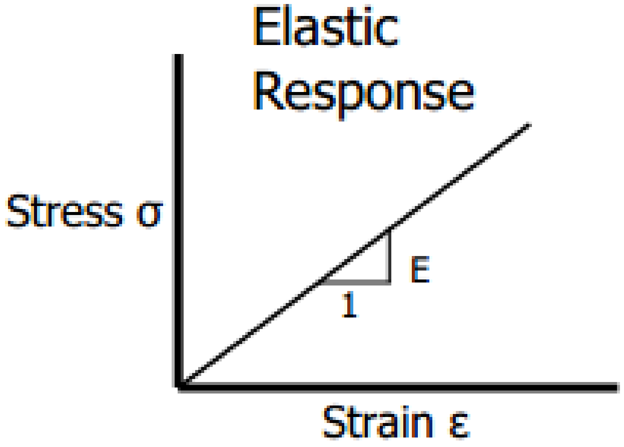
Nano-mechanics for Intelligent Materials (2/2)

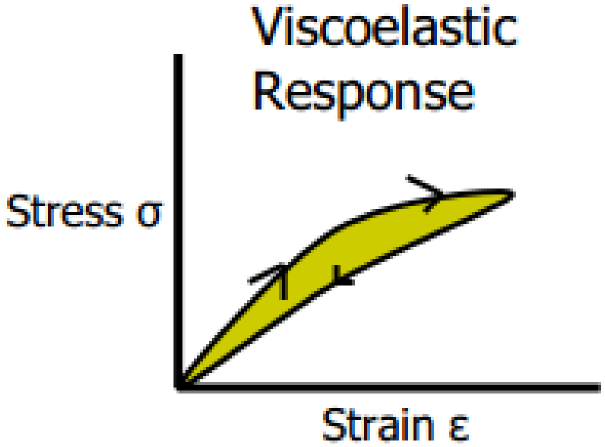
Giorgio MATTEI

Viscoelasticity

- **Viscoelastic materials** exhibit the characteristics of both elastic and viscous materials
  - Viscosity → resistance to flow (damping)
  - Elasticity → ability to revert back to the original shape

- **Elastic vs. viscoelastic** stress-strain response



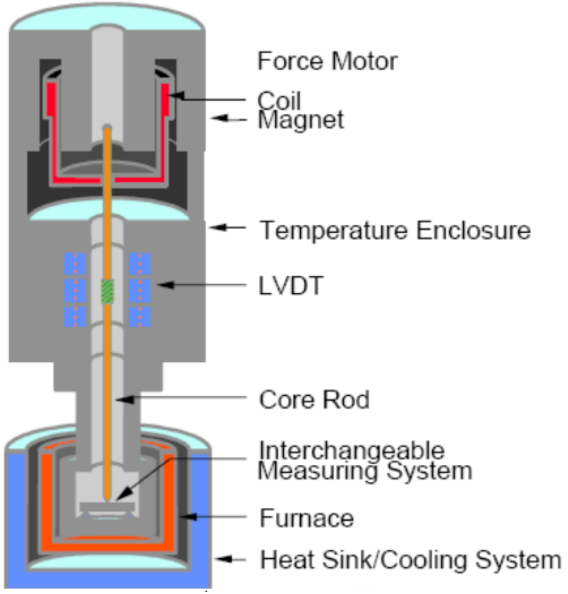


Methods to characterise viscoelasticity

- **Time** domain
  - Creep response
  - Stress relaxation
- **Frequency** domain
  - Dynamic mechanical analysis (DMA)
  - Dynamic mechanical thermal analysis (DMTA)
- **Strain-rate** domain
  - Epsilon dot Method
- **Stress-rate** domain
  - Sigma dot Method

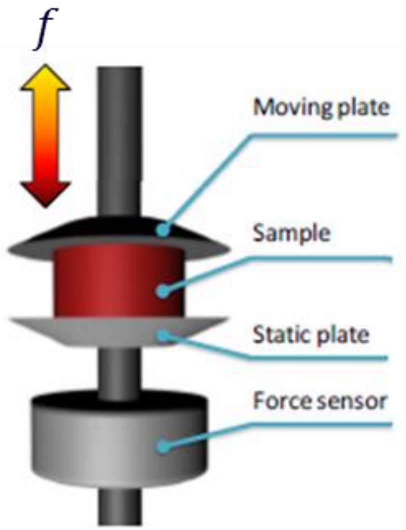
Dynamic mechanical analysis (DMA)

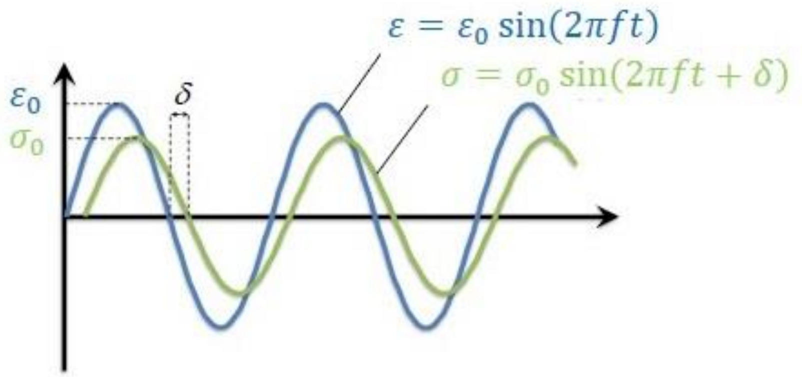
- Dynamic mechanical analysis (DMA) is a standard **force-triggered method** to **determine viscoelastic properties** of materials by **applying a small amplitude cyclic strain** on a sample and **measuring the resultant cyclic stress response**.



Dynamic mechanical analysis (DMA)

- For a given **sinusoidal strain input** the resulting **stress will be sinusoidal if the applied strain is small enough** so that the tissue can be approximated as **linearly viscoelastic**.

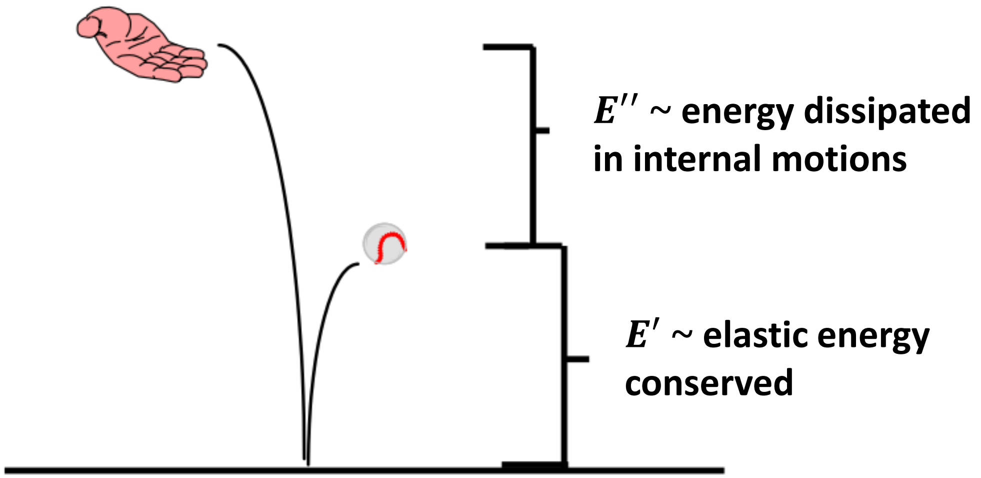




Viscoelastic material response is characterised by a **phase lag** (**δ**) between the strain input and the stress response, which is comprised **between 0°** (**purely elastic**) and **90°** (**purely viscous**). This phase lag is **due to** the excess **time necessary for molecular motions and relaxations** to occur.

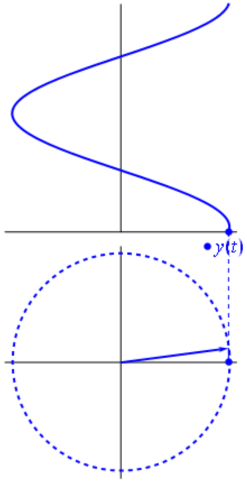
Complex, storage and loss moduli

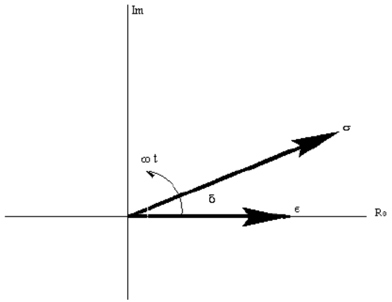
- The dynamic mechanical properties are quantified with the **complex modulus** (E^*), which can be thought as an **overall resistance** to deformation under dynamic loading. The complex modulus is composed of the **storage** (E', elastic component) and the **loss** (E'', viscous component) moduli, that are **additive under the linear theory of viscoelasticity** ($E^* = E' + iE''$).



- It is convenient to represent the sinusoidal stress and strain functions as complex quantities (called rotating vectors, or **phasors**) with a **phase shift** of **δ**.

$$\varepsilon = \varepsilon_0 \, e^{i\omega t} \quad \sigma = \sigma_0 \, e^{i(\omega t+\delta)}$$





Observable σ and ε can be viewed as the projection on the real axis of vectors rotating in the complex plane at the same frequency ω

Rotating vector representation of harmonic stress and strain

$$\boldsymbol{E}^* = \frac{\sigma}{\varepsilon} = \frac{\sigma_0}{\varepsilon_0} e^{i\delta} =$$

$$= \frac{\sigma_0}{\varepsilon_0}(\cos\delta + i\sin\delta) =$$

$$= \boldsymbol{E'} + \boldsymbol{iE''}$$

Storage modulus

$E' = E^* \cos(\delta)$

Loss modulus

$E'' = E^* \sin(\delta)$

$\tan(\delta) = E''/E'$ Damping factor

$\eta' = E''/\omega$ Dynamic viscosity

Test modes

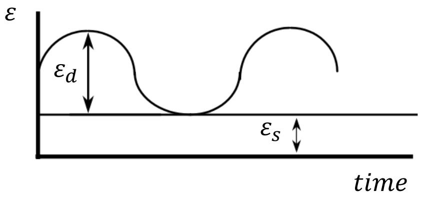
- **Temperature sweep**: Modulus and damping are recorded as the sample is heated
- **Frequency sweep**: Modulus and damping are recorded as the sample is loaded at increasing (or decreasing) frequencies
- **Stress amplitude sweep**: Modulus and damping are recorded as the sample stress is increased
- **Strain amplitude sweep**: Modulus and damping are recorded as the sample strain is increased
- **Combined sweep**: Combinations of above methods

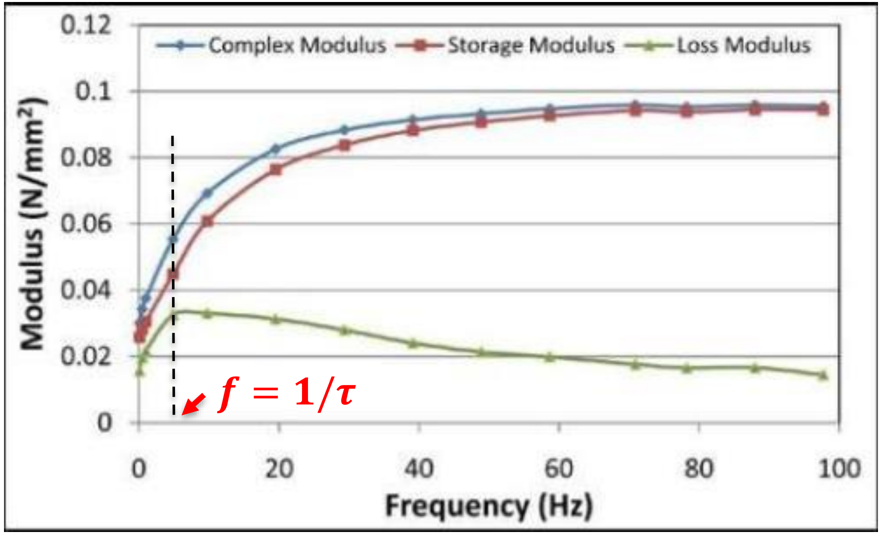
Frequency sweep tests

- A sample is held to a **fixed temperature** and tested at **varying frequency**.

Test parameters:

- Temperature (T)
- Frequency range (f)
- Static strain (ε_s)
- Dynamic strain (ε_d)

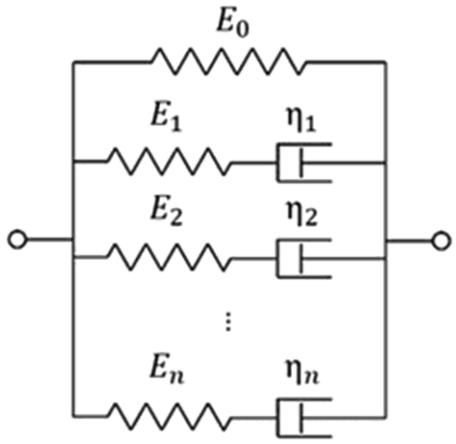




- **Peaks** in $\tan(\delta)$ or E'' with respect to frequency identify the **characteristic relaxation frequencies** of the viscoelastic sample under testing, defined as $f = 1/\tau$ (where τ is the **characteristic relaxation time**)

Lumped models to describe **linear viscoelastic response**

- The most general form of linear viscoelastic model is called the **Generalised Maxwell (GM)** model and consists of a **pure spring ($\boldsymbol{E_0}$)** with $\boldsymbol{n}$ **Maxwell arms** (i.e. spring E_i in series with a dashpot η_i) assembled **in parallel**, thus defining a set of $\boldsymbol{n}$ **different characteristic relaxation times** (i.e. $\boldsymbol{\tau_i = \eta_i / E_i}$)



$$H_{GM}(s) = \frac{\bar{\sigma}}{\bar{\epsilon}} = E_0 + \sum_{i=i}^{n} \frac{E_i \, \eta_i \, s}{E_i + \eta_i \, s}$$

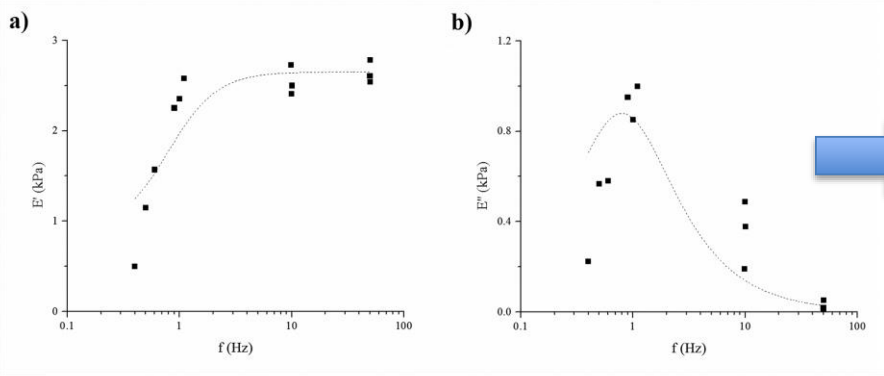
GM model transfer function in the Laplace domain

Lumped parameters derivation from frequency sweep

- Calculate the **complex conjugate of the GM modulus** ($\boldsymbol{E^*_{GM}}$) by substituting $s = i\,\omega = i\,2\pi f$ in $H_{GM}(s)$, then **split the expression into** its **real (*Re*)** and **imaginary (*Im*)** parts to obtain the **frequency-dependent relations** for the **storage** and **loss** moduli, respectively

$$E^*_{GM}(f) = \underbrace{\left(E_0 + \sum_{i=i}^{n} \frac{4\,E_i\,\eta_i^2\,f^2\,\pi^2}{E_i^2\ +\ 4\,\eta_i^2 f^2 \pi^2}\right)}_{\boldsymbol{E'(f)}} + i\underbrace{\left(\sum_{i=i}^{n} \frac{2\,E_i^2\,\eta_i\,f\,\pi}{E_i^2\ +\ 4\,\eta_i^2 f^2 \pi^2}\right)}_{\boldsymbol{E''(f)}}$$

- **Global fitting** with **shared parameters** (χ^2 minimisation)



Nano-DMA

- Experimental data obtained for a given frequency can be used to compute the frequency-dependent storage (E') and loss (E'') moduli as:

$$\frac{E'(f)}{(1-\upsilon^2)} = \frac{P_0}{h_0} cos(\phi) \frac{1}{\sqrt{hR}}$$

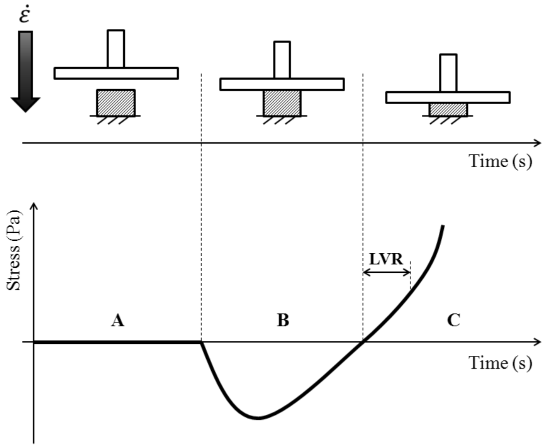
$$\frac{E''(f)}{(1-\upsilon^2)} = \frac{P_0}{h_0} sin(\phi) \frac{1}{\sqrt{hR}}$$

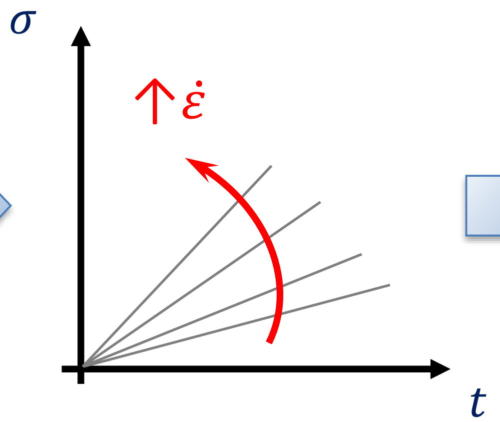
Herbert EG et al, *J. Phys. D. Appl. Phys.* 41 (2008)

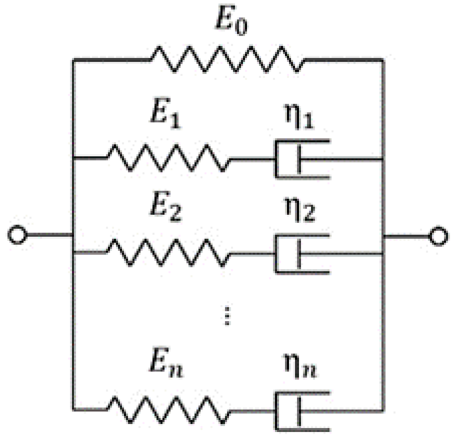
- Frequency spectra of storage and loss moduli can then be fitted to lumped parameter rheological models to derive material viscoelastic constants as previously described

Epsilon dot method ($\dot{\varepsilon}M$)

$\dot{\varepsilon}M$ paradigm: characterise the material **viscoelastic behaviour** testing samples at **different constant strain rates** ($\dot{\varepsilon}$), then **analyse $\sigma(t)$ curves** within the **LVR**







1. Bulk test at constant $\dot{\varepsilon}$

- ✓ **Force-displacement time recording** starts **prior to** sample **contact** → **no pre-load**
- ✓ **Short test** duration → **no sample deterioration**
- ✓ **LVR** determined through **measured σ-ε curves**

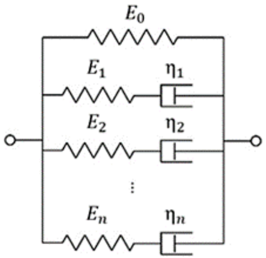
2. Global fitting of **LVR stress-time** (σ-t) **series** measured **at different $\dot{\varepsilon}$**

3. Lumped parameter estimation

Tirella A et al., *JBMR A* 102 (2014)

$\sigma(t)$ response to a constant $\dot{\varepsilon}$ for the $\dot{\varepsilon}M$ global fitting

1. Calculate the **transfer function** of a **lumped parameter model** in the Laplace domain



$\tau_i = \eta_i / E_i$ i^{th} relaxation time

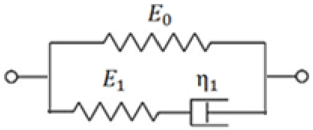
$$H_{GM}(s) = \frac{\bar{\sigma}}{\bar{\epsilon}} = E_0 + \sum_{i=i}^{n} \frac{E_i\,\eta_i\,s}{E_i + \eta_i\,s}$$

2. Derive the **model response** to a **constant $\dot{\varepsilon}$ input** with **amplitude $|\dot{\varepsilon}|$** in the Laplace domain

$$\bar{\sigma} = H_{GM}(s) \cdot \frac{|\dot{\boldsymbol{\varepsilon}}|}{\boldsymbol{s}^2}$$

3. Get the $\boldsymbol{\sigma(t)}$ **response** through **Inverse Laplace transformation**

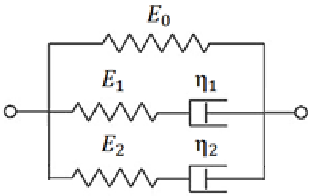
n=1



Maxwell Standard Linear Solid (**SLS**)

$$\sigma_{ind}(t) = \dot{\varepsilon}_{ind}\left[E_0 t + \eta_1\left(1 - e^{-\frac{E_1}{\eta_1}t}\right)\right]$$

n=2



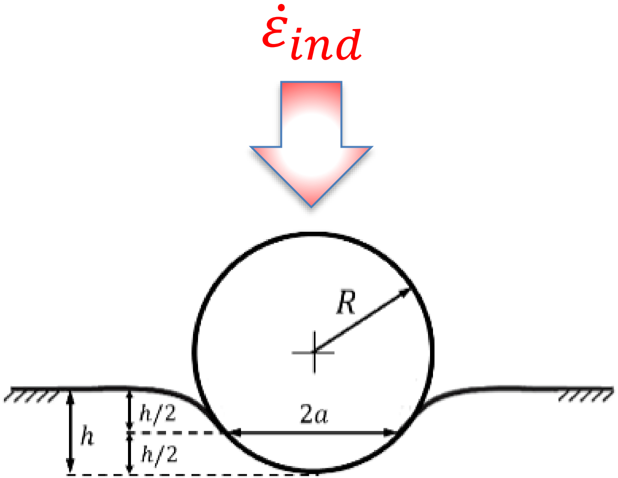
2-arm generalised Maxwell (**GM2**)

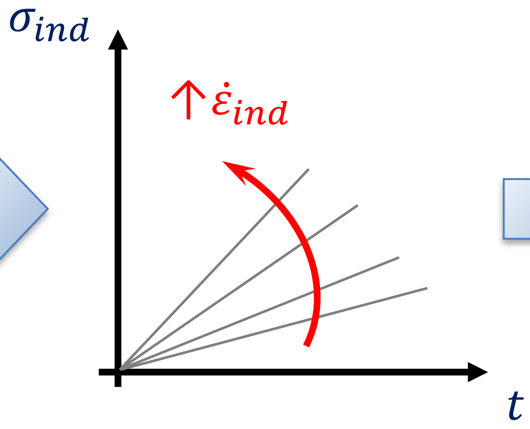
$$\sigma_{ind}(t) = \dot{\varepsilon}_{ind}\left[E_0 t + \eta_1\left(1 - e^{-\frac{E_1}{\eta_1}t}\right) + \eta_2\left(1 - e^{-\frac{E_2}{\eta_2}t}\right)\right]$$

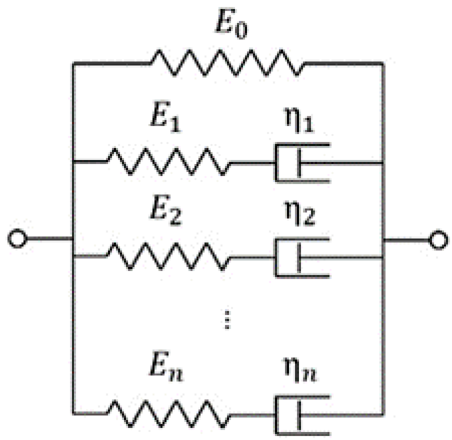
Tirella A et al., *JBMR A* 102 (2014)

Nano-epsilon dot method (nano-$\dot{\varepsilon}M$)

- **Estimate material viscoelastic constant** through **nano-indentation** at **different constant strain rates** ($\dot{\varepsilon}_{ind}$), then **analyse $\boldsymbol{\sigma_{ind}(t)}$ curves** within the **LVR**







1. Nano-indentation at constant $\dot{\varepsilon}_{ind}$

- ✓ **Force-displacement time recording** starts **prior to** sample **contact** → **no pre-load**
- ✓ **Short test** duration → **no sample deterioration**
- ✓ **LVR** determined through **measured σ_{ind}-ε_{ind} curves**

2. Global fitting of **LVR stress-time** (σ_{ind}-t) **series** measured **at different $\dot{\varepsilon}_{ind}$**

3. Lumped parameter estimation

Mattei G et al., *JMBBM* (2015)

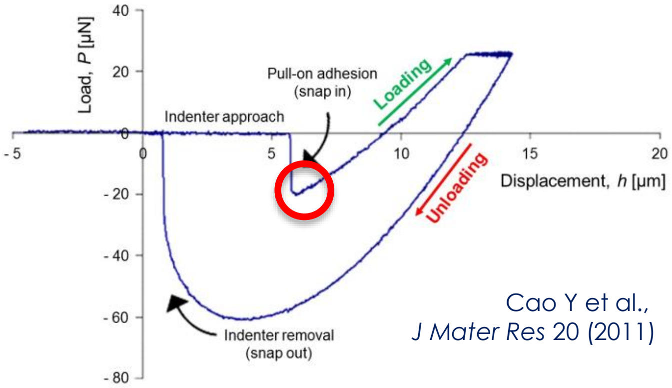
How to get σ-ε curves from P-h measurements?

- **1st issue:** identifying the **initial contact point**

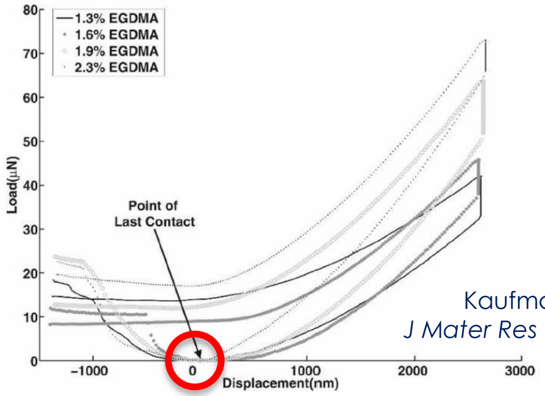
Commercial load-controlled nano-indenters → Load-based contact determination → ⚠

Even **small trigger load** can **cause significant pre-stress** on **soft samples**

Ideal tests should **start out of contact with the sample** ⇒ need of **displacement-controlled experiments** and **post-measurement identification** of the **initial contact point**



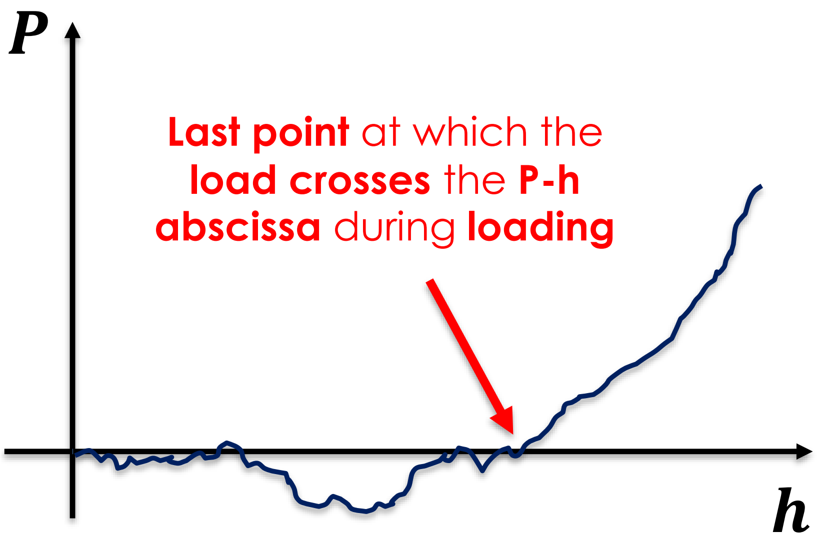
Cao Y et al., *J Mater Res* 20 (2011)



Kaufman et al., *J Mater Res* 23 (2008)

How to get σ-ε curves from P-h measurements?

- **1st issue:** our **solution**



Unique identification of the **contact point** both when

- ✓ **Snap into contact** is **poorly evident**
- ✓ **Noise around zero load** is present

Tirella A et al., *JBMR A* 102 (2014)
Mattei G et al., *J Biomech* 47 (2014)
Mattei G et al., *JMBBM* (2015)

How to get σ-ε curves from P-h measurements?

- **2nd issue:** nano-$\dot{\varepsilon}M$ needs $\sigma_{ind}(t)$ response to constant $\dot{\varepsilon}_{ind}$

Most of the studies define $\dot{\boldsymbol{\varepsilon}}_{ind} = \dot{\boldsymbol{h}}/\boldsymbol{h}$ [1,2] and $\boldsymbol{\varepsilon}_{ind} = \sqrt{\boldsymbol{h}/\boldsymbol{R}}$ [3], **but…**

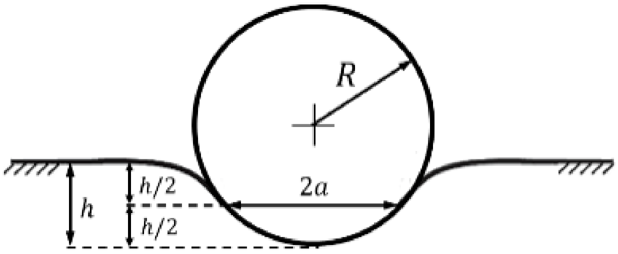
- $\dot{\boldsymbol{\varepsilon}}_{ind} \neq \frac{\partial \boldsymbol{\varepsilon}_{ind}}{\partial \boldsymbol{t}}$
- both $\dot{\boldsymbol{\varepsilon}}_{ind} = \dot{\boldsymbol{h}}/\boldsymbol{h}$ **and** $\frac{\partial \boldsymbol{\varepsilon}_{ind}}{\partial \boldsymbol{t}} = \dot{\boldsymbol{h}}/2\sqrt{\boldsymbol{Rh}}$ are **depth-dependent** functions

NOT suited for the **nano-$\dot{\varepsilon}M$**

[1] Haghshenas M et al., *Mater Sci Eng* 572 (2013)
[2] Maier V et al., *J Mater Res* 26 (2011)
[3] Basu S et al., *J Mater Res* 21 (2006)

How to get σ-ε curves from P-h measurements?

- **2nd issue:** our **solution**



Hertz model $P = \frac{4}{3} E_{eff} R^{1/2} h^{3/2}$

Sneddon relation $h = \frac{a^2}{R}$

$$\frac{h}{a} \cdot \frac{P}{\pi a^2} = \frac{4}{3\pi} E_{eff} \left(\frac{a}{R}\right) \cdot \frac{h}{a}$$

$$\left(\frac{1}{E_{eff}} = \frac{1-\upsilon^2}{E} + \frac{1-\upsilon'^2}{E'} \approx \frac{1-\upsilon^2}{E}\right)$$

$$\sigma_{ind} = \frac{P}{R\sqrt{hR}}$$

$$\varepsilon_{ind} = \frac{4}{3(1-\upsilon^2)}\left(\frac{h}{R}\right)$$

✓ $\sigma_{ind}/\varepsilon_{ind} = E$ (in case of soft materials where $E' \gg E$)

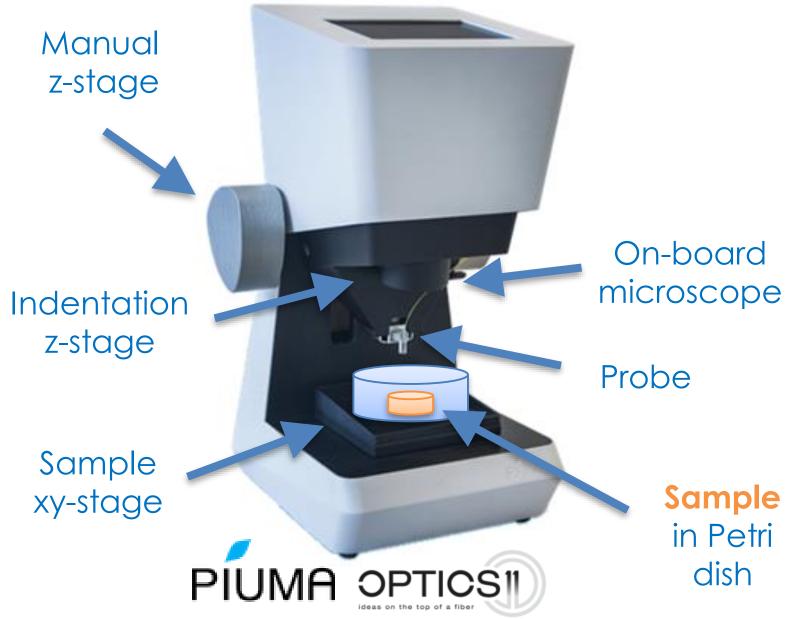
✓ $\dot{\varepsilon}_{ind} = \frac{\partial \varepsilon_{ind}}{\partial t} = \frac{4}{3(1-\upsilon^2)}\left(\frac{\dot{h}}{R}\right)$

A **constant indentation rate** ($\dot{h}$) yields a **constant strain rate** ($\dot{\varepsilon}_{ind}$)

Mattei G et al., *JMBBM* (2015)

An example: PDMS and gelatin characterisation

- **2 different samples: PDMS** (Sylgard 184, 10:1 base to catalyst), **5% w/v gelatin** (type A)
- **Constant $\dot{\varepsilon}_{ind}$ tests** in **dH_2O** at **RT** using the **PIUMA Nanoindenter** (Optics11)
- **Measurements started above** the **sample surface** to **avoid pre-stress** (different tests on different surface points spaced by 200 μm)



Indentation rate $\dot{h}$ to **obtain** a given **indentation strain rate $\dot{\varepsilon}_{ind}$**

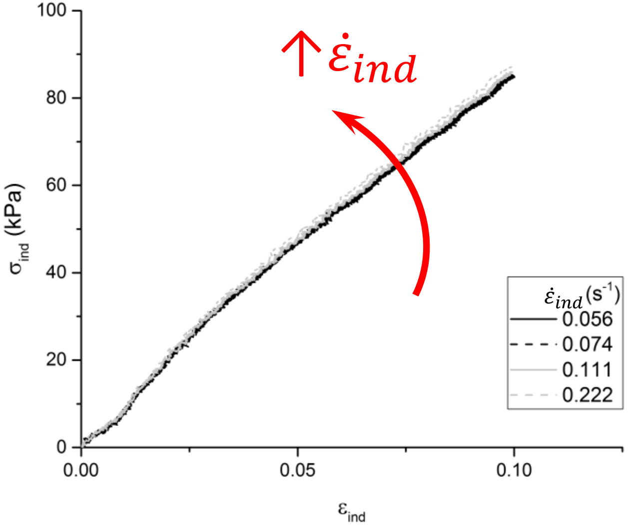
$$\dot{h} = \frac{4}{3(1-\upsilon^2)}\left(\frac{\dot{\varepsilon}_{ind}}{R}\right)$$

Indentation stress-strain curves at different $\dot{\varepsilon}_{ind}$

- Tests at **4 different indentation strain rates** (n = 10 per each $\dot{\varepsilon}_{ind}$)

PDMS

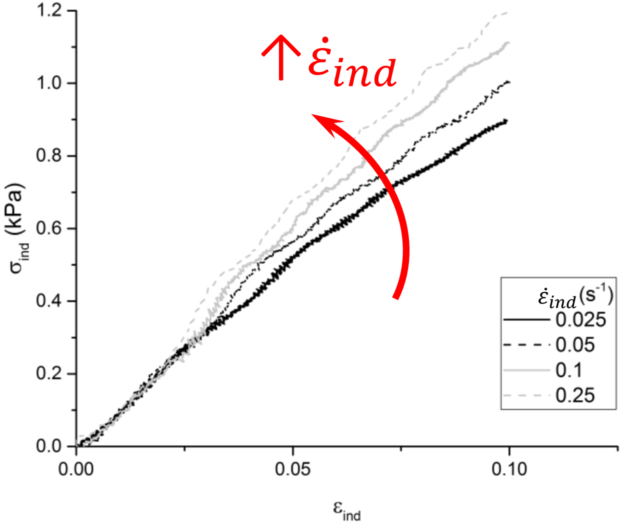
(R = 64 μm, k = 13.4 N/m)



- ✓ **LVR** up to **10%** $\boldsymbol{\varepsilon_{ind}}$
- ✓ Fairly **rate-independent** behaviour

Gelatin

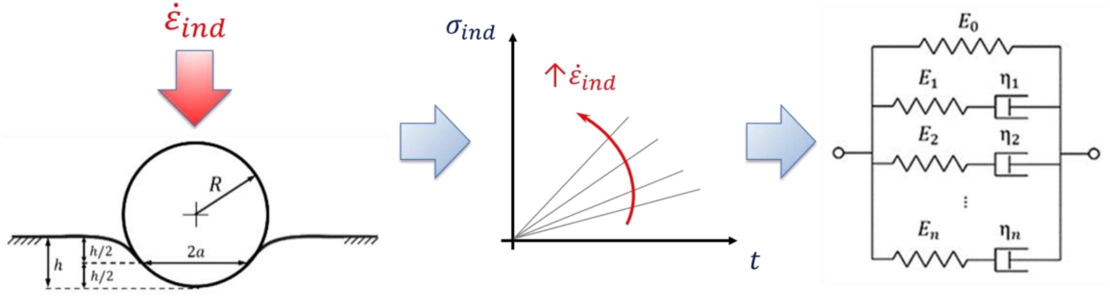
(R = 70 μm, k = 0.55 N/m)

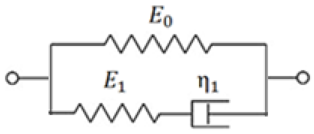


- ✓ **LVR** up to **10%** $\boldsymbol{\varepsilon_{ind}}$
- ✓ Pronounced **rate-dependent** behaviour

Viscoelastic lumped parameter estimation

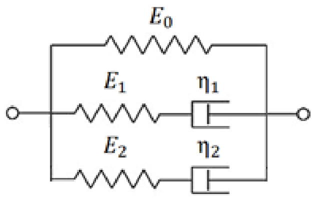
- **Indentation stress-time data within LVR** obtained at **different $\dot{\varepsilon}_{ind}$** were analysed with **$\dot{\varepsilon}M$ global fitting** procedure **sharing the viscoelastic parameters** to estimate





Maxwell Standard Linear Solid (**SLS**)

$$\sigma_{ind}(t) = \dot{\varepsilon}_{ind}\left[E_0 t + \eta_1\left(1 - e^{-\frac{E_1}{\eta_1}t}\right)\right]$$



2-arm generalised Maxwell (**GM2**)

$$\sigma_{ind}(t) = \dot{\varepsilon}_{ind}\left[E_0 t + \eta_1\left(1 - e^{-\frac{E_1}{\eta_1}t}\right) + \eta_2\left(1 - e^{-\frac{E_2}{\eta_2}t}\right)\right]$$

Global fitting results

Viscoelastic parameters estimated using the nano-$\dot{\varepsilon}M$ (est. value ± standard error)

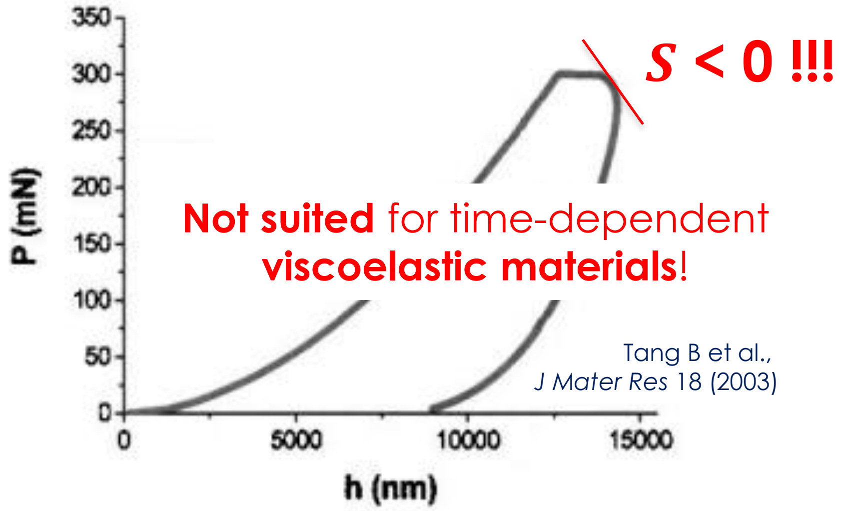
	PDMS		Gelatin	
Parameter	*Maxwell SLS*	*GM2*	*Maxwell SLS*	*GM2*
E_{inst} **(kPa)**	$1.74 \cdot 10^3 \pm 1.47 \cdot 10^1$	$1.74 \cdot 10^3 \pm 1.02 \cdot 10^2$	14.08 ± 0.58	$14.08 \pm 1.37 \cdot 10^3$
E_{eq} **(kPa)**	$8.82 \cdot 10^2 \pm 8.72 \cdot 10^{-1}$	$5.98 \cdot 10^2 \pm 7.14 \cdot 10^1$	1.84 ± 0.42	$4.07 \cdot 10^{-4} \pm 4.65 \cdot 10^2$
τ_1 **(s)**	$0.26 \pm 4.93 \cdot 10^{-3}$	$0.26 \pm 4.93 \cdot 10^{-3}$	6.90 ± 0.60	$14.78 \pm 2.36 \cdot 10^3$
τ_2 **(s)**	-	$1.04 \cdot 10^{12} \pm 3.68 \cdot 10^{11}$	-	$5.57 \pm 1.36 \cdot 10^3$
R^2	0.97	0.97	0.99	0.99

Values in **red cannot be considered as significant** since they are **almost meaningless** with **very large standard errors**, clearly indicating **GM2 model over-parameterisation**

Advantages of the nano-$\dot{\varepsilon}M$

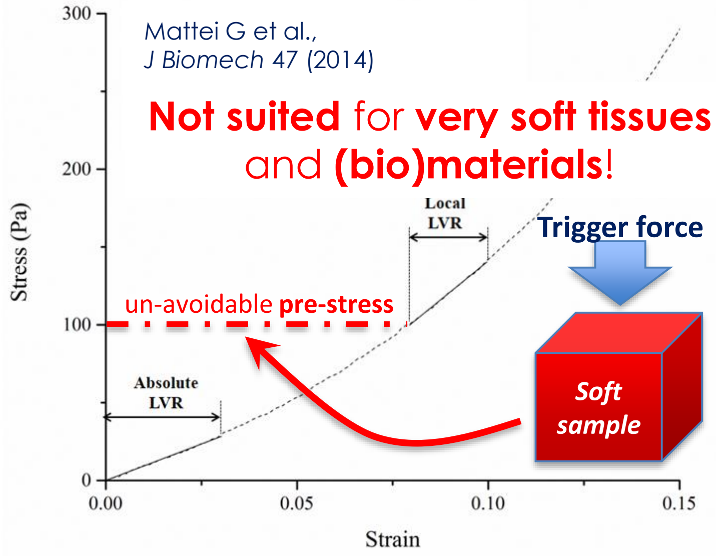
Oliver-Pharr

Oliver W, Pharr G, *J Mater Res* 7, 1564-83 (1992)



- **Elastic-plastic** contact model
- Analysis based on the **unloading curve** using P_{max}, h_{max} and **unloading slope** S
- **Characterises** material **elastic properties only**

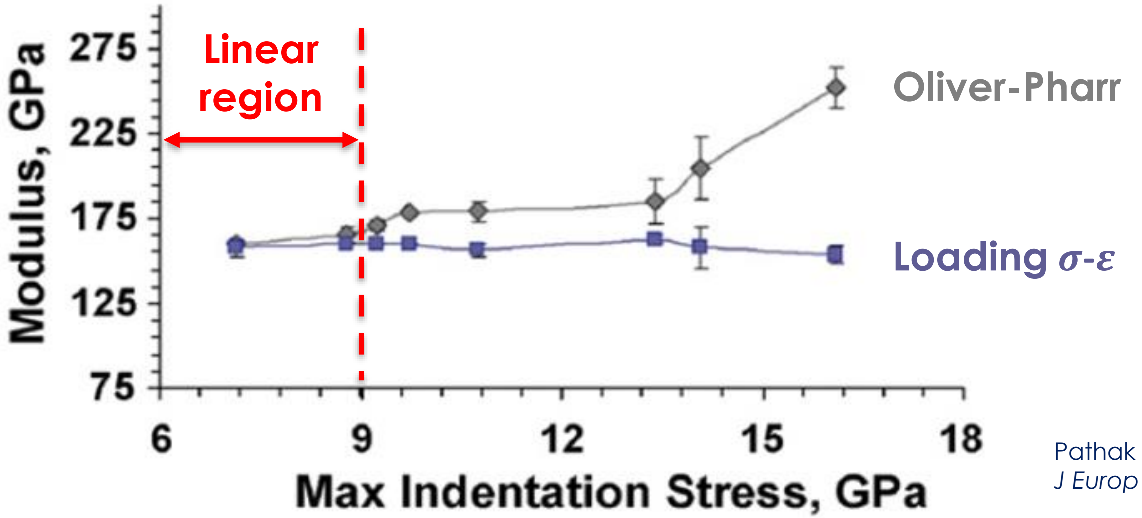
Nano-DMA



- **Micro-scale equivalent** of the **bulk DMA**
- **CSM reduces** the **reliance on unloading curve** and **provide results** as a function of **indentation depth**
- **Need a small** but measureable **trigger force**

Advantages of the nano-$\dot{\varepsilon}M$

- **Mechanical properties** representative of the **"virgin" material**
- **Constant** derived **(visco)elastic parameters, regardless of** the **max load** (or **displacement**) chosen for the experiment



Pathak et al.,
J Europ Cer Soc 28 (2008)

- **During unloading** only the **elastic displacements** are **recovered**

Micro- vs. macro-scale results?

SLS viscoelastic parameters obtained at the micro- and macro-scale* (est. value ± standard error)

** macro-scale values are taken from Tirella A et al., JBMR A 102 (2014)*

	PDMS			Gelatin		
Parameter	*Micro-scale* (*nano-$\dot{\varepsilon}M$*)		*Macro-scale* (*$\dot{\varepsilon}M$*) [1]	*Micro-scale* (*nano-$\dot{\varepsilon}M$*)		*Macro-scale* (*$\dot{\varepsilon}M$*) [1]
E_{inst} **(kPa)**	$(1.74 \pm 0.01) \cdot 10^3$	<	$(2.55 \pm 0.04) \cdot 10^3$	14.08 ± 0.58	>	[illegible]23 ± 0.45
E_{eq} **(kPa)**	$(8.82 \pm 0.01) \cdot 10^2$	<	$(2.14 \pm 0.01) \cdot 10^3$	1.84 ± 0.42	[illegible]	[illegible] ± 0.10
τ_1 **(s)**	0.26 ± 0.01	<	0.66 ± 0.25	6.90 ± 0.60	>	[illegible] 0.19

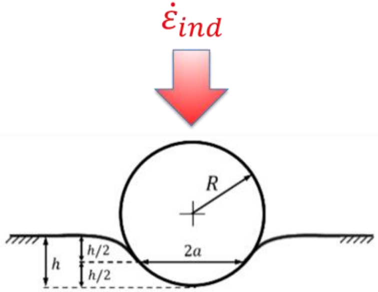
- $\boldsymbol{\tau_1}$ **decrease** from macro- to **micro-scale** observed for **PDMS** is **consistent with literature** [1,2,3]
- **Variations** between results at the **micro-** and **macro-scale** may be due to
  - **real differences** between the **bulk** and **surface** mechanical properties [1]
  - nano-indentation $\boldsymbol{\sigma_{ind}}$ and $\boldsymbol{\varepsilon_{ind}}$ are **not the same** as engineering $\boldsymbol{\sigma}$ and $\boldsymbol{\varepsilon}$

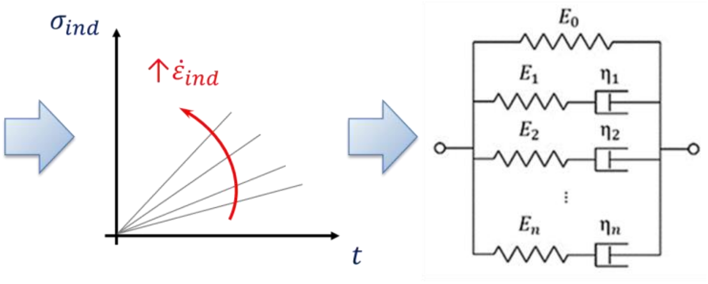
[1] Kaufman J et al., *J Mater Res* 23 (2008) [2] Sasaki S et al., *J Chem Phys* 120 (2004) [3] Mak A et al., *J Biomech* (1987)

Conclusions

- The **nano-$\dot{\varepsilon}M$**
  - combines the **advantages** of the **$\dot{\varepsilon}M$** and **nano-indentation** techniques
  - allows to **locally map** the **viscoelastic properties** of **“virgin” materials** in **absence** of **pre-stress,** being **advantageous over methods** based on the **unloading curve** or **requiring** a **force trigger**
  - very **suited** for **soft biological tissues** and **biomaterials**
  - can be **implemented** with **any displacement-controlled nano-indenter** (e.g. Optics 11 PIUMA)

PIUMA OPTICS11



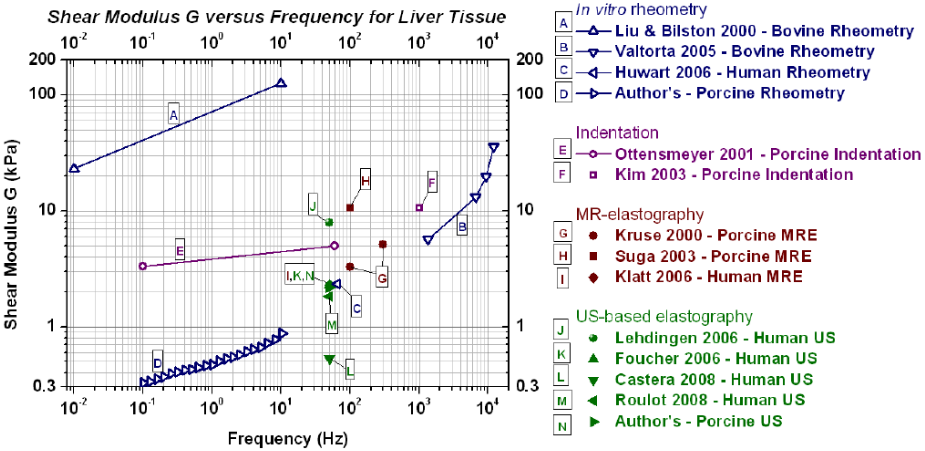


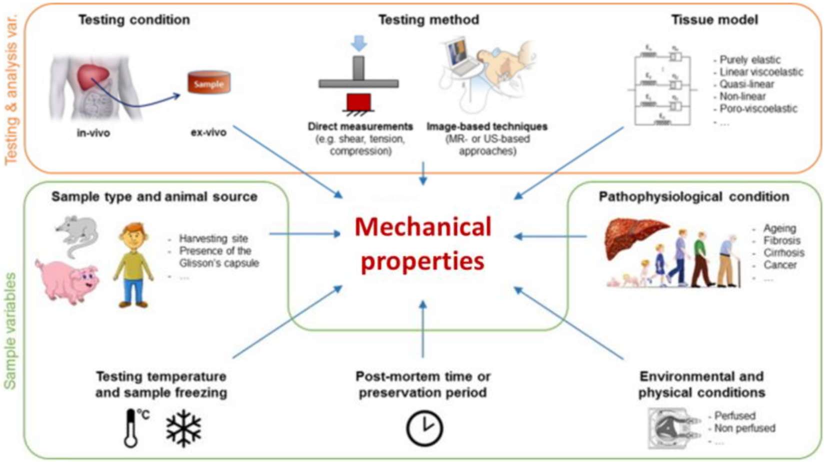
Mattei G et al., *JMBBM* (2015)

Does measuring in the frequency or strain-rate domain affect mechanical results?

Material mechanical *properties*

- **Little consensus** in the literature

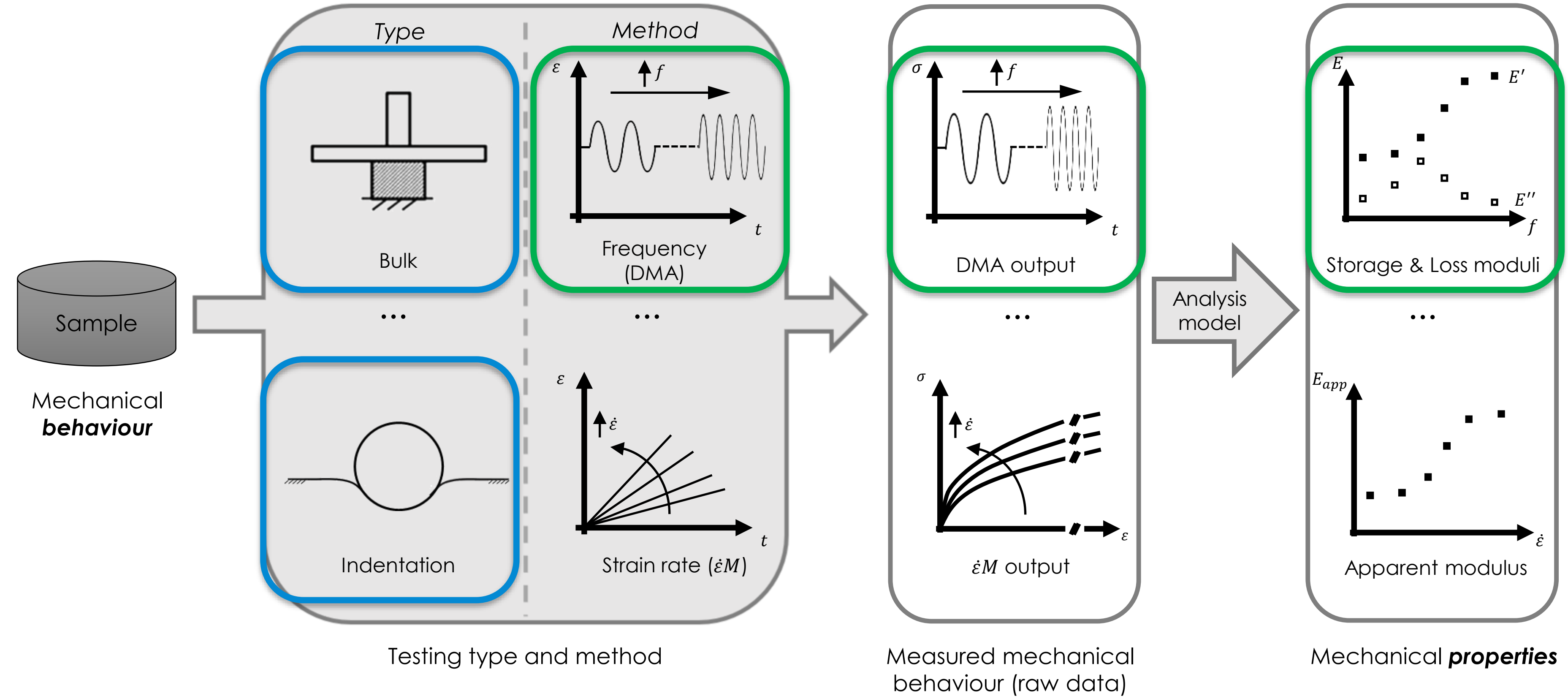




S Marchesseau et al, *Progr in Biophys and Mol Biol* 103:185-96 (2010)

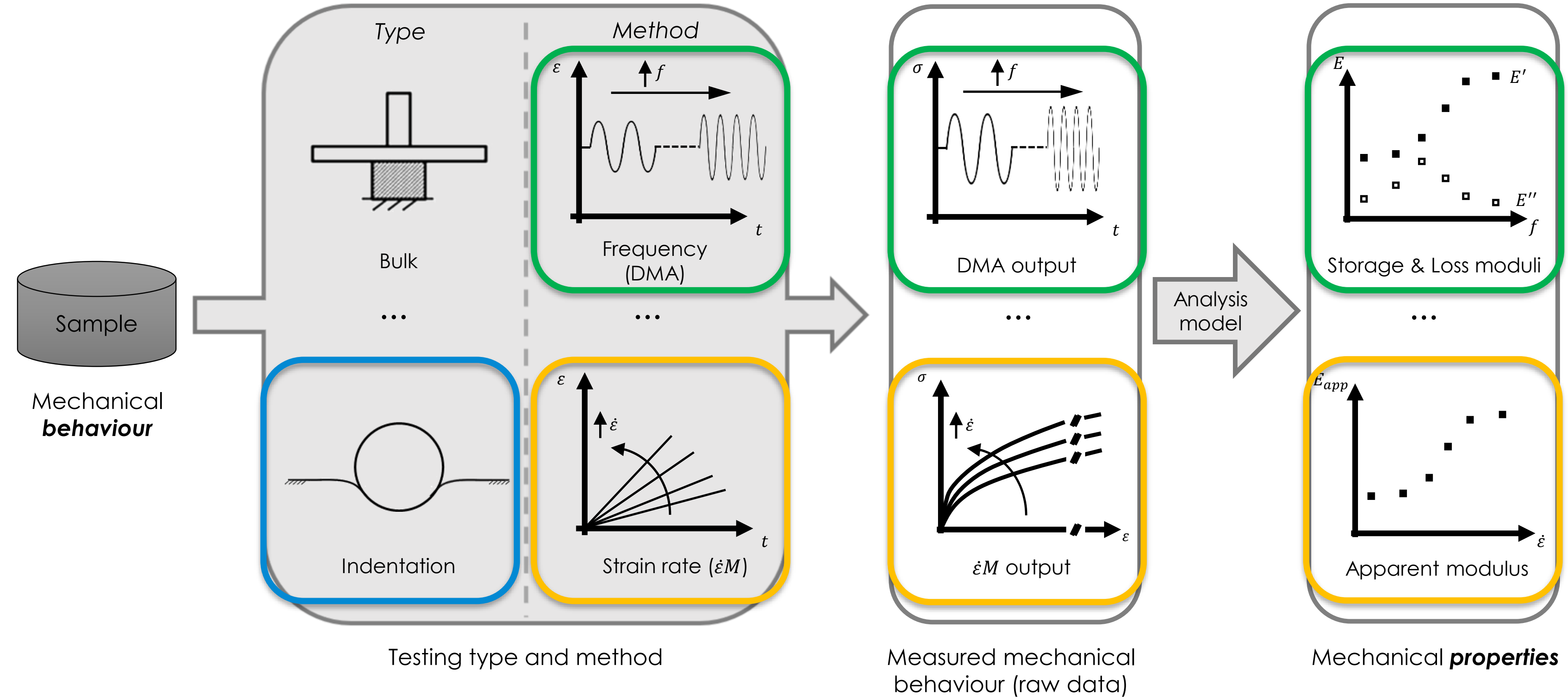
G Mattei and A Ahluwalia, *Acta Biom* 45:60-71 (2016)

From sample mechanical *behaviour* to *properties*



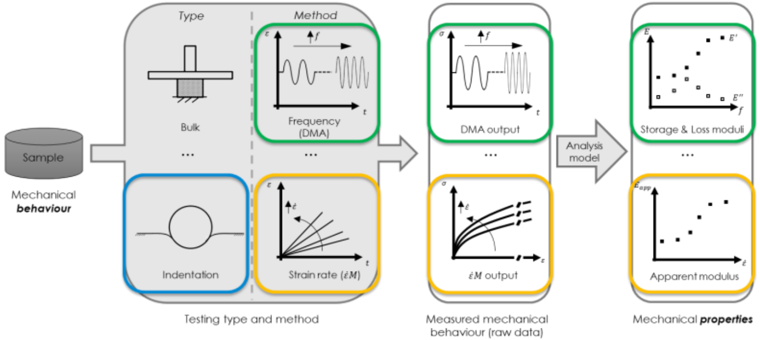
L Bartolini et al, *Sci Rep* 8:13697 (2018)

From sample mechanical *behaviour* to *properties*

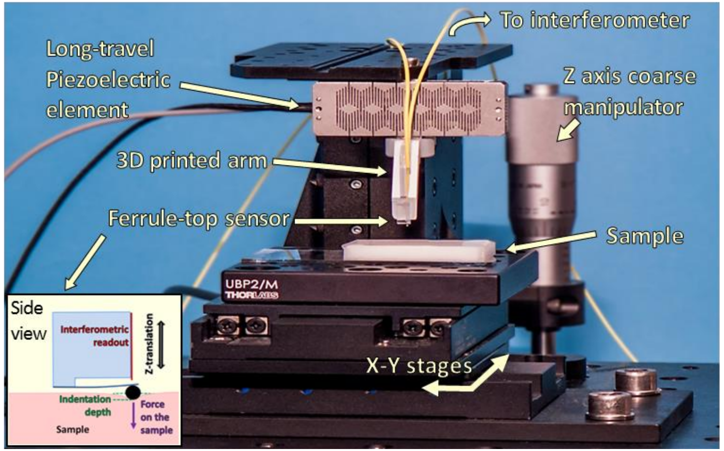


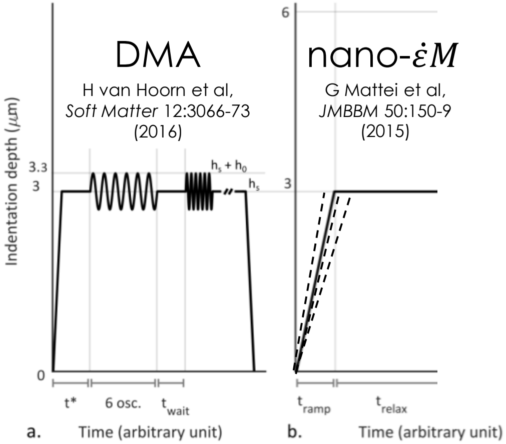
L Bartolini et al, *Sci Rep* 8:13697 (2018)

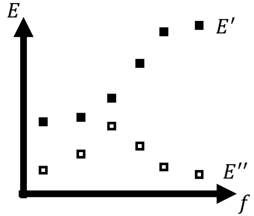
Aim and Strategy



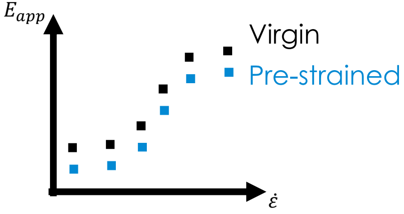
Does measuring in the **frequency** or **strain-rate domain** affect **micro-mechanical properties**?







Storage & Loss moduli



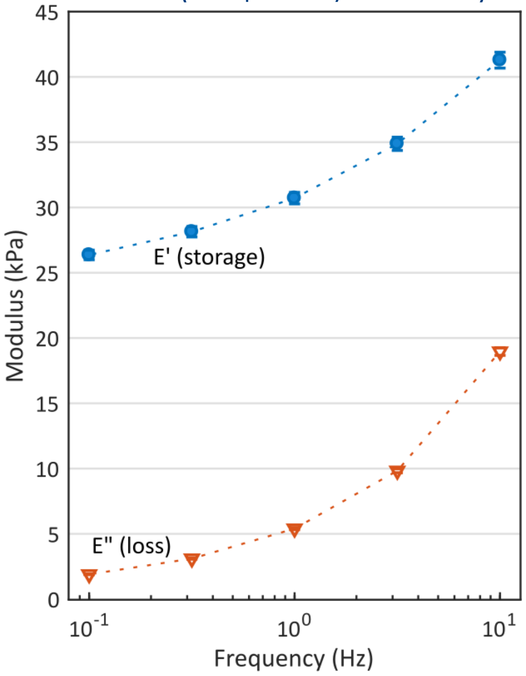
Apparent modulus

L Bartolini et al, *Sci Rep* 8:13697 (2018)

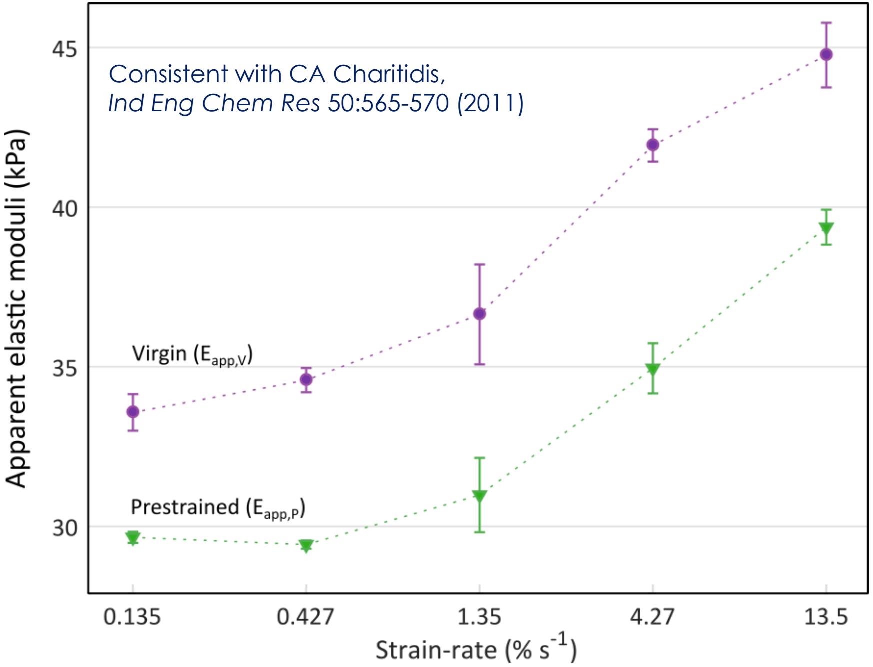
PDMS mechanical properties

- **PDMS** (Sylgard 184, 50:1) samples tested at **25°C** with a **248 μm** radius probe

DMA (frequency domain)



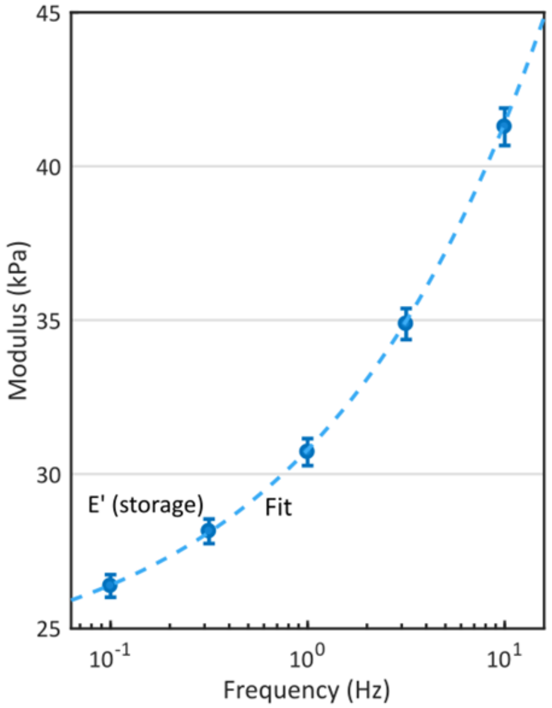
nano-$\dot{\varepsilon}M$ (strain-rate domain)



L Bartolini et al, *Sci Rep* 8:13697 (2018)

Comparing DMA and nano-$\dot{\varepsilon}M$ results

- **Conversion from f to $\dot{\varepsilon}$ domain** based on SE Zeltman et al, *Polymer* 101:1-6 (2016), assuming **1 f_c** or τ



1) **Storage modulus master curve** from DMA data

$$E'(\omega) = a \cdot \tanh\big(b \cdot (\ln(\omega) + c)\big) + d \qquad \omega = 2\pi f$$

2) $E'(\omega)$ converted into **time-domain relaxation modulus**

$$E(t) = \frac{2}{\pi}\int_0^{\infty} \frac{E'(\omega)}{\omega} sin(\omega t)\, \partial\omega$$

3) **Stress-time response** to a given strain history

$$\sigma(t) = E * \partial\varepsilon = \int_{-\infty}^{t} E(t-\tau)\frac{\partial\varepsilon(\tau)}{\partial\tau}\partial\tau \xrightarrow{\text{constant } \dot{\varepsilon}} \sigma(t) = \dot{\varepsilon}\int_0^t E(\tau)\,\partial\tau$$

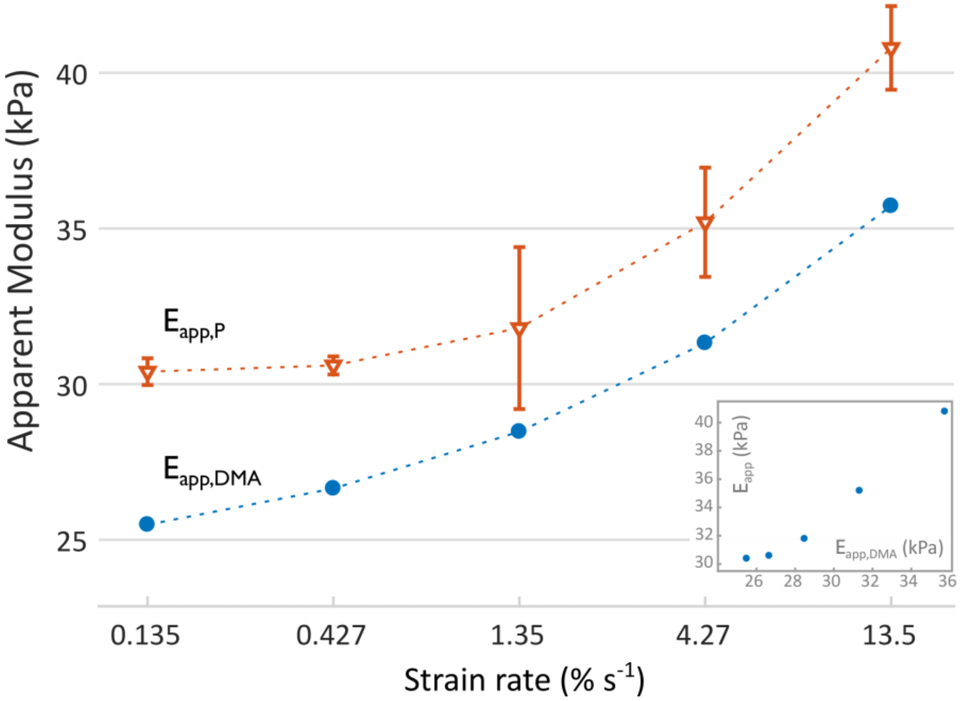
4) **Stress-strain response** by linear transformation of $\varepsilon = \dot{\varepsilon} \cdot t$

$E_{app,DMA}(\dot{\varepsilon})$ as **stress-strain slope** within LVR

L Bartolini et al, *Sci Rep* 8:13697 (2018)

Results: $E_{app,DMA}$ vs $E_{app,P}$

- DMA-derived apparent elastic moduli ($E_{app,DMA}$) *versus* nano-$\dot{\varepsilon}M$ pre-strained ones ($E_{app,P}$)

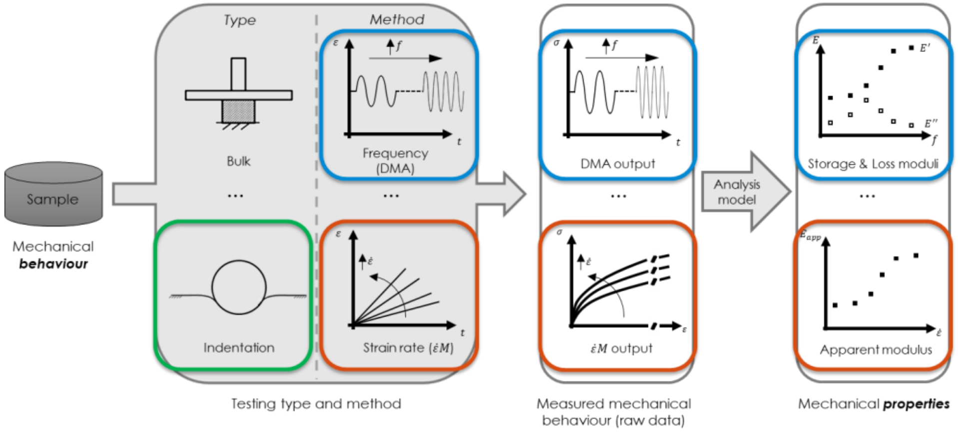


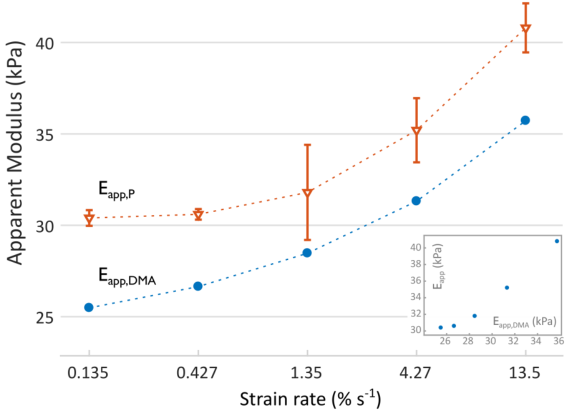
- **Increase** with **strain rate** (as expected)
- $E_{app,P} > E_{app,DMA}$, regardless of $\dot{\varepsilon}$
- **Optimal correlation** (r = 0.99) with almost **constant difference** between moduli (~10%) (consistent with SE Zeltman et al, *Polymer* 101:1-6, 2016)
  - **Good agreement** between f and $\dot{\varepsilon}$ results
  - **Systematic error** possibly due to **narrow f range attainable** by our setup (0.1-10 Hz)

2% increase in ***d*** resulted in:
- small master-curve transition upward
- almost **perfect match** of the **moduli**

L Bartolini et al, *Sci Rep* 8:13697 (2018)

- **Frequency** and **strain-rate** domain results **directly compared**, **without** any **other source of variability**





The observed **compatibility allows** to **combine** these **methods** towards a **more comprehensive understanding** of **material viscoelastic** (time-dependent) **behaviour**, critical for several applications (biomechanics, tissue engineering, mechanobiology, …)

L Bartolini et al, *Sci Rep* 8:13697 (2018)

Useful references

- G. Mattei, G. Gruca, N. Rijnveld, A. Ahluwalia, The nano-epsilon dot method for strain rate viscoelastic characterisation of soft biomaterials by spherical nano-indentation, J. Mech. Behav. Biomed. Mater. 50 (2015) 150–159. doi:10.1016/j.jmbbm.2015.06.015.
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Contacts

Giorgio MATTEI
Assistant Professor in Bioengineering

University of Pisa
Dept. of Information Engineering
Largo Lucio Lazzarino 1
56122 Pisa – Italy

Phone: +39 050 2218239
Mail: giorgio.mattei@unipi.it