

ISTANBUL TECHNICAL UNIVERSITY ★ GRADUATE SCHOOL

**4D PRINTING OF BODY TEMPERATURE RESPONSIVE HYDROGELS
WITH SELF-HEALING AND SHAPE-MEMORY ABILITIES**



M.Sc. THESIS

Gamze AYDIN

Department of Polymer Science and Technology

Polymer Science and Technology Programme

JUNE 2024

ISTANBUL TECHNICAL UNIVERSITY ★ GRADUATE SCHOOL

**4D PRINTING OF BODY TEMPERATURE RESPONSIVE HYDROGELS
WITH SELF-HEALING AND SHAPE-MEMORY ABILITIES**

M.Sc. THESIS

**Gamze AYDIN
(515211007)**

Department of Polymer Science and Technology

Polymer Science and Technology Programme

**Thesis Advisor: Prof. Dr. Oğuz OKAY
Thesis Co-Advisor: Dr. Turdimuhammad ABDULLAH**

JUNE 2024

İSTANBUL TEKNİK ÜNİVERSİTESİ ★ LİSANSÜSTÜ EĞİTİM ENSTİTÜSÜ

**KENDİ KENDİNİ ONARMA VE ŞEKİL HAFIZA ÖZELLİKLERİNE SAHİP
VÜCUT SICAKLIĞINA DUYARLI HİDROJELLERİN DÖRT BOYUTLU
BASKISI**

YÜKSEK LİSANS TEZİ

**Gamze AYDIN
(515211007)**

Polimer Bilim ve Teknolojisi Anabilim Dalı

Polimer Bilim ve Teknolojisi Programı

**Tez Danışmanı: Prof. Dr. Oğuz OKAY
Tez Eş-Danışmanı: Dr. Turdimuhammad ABDULLAH**

HAZİRAN 2024

Gamze AYDIN, a M.Sc. student of ITU Graduate School student ID 515211007, successfully defended the thesis/dissertation entitled “4D PRINTING OF BODY TEMPERATURE RESPONSIVE HYDROGELS WITH SELF-HEALING AND SHAPE-MEMORY ABILITIES”, which she prepared after fulfilling the requirements specified in the associated legislations, before the jury whose signatures are below.

Thesis Advisor : **Prof. Dr. Oğuz OKAY**

Istanbul Technical University

Thesis Co-Advisor : **Dr. Turdimuhammad ABDULLAH**

Istanbul Technical University

Jury Members : **Prof. Dr. Oğuz OKAY**

Istanbul Technical University

Dr. Turdimuhammad ABDULLAH

Istanbul Technical University

Prof. Dr. Deniz CEYLAN

Bezmiâlem Vakıf University

Assoc. Prof. Dr. Erdem YAVUZ

Istanbul Technical University

Dr. Esra SU

Istanbul University

Date of Submission : 23 May 2024

Date of Defense : 25 June 2024





With love and gratitude to my family,



FOREWORD

I have become aware that I am not the same person I was at the beginning of my master's studies thanks to Polymeric Gels Research Lab (PGRL). Joining the PGRL was not just a researching experience; it was a masterclass. I am forever grateful to Prof. Dr. Oğuz OKAY for accepting me as his student. His guidance and unwavering encouragement fostered a courage I never knew I had. Witnessing his working discipline will forever shape my scientific journey and I will always be proud to be a student of him.

Prof. Dr. Oğuz OKAY's mentorship was not singular. On day one, he introduced me to the invaluable guidance of Dr. Turdimuhammad ABDULLAH. Dr. Abdullah's constant support, especially during the initial phases of my studies and his encouraging leadership on my research were also instrumental in my success.

The warmth of our office would not have been the same without Çiğdem Buse ORAL, a former member of PGRL. Her hard-working lab routines and infectious friendly spirit were a constant source of encouragement throughout my studies. I am also grateful for the former members of the PGRL: Dr. Esra SU, Büşra SEKİZKARDEŞ, Sümeyye ÖZER, and many others. Their research experiences and warm conversations were a valuable source of support.

My deepest gratitude goes to my family, especially my sister Tuğba, whose cheerful spirit has always been a source of light. I also owe a debt of endless thanks to my husband İlker. His unwavering empathy and support were essential on this journey.

Finally, I would also like to thank to The Scientific and Technological Research Council of Turkey (TUBITAK) for their financial and material support (Project No. 121Z568).

June 2024

Gamze AYDIN
(Chemist)

TABLE OF CONTENTS

	<u>Page</u>
FOREWORD	ix
TABLE OF CONTENTS	xi
ABBREVIATIONS	xiii
SYMBOLS	xv
LIST OF TABLES	xvii
LIST OF FIGURES	xix
SUMMARY	xxiii
ÖZET	xxvii
1. INTRODUCTION	1
2. THEORETICAL PART	5
2.1 Hydrogels	6
2.2 3D Printing	14
2.3 4D Printing	21
3. EXPERIMENTAL PART	25
3.1 Chemicals	25
3.1.1 Monomers	25
3.1.2 Photoinitiator system	26
3.2 Tools and Instruments	26
3.2.1 Resin printer	26
3.2.2 Analytical balance	27
3.2.3 Digital compass	27
3.2.4 Vacuum oven	27
3.2.5 Differential scanning calorimetry	27
3.2.6 Rheometer	28
3.2.7 Light microscope	28
3.2.8 Mechanical test device	28
3.2.9 Fourier-transform infrared spectroscopy (FTIR)	28
3.2.10 X-ray powder diffraction (XRD)	28
3.3 Experimental Procedure	28
3.4 Hydrogel Characterization	31
3.4.1 Swelling measurements	31
3.4.2 (ATR)-FTIR measurements	32
3.4.3 Differential scanning calorimetry (DSC) measurements	32
3.4.4 X-ray powder diffraction (XRD) measurements	32
3.4.5 Rheological measurements	33
3.4.6 Mechanical tests	33
3.4.7 Determination of shape-memory behavior	34
4. RESULTS AND DISCUSSION	35
4.1 x_{DMAA} Effect	36
4.2 C16A Effect	39

5. CONCLUSIONS	53
REFERENCES.....	55
APPENDICES	65
CURRICULUM VITAE.....	67



ABBREVIATIONS

3D	: Three-Dimensional
4D	: Four-Dimensional
ASTM	: The American Society for Testing and Materials
C16A	: Hexadecyl Acrylate
CAD	: Computer-Aided Design
DMAA	: <i>N, N</i> -Dimethyl Acrylamide
DSC	: Differential Scanning Calorimetry
ECM	: Extracellular Matrix
FDM	: Fused Deposition Modeling
FTIR	: Fourier-Transform Infrared Spectroscopy
ISO	: The International Organization for Standardization
LA	: Lauryl Acrylate
LCST	: Lower Critical Solution Temperature
MAA	: Methacrylic Acid
mm	: Milimeter
μm	: Micrometer
NIPAAm	: <i>N</i> -isopropylacrylamide
PBF	: Powdered Bed Fusion
PNIPAAm	: Poly(<i>N</i> -isopropylacrylamide)
SA	: Stearyl Acrylate
SLA	: Stereolithography Apparatus
SLS	: Selective Laser Sintering
TPO	: Diphenyl (2,4,6-trimethylbenzoyl) Phosphine Oxide
UCST	: Upper Critical Solution Temperature
UV	: Ultraviolet
XRD	: X-Ray Powder Diffraction



SYMBOLS

C16A mol%	: C16A Mole Percentage in Overall Precursor Mixture
ΔG	: Change in Gibbs Free Energy
$\Delta H_{m,C16A}$	: Melting Enthalpy of Hexadecyl Acrylate
ΔH_{C16A}^o	: Melting Enthalpy of 100% Crystalline Hexadecyl Acrylate
E	: Young's Modulus
ϵ	: Strain
$\dot{\epsilon}$	: Strain Rate
ϵ_f	: Fracture Strain
f_{cry}	: Fraction of Crystallinity
γ_o	: Strain Amplitude
G'	: Storage Modulus
λ	: Stretch Ratio
m_0	: As-prepared Sample Mass
m_{rel}	: Relative Weight Swelling Ratio
m_{swl}	: Mass of Swollen Sample
ω	: Angular Frequency
q_w	: Weight Swelling Ratio
$R_f(\theta)$	: Angular Shape Fixity Ratio
$R_r(\theta)$	: Angular Shape Recovery Ratio
σ_f	: Fracture Stress
σ_{nom}	: Nominal Stress
σ_{true}	: True Stress
σ_y	: Yield Stress
$\tan \delta$	: Loss Factor
T_{cry}	: Crystallization Temperature
T_g	: Glass Transition Temperature
T_{high}	: Temperature Exceeding the Critical Temperature
T_{low}	: Temperature Below the Critical Temperature
T_m	: Melting Temperature

V_a	: Activation Volume
W	: Toughness
W_g	: Gel Fraction
x_{DMAA}	: Mole Fraction of DMAA in DMAA-MAA Mixture



LIST OF TABLES

	<u>Page</u>
Table 3.1 : Synthesis conditions of hydrogels at various DMAA to MAA mole ratios. C16A content was fixed at 40 mol %.....	29
Table 3.2: Synthesis conditions of hydrogels at various C16A contents. The mole ratio of DMAA to MAA was fixed at 0.5.	30
Table A.1 : Preparation conditions of hydrogels at various mole ratios of DMAA to MAA. C16A= 40 mol %.	66
Table A.2 : Preparation conditions of hydrogels at various C16A contents. $x_{\text{DMAA}}=0.5$	66



LIST OF FIGURES

	<u>Page</u>
Figure 2.1 : From left to right, skeletal structures of linear, branched, and cross-linked polymer structures.	5
Figure 2.3 : Illustration of swelling and deswelling in hydrogels.....	7
Figure 2.4 : Cartoon for comparison of (a) chemically crosslinked and (b) physically crosslinked hydrogel structures under stress exposure (Abdurrahmanoglu et al., 2009).....	10
Figure 2.5 : Cell attachment-detachment control via thermo-responsive PNIPAAm compared to enzymatic treatment (Kumashiro et al., 2010).	12
Figure 2.6 : Shape-memory behavior: (a) Shape fixing of the dog-bone sample, (b) angular recovery of a curved sample (Lendlein & Gould, 2019).....	13
Figure 2.7 : (a) 3D model of the heart, (b) Sliced output of the printer program, (c) a layer/slice of the 3D model, (d) printed heart in the gelatin support medium, (e) robust cardiac patches keeping their integrity, (e) perfusion in the patch (Noor et al., 2019).....	16
Figure 2.8 : Side-by-side comparison of subtractive manufacturing and additive manufacturing (Chen et al., 2020).....	17
Figure 2.9 : The method of operation in FDM (Dvořák et al., 2018).	19
Figure 2.10 : The method of operation in PBF (Kumar et al., 2023).....	20
Figure 2.11 : Printing mechanism of an SLA printer (Lakkala et al., 2023).	21
Figure 2.12 : Contrasting 3D and 4D printing workflows (Momeni et al., 2017). ...	22
Figure 3.1 : The chemical structures of monomers: (a) hexadecyl acrylate (C16A), (b) <i>N</i> , <i>N</i> -dimethyl acrylamide (DMAA), (c) methacrylic acid (MAA)...	25
Figure 3.2 : The chemical structures of photoinitiator Diphenyl (2,4,6-trimethylbenzoyl) phosphine oxide (TPO).....	26
Figure 3.3 : Validation of printing parameters through printed dummy models. From left to right: printed dummy models with exposure times of a 7, 8, 9, and 10 seconds.	27
Figure 4.1 : (a) IR spectra of poly(DMAA-co-MAA), poly(DMAA), and poly(MAA), (b) IR spectra of poly(C16A-co-DMAA-co-MAA).....	35
Figure 4.2 : Effect of x_{DMAA} on hydrogel properties. (a) Changes in relative weight swelling ratio m_{rel} and swelling ratio q_w with increasing x_{DMAA} . (b) Gel fraction W_g of the hydrogels at different x_{DMAA} values.	37
Figure 4.3 : (a) Stress-strain curves of hydrogels with various x_{DMAA} , (b) Young's modulus E , fracture stress σ_f , (c) toughness W and fracture strain ε_f of hydrogels with various x_{DMAA}	38
Figure 4.4 : Relative weight swelling ratio m_{rel} ($\square$), weight swelling ratio q_w ($\circ$), and gel fraction W_g ($\triangle$) of hydrogels with varying C16A mol%.	39

- Figure 4.5** : Thermal properties of hydrogels with varying C16A mol%. (a) Heating scans, (b) cooling scans, and (c) changes in melting temperature T_m , crystallization temperature T_{cry} , and fraction of crystallinity f_{cry} of the hydrogels. 41
- Figure 4.6** : XRD patterns of hydrogels with various C16A mol %= (a) 0, (b) 10, (c) 20, (d) 30, (e) 40, (f) 50, (g) 60, (h) 70, (i) 100..... 42
- Figure 4.7** : The combined XRD patterns of hydrogels with various C16A content. 42
- Figure 4.8** : The change in G' (a) and $\tan \delta$ (b) of hydrogels with different C16A contents during a heating-cooling cycle in the range of 25 – 65 °C. $\omega = 6.28 \text{ rad.s}^{-1}$. $\gamma_0 = 0.1\%$ 43
- Figure 4.9** : (a) Stress-strain curves of hydrogels at a fixed $\dot{\epsilon}$ of $1.89 \times 10^{-2} \text{ s}^{-1}$, the changes in $E - \sigma_f$ (b) and $W - \varepsilon_f$ (c) of hydrogels with various C16A contents. (d) Stress-strain curves of hydrogel with 0.5 x_{DMAA} and 50 mol% C16A at various strain rates $\dot{\epsilon}$ as indicated. (e) Yield stress σ_y of the hydrogels plotted against the strain rate $\dot{\epsilon}$ in which symbols represent the experimental data while the solid line depicts the best-fitting line. ... 46
- Figure 4.10** : Images of a bending test: shape fixing and shape recovery of a hydrogel sample with 50 mol% C16A. 47
- Figure 4.11** : (a) Angular shape-recovery ratios of a rectangular-shaped hydrogel sample (40 mm $\times$ 12 mm $\times$ 1 mm) as a function of temperature. (b) Angular shape-fixity ratio $R_f(\theta)$ of samples with 40, 50, 60, and 70 C16A mol%. 47
- Figure 4.12** : 4D-printed paperclip-shaped hydrogel with 50 mol% C16A and 0.5 x_{DMAA} . From left to right: as prepared state, programmed temporary shape, and recovered permanent shape. 48
- Figure 4.13** : Real-time shape recovery images of petal-shaped hydrogel with 50% C16A and 0.5 x_{DMAA} . (a) 3D-printed flower permanent shape, (b) programmed temporary shape view from the side and (c) up, (d) transition to permanent shape as dispensing hot water from a syringe, (e) recovered permanent shape view from up and (f) side. 49
- Figure 4.14** : Self-healing images of the specimen with 50 mol% C16A; prior to any treatment (a), a scratched specimen surface under 4x magnification (b), after being submerged in water at 60 °C for 10 minutes under 4x (c) and 10x magnification (d). 50
- Figure 4.15** : Self-healing images of the specimen with 60 mol% C16A; prior to any treatment (a), a scratched specimen surface under 4x magnification (b), after being submerged in water at 60 °C for 10 minutes under 4x (c) and 10x magnification (d). 50
- Figure 4.16** : Self-healing images of the specimen with 70 mol% C16A; prior to any treatment (a), a scratched specimen surface under 4x magnification (b), after being submerged in water at 60 °C for 10 minutes under 4x (c) and 10x magnification (d). 51
- Figure 4.17** : Images of a specimen with 70 mol% C16A before (a) and after cutting (b), after coloring one of its parts with a marker for clarity (c), and after healing at 60 °C for 10 minutes (d). The healed specimen (e) can be twisted (g), bent (f), and stretched to a stretch ratio of $283 \pm 10 \%$ 51

Figure 4.18 : (a) The stress-strain curves for the virgin (solid line) and healed (dashed line) specimens with 50, 60, and 70 mol% C16A, (b) efficiencies of Young's modulus E , fracture stress σ_f , and fracture strain ε_f of specimens. 52





4D PRINTING OF BODY TEMPERATURE RESPONSIVE HYDROGELS WITH SELF-HEALING AND SHAPE-MEMORY ABILITIES

SUMMARY

The power of additive manufacturing (AM), also known as three-dimensional (3D) printing, is being continuously expanded by researchers who are pushing the boundaries of this transformative technology. This technique, which is a cornerstone of the fourth industrial revolution allows for the creation of complex, customised objects directly from computer-aided designs. In recent years, a revolutionary advancement within AM has emerged: four-dimensional (4D) printing. 4D printing integrates smart materials with 3D printing, enabling objects to change properties over time. The fourth dimension in the name refers to the time parameter, which arises from the change in the properties of the printed material over time. Leveraging the inherent design freedom and moldless fabrication of 3D printing, this thesis utilized stereolithography, a high-precision photopolymerization technique, for rapid, low-volume, and customized manufacturing applications. In this thesis, a hydrophobic hexadecyl acrylate (C16A) and hydrophilic *N,N*-dimethyl acrylamide (DMAA) and methacrylic acid (MAA) monomers were polymerized in the presence of diphenyl (2,4,6-trimethylbenzoyl) phosphine oxide (TPO) photoinitiator using a commercial stereolithography (SLA) device without any solvent or crosslinker. In this wise, the possible toxic side effects of solvent and chemical crosslinker were avoided which are critical concerns for a potential medical application.

Among the materials employed in 4D printing, smart hydrogels are highly promising. Hydrogels are particularly attractive for biomedical applications due to their biomimetic properties, which enable them to resemble the structure and function of natural tissues. However, traditional hydrogels exhibit shortcomings, including low mechanical strength and slow response times. Consequently, the development of printable hydrogels that are mechanically robust, capable of actuation at body temperature, and capable of maintaining their actuated form represents a key area of ongoing research. In this study, mechanical characteristics of hydrogels were enhanced by the synergic effect of hydrophobic interactions in C16A and hydrogen bonding in between DMAA and MAA monomers. The mechanical behavior of these physically crosslinked hydrogel was governed by the DMAA:MAA mole ratio, denoted as x_{DMAA} . Rapid temperature-induced actuation were also achieved successfully in less than 30 seconds. These actuations were based on the melting of the hydrophobic domains formed by the C16A units. Moreover, the melting temperature could be tailored by adjusting C16A amount in between 34 °C to 40 °C at 10 mol% and 70 mol% C16A, respectively.

Since the hydrogels are composed of hydrophobic and hydrophilic monomers in their formulations, the experiments were designed in two steps. In the first step, the mole ratios of the two hydrophilic monomers were differentiated by keeping the C16A ratio constant at 40% by mole. The mole fraction of DMAA in DMAA-MAA was expressed

as x_{DMAA} . DSC results revealed that the melting temperature T_m , crystallization temperature T_{cry} and crystallinity fraction f_{cry} remained relatively constant in all hydrogel samples (approximately 35 ± 1 °C, 26 ± 2 °C, and 27 ± 18 °C, respectively). This observation can be attributed to the constant content of the hydrophobic C16A monomer, which dominates the thermal behavior. The variation in x_{DMAA} significantly affected the mechanical behavior of the hydrogels. Samples with x_{DMAA} values of 0.5, 0.67, and 0.8 showed ductile behavior and therefore their toughness was high compared to the other samples. Overall, the sample with 0.5 x_{DMAA} showed the highest toughness as 2.23 ± 0.2 MJ.m⁻³, thus the x_{DMAA} value of 0.5 was chosen as the optimal ratio. Reasonably, the molar equivalence of DMAA, the hydrogen bond acceptor, and MAA, the donor, leads to the highest efficiency of hydrogen bonding and accordingly, the highest mechanical toughness is observed in the hydrogel with x_{DMAA} value of 0.5.

In the second part of the experimental study, the mole percentage of C16A monomer was varied between 10 and 70 by taking x_{DMAA} value as 0.5. The data on swelling showed a decrease in water uptake with an increase in C16A content. To gain further insight into this finding, DSC was used to analyze the thermal properties of the hydrogels. This analysis provides evidence for the presence of crystalline domains formed primarily by the hydrophobic segments of the C16A molecules. As expected, the melting temperature of the crystalline domains of the hydrogel increased proportionally with the increase in the C16A mole percentage. It is important to note that this resulted in a wide range of T_m values ranging from 34 °C to 40 °C at 10 mol% and 70 mol% C16A, respectively. The results show that there is a correlation between the amount of C16A and the T_m values of the hydrogels. The f_{cry} value of the hydrogels increased in proportion to the amount of C16A present, from 0.8 ± 0.2 to 39.7 ± 3 for 10% and 70% C16A samples, respectively. This can be attributed to the formation of highly ordered and stable crystal structures due to the presence of longer hydrophobic segments within the C16A units. These well-defined crystalline domains require a higher temperature to melt, resulting in the observed increase in T_m . XRD measurements show that as the mole percentage of C16A increases, the broad peak around 16°, characteristic of amorphous polymers, is replaced by an increasingly sharp 21.3° peak. The peak at 21.3° corresponds to a d-spacing of 0.42 nm. This value is typical for a crystalline structure with paraffin-like alkyl chains packed side by side. In our case, it refers to the alignment of hexadecyl side chains originating from C16A units within the hydrogel network. The d-spacing of 0.42 nm reflects the average distance between two neighboring hexadecyl side chains when aligned along the same polymer backbone. XRD results, indicating increased crystallinity with higher C16A content, are in good agreement with DSC results, which show higher melting temperatures for hydrogels containing more C16A. All tested hydrogels show net reversible changes in storage modulus G' and loss factor $\tan \delta$ above and below the phase transition temperatures. It is important to note that the magnitude of these changes is highly proportional to the amount of C16A present in the hydrogels. At temperatures below the phase transition temperature, the modulus increases as a result of the increasing C16A content due to the increasing degree of crystallinity. Conversely, at temperatures above the phase transition temperature, the modulus decreases with increasing C16A content. Below the transition temperature, a decreasing trend in $\tan \delta$ is observed for all hydrogels. This is due to the fact that the polymer chains are more densely packed and have greater stiffness as a result of the increase in crystallinity. Above the T_m value of the hydrogels, the crystalline domains acting as physical cross-linking points melt reversibly, causing them to become relatively soft. The rheological measurements were found to be consistent with the

DSC measurements. In the mechanical tests, both the fracture stress σ_f and the fracture strain ε_f increase with increasing C16A content, reaching a maximum at 50 mol%. Thereafter, these values decrease slightly at C16A concentrations higher than 50 mol%. As a result, the hydrogel containing 50 mol% C16A exhibits superior mechanical properties with maximum values of 50.8 ± 0.7 MPa Young's modulus (E) and 8.15 ± 0.7 MJ.m⁻³ toughness (W). Thus, the overall findings indicate that the balanced incorporation of C16A at an optimum mole level of 50 mol% is crucial to achieving superior mechanical properties in these hydrogels. The network structure of the hydrogel responds differently to stress at different rates affecting its mechanical properties. Faster deformation makes hydrogels stiffer and stronger. The yield stress σ_y is directly proportional to the logarithm of the strain rate $\dot{\varepsilon}$ according to Eyring's theory, which assumes that polymer segments must overcome an energy barrier at the yield point. To evaluate this effect, mechanical testing was performed by keeping the strain rate $\dot{\varepsilon}$ in the range from 0.95×10^{-2} s⁻¹ to 4.54×10^{-2} s⁻¹. It reveals that the E , σ_y and σ_f of hydrogels increase with increasing $\dot{\varepsilon}$, while ε_f decreases. This analysis also reveals an activation volume V_a of 2.3 ± 0.2 nm³. It shows that the Kuhn length is about 1.3 nm, which is equivalent to the length of the bond between two neighboring carbon atoms. This is an indication that at the strain rates tested, each bond plays an important role in the mechanical response.

The shape memory properties of hydrogels were revealed by utilizing the bending test. The angular shape recovery ratio $R_r(\theta)$ of the hydrogels was calculated by recording the recovery angle of the bent sample for each 1 °C increase in temperature. The angular fixing ratio $R_f(\theta)$ of the hydrogels was calculated as 98%, 98.8%, 99.2%, and 99.4% for samples with C16A mole percentages of 40, 50, 60, and 70, respectively. All samples except the sample containing 40 moles of C16A achieved a shape recovery ratio of 100%. This ability can be attributed to the high crystallinity of the hydrogels. A significant increase in the $R_r(\theta)$ was observed around 34 – 36 °C for tested samples, and increasing the amount of C16A resulted in a faster angular recovery. The self-healing behavior of the specimens was observed in a series of experiments. The specimens were scratched with a razor blade and then immersed in a 60 °C aqueous medium for 10 minutes. The results reveal a clear correlation between the C16A content and the degree of healing. Although the scratch is not completely healed in the sample containing 50 mol% C16A, its width decreases significantly from an average of 82.3 μm to 44.8 μm. Regarding 60 mol% C16A, the scratch is almost healed, showing that the width decreases from an initial 29.9 μm to only 5.7 μm on average. On top of that, the sample with 70 mol% C16A was revealed to be completely healed with the disappearance of the 109.6 μm wide scratch. The self-healing abilities of the hydrogels were evaluated through mechanical tests, considering both their virgin and healed states. It was found that increasing C16A content significantly improved the efficiencies of E , σ_f , ε_f . E efficiencies were 91.8% for 50 mol%, 97.4% for 60 mol%, and 99.5% for 70 mol% C16A. σ_f efficiencies were 83.9% for 50 mol%, 87.8% for 60 mol%, 88.5% for 70 mol% C16A. efficiency initially lagged behind at 50 mol% C16A (11.5%) but showed improvement with higher C16A content, reaching 63.2% for 60 mol% and 74% for 70 mol%.

In conclusion, this thesis successfully developed a novel class of 4D-printed hydrogels with tunable mechanical properties, rapid temperature-induced actuation, and self-healing behavior. By strategically incorporating hydrophobic and hydrophilic monomers, a balance between these functionalities was achieved.



KENDİ KENDİNİ ONARMA VE ŞEKİL HAFIZA ÖZELLİKLERİNE SAHİP VÜCUT SICAKLIĞINA DUYARLI HİDROJELLERİN DÖRT BOYUTLU BASKISI

ÖZET

Dört boyutlu baskı, temelinde üç boyutlu baskı teknolojisine akıllı malzemelerin entegre edilmesiyle oluşturulmuş bir kavramdır. Adında geçen dördüncü boyut zaman parametresi olup, basılan malzemenin zaman içerisinde özelliklerinin farklılaşmasından ileri gelir. Yenilikçi bir üretim yöntemi olarak üç boyutlu baskı, diğer adıyla eklemeli üretim, sağladığı avantajlardan ötürü geleneksel üretim yöntemi olan çıkarmalı üretime hem alternatif hem de tamamlayıcı bir niteliğe sahiptir. Çıkarmalı üretimin aksine eklemeli üretimin, kompleks üretim araçlarına gereksinim duymayarak ve bu sayede lojistik faaliyetlerine ihtiyaç bırakmadan yerinde üretim anlayışıyla çevresel bir yaklaşıma hizmet ettiği söylenebilir. Üç boyutlu baskıda kullanılan hammadde ve üretim sonrası zayıat görece çok daha minimal düzeylerde seyretmektedir. En önemli farklarından biri de çıkarmalı imalatın daha karmaşık üretim sistemleri entegre ederek yerini doldurabileceği kompleks geometriye sahip tasarımları, üç boyutlu baskı teknolojisinin oldukça basit şekilde hayata geçirebilmesidir. Ayrıca, geleneksel üretim yöntemlerinde kullanıcıya özel üretim gerçekleştirmek kullanıcıya uygun bir kalıp hazırlamaktan geçerken, üç boyutlu baskı sistemlerinde herhangi bir kalıba gerek duymaksızın gerçekleşmektedir. Bu sayede hastaya veya müşteriye özel üretim hizmetlerinde kendine sağlam bir yer edinen üç boyutlu baskının günümüz dünyasında gün geçtikçe popülerliği artmakta ve hem akademik dünyada hem de endüstride bu alanda pek çok çalışma yapılmasına neden olmaktadır. Üç boyutlu baskı sistemlerindeki üretim süreci, tipik olarak nesnenin bilgisayar destekli tasarım (CAD) yazılımında tasarlanmasıyla başlar. Üç boyutlu (3D) model daha sonra yazıcının takip etmesi için uygun bir yazılım ile ince katmanlar haline getirilir. Bu aşamadaki temel amaç, dizayn edilen 3D modelin yazıcı tarafından X, Y, Z koordinatlarıyla anlamlandırabileceği bir komut dosyasına çevirmektir. Son olarak yazıcı bu komut dosyasının ışığında, termoplastik bir polimerin ekstrüzyonu, toz halindeki bir malzemenin sinterlenmesi veya sıvı bir reçinenin ışıkla kürlenmesi gibi çeşitli teknikler kullanarak nesneyi katmanlar halinde oluşturur. Bu tez çalışmasında yüksek hassasiyete sahip bir baskılama yöntemi olan ve fotopolimerizasyona dayanan stereolitografi yöntemi kullanılmıştır.

Tez kapsamında öncelikle dört boyutlu baskının hammaddesi olan akıllı hidrojeller üzerinde durulmuş ardından üç boyutlu baskı yöntemleri temel prensipleriyle ele alınmış ve literatürdeki çalışmalardan örnekler gösterilmiştir. Dört boyutlu baskıyı mümkün kılan akıllı malzemeler, uyaranlar açısından ele alındığında, sıcaklık, ışık, elektrik alan, pH, ve solvent gibi çeşitli etkenlerden bahsedilebilir. Bu etkenlere maruziyetleri ile akıllı malzemeler, alternatif bir şekle ve özelliğe kavuşurlar. Bu değişime uğramaları kadar değişimin ne kadar süre zarfında gerçekleştiği de önemli bir araştırma konusudur. Literatürde görüldüğü üzere akıllı hidrojellerin başlıca dezavantajlarından biri olan yavaş harekete geçme, bu malzemelerin gerçek dünya

uygulamalarına ket vurmaktadır. Örneğin uyarıcı faktör eğer solvent gibi difüze olma yoluyla malzemeye etkileşime geçiyorsa, şekil değiştirme aktivasyonu görece daha yavaş gerçekleşmektedir. Doğası gereği solvent moleküllerinin hidrojel polimer zincirleri arasına difüze olması belli bir miktar süreye ihtiyaç duyar. Bu noktada sıcaklık, ışık ve manyetik alan gibi uyaranlar aktivasyon hızı açısından öne geçmektedir. Bunlar arasında sıcaklık, daha doğrudan bir uyaranıdır. Hidrojel kullanım amacına uygun kritik sıcaklığa sahip bir malzemeyi dikkatlice seçerek şekil değişimini programlamak sıcaklık uyarısını makul ve uygulanabilir kılmaktadır. Hidrojellerin bir diğer dezavantajı ise mekanik dayanımsızlıklarıdır. Bu mekanik güçsüzlüğü aşmak için alınan aksiyon hidrojel çapraz bağlar yönünden zenginleştirilmesidir. Zaman içerisinde bozulmaya yatkın olan ve biyolojik sistemler için zararlılık teşkil eden kimyasal çapraz bağlayıcılara bağımlılığı ortadan kaldırmak için literatürdeki önemli bir strateji ise, hidrofobik etkileşimlerin ve hidrojen bağının sinerjik etkilerini kullanan polimerik sistemlerin tasarımına odaklanmaktadır. Fiziksel olarak çapraz bağlanmış hidrojel, kimyasal olarak çapraz bağlanmış muadillerine göre belirgin bir avantaj sunar: Stres durumunda, kimyasal olarak çapraz bağlı hidrojel, uygulanan kuvveti çarpma noktasında yoğunlaştırmaya eğilimlidir ve bu da kopma riskini artırır. Buna karşılık, fiziksel olarak çapraz bağlanmış hidrojel enerjiyi daha eşit bir şekilde dağıtır. Hidrofobik etkileşim sağlayıcı birimler olarak N-süstitüe akrilamidlerin kullanımı oldukça yaygındır. Sıcaklık uyarısına duyarlı bu akrilamidlerde önemli bir diğer parametre ise yan zincirinin uzunluğudur. Literatürde görüldüğü üzere, görece daha uzun alkil yan zincire sahip akrilamidler daha elastik bir karaktere sahiptir. Bu elastikiyet avantajı göz önüne alınarak bu tez çalışmasında hidrofobik etkileşim sağlayan birim olarak yan zincirinde 16 karbon barındıran heksadesil akrilat (C16A) monomeri seçilmiştir. Mekanik mukavemete katkısının ötesinde, C16A monomeri önemli bir pratik avantaj sunar: 18 °C'lik nispeten düşük erime noktası. Bu, reçinenin, yani polimer öncülünün oda sıcaklığında sıvı halde kalmasını sağlar ve piyasada bulunan çoğu reçinede olduğu gibi baskı öncesi ısıtma ihtiyacını ortadan kaldırır. Hidrofilik etkileşimler oluşturmak üzere ise *N*, *N*-dimetil akrilamid (DMAA) ve metakrilik asit (MAA) monomerleri belirlenmiştir. DMAA birimleri hidrojen bağı alıcısı, MAA birimleri ise hidrojen bağı donörü fonksiyonu görmüştür. Bu haliyle, bir hidrofobik ve iki hidrofilik monomer, diphenyl (2,4,6-trimethylbenzoyl) phosphine oxide (TPO) fotobaşlatıcı eşliğinde, herhangi bir çözücü veya çapraz bağlayıcı olmaksızın ticari tip bir stereolitografi (SLA) cihazı yardımıyla polimerize edilmiştir.

Bu tez çalışmasında hidrojel formülasyonları itibarıyla hidrofobik ve hidrofilik monomerlerden oluşturulduğu için, deneyler de iki aşamalı olacak şekilde kurgulanmıştır. İlk aşamada C16A oranı molce %40'ta sabit tutularak hidrofilik iki monomerin birbirine göre mol oranları farklılaştırılmıştır. DMAA-MAA içerisindeki DMAA mol kesri x_{DMAA} olarak ifade edilmiştir. 1–0 arasından belirlenen 10 örnek incelenmiştir. DSC sonuçları, erime sıcaklığı T_m , kristalleşme sıcaklığı T_{cry} ve kristallik kesri f_{cry} tüm hidrojel numunelerinde nispeten sabit kaldığını (sırasıyla yaklaşık 35 ± 1 °C, 26 ± 2 °C ve 27 ± 18) ortaya koymuştur. Bu gözlem, muhtemelen termal davranışa hâkim olan hidrofobik C16A monomerinin sabit içeriğine atfedilebilir. Mekanik test ölçümleri, gerinim hızı $1,89 \times 10^{-2} \text{ s}^{-1}$ olarak sabit tutulmuştur. x_{DMAA} 'daki değişim hidrojel mekanik davranışını önemli ölçüde etkilemiştir. x_{DMAA} değeri 0,5, 0,67, ve 0,8 olan örnekler sünek davranış gösterip toklukları diğer örneklerle kıyasla yüksek seyretmektedir. Maksimum tokluk ise, $2,23 \pm 0,2 \text{ MJ.m}^{-3}$ değeri ile 0,5 x_{DMAA} örneğinde elde edilmiştir ve böylelikle en iyi mekanik özellikler için x_{DMAA} değeri 0,5 olarak belirlenmiştir. Makul olarak hidrojen

bağı alıcısı olan DMAA ve donörü olan MAA'nın molce eşitliğinin de en yüksek verimlilikte hidrojen bağı kurulmasına neden olmakta ve buna bağlı olarak en yüksek mekanik mukavet de x_{DMAA} değeri 0,5 olan hidrojelde gözlenmektedir.

Bu bulgular ışığında deneysel çalışmanın ikinci kısmında x_{DMAA} değeri 0,5 olarak alınarak, diğer bir deyişle eşmolar DMAA ve MAA kullanılarak, C16A monomerinin mol yüzdesi 10 ila 70 arasında değiştirilmiştir. Şişme ile ilgili veriler, C16A içeriğindeki artışla birlikte şişmede azalma olduğunu göstermiştir. Bu bulgu hakkında daha fazla bilgi edinmek için hidrojellerin termal özelliklerini analiz etmek üzere DSC kullanılmıştır. Bu analiz, öncelikle C16A moleküllerinin hidrofobik segmentleri tarafından oluşturulan kristal alanların varlığına dair kanıt sağlamaktadır. Beklendiği gibi, hidrojin kristalin alanlarının erime sıcaklığı C16A mol yüzdesindeki artışla orantılı olarak artmıştır. Bunun, sırasıyla %10 mol ve %70 mol C16A'da 34 °C ila 40 °C arasında değişen geniş bir T_m değerleri aralığıyla sonuçlandığını belirtmek önemlidir. Sonuçlar, C16A miktarıyla hidrojellerin T_m değerleri arasında bir korelasyon olduğunu göstermektedir. Hidrojellerin f_{cry} değeri mevcut C16A miktarı ile orantılı olarak sırasıyla %10 ve %70 C16A örnekleri için, $0,8 \pm 0,2$ 'den $39,7 \pm 3$ 'e kadar artmıştır. Bu durum, C16A birimleri içinde daha uzun hidrofobik segmentlerin varlığına bağlı olarak oldukça düzenli ve kararlı kristal yapıların oluşumuna bağlanabilir. Bu iyi tanımlanmış kristalin alanlar erimek için daha yüksek bir sıcaklık gerektirir ve bu da T_m 'de gözlemlenen artışa neden olur. Yapılan XRD ölçümleri sonucunda C16A mol yüzdesi arttıkça amorf polimerlerin karakteristiği olan 16° civarında geniş pik, yerini git gide keskinliği artan 21,3° pike bırakmaktadır. 21,3°'deki bu daha keskin pik, 0,42 nm'lik bir d aralığına karşılık gelir. Bu değer, yan yana paketlenmiş, parafin benzeri alkil zincirlerine sahip kristal bir yapı için tipiktir. Bizim durumumuzda, hidrojel ağı içindeki C16A birimlerinden kaynaklanan hegzadesil yan zincirlerinin hizalanmasını ifade eder. 0,42 nm'lik d aralığı, aynı polimer omurgası boyunca hizalandığında iki komşu hegzadesil yan zinciri arasındaki ortalama mesafeyi yansıtmaktadır. Daha yüksek C16A içeriği ile artan kristallliği gösteren XRD'den elde edilen bu gözlemler, daha fazla C16A içeren hidrojeller için daha yüksek erime sıcaklıkları gösteren DSC sonuçları ile iyi bir uyum içindedir. Test edilen tüm hidrojeller, faz geçiş sıcaklıklarının altında ve üstünde depolama modülü G' ve kayıp faktörü $\tan \delta$ 'da net tersinir değişiklikler göstermektedir. Bu değişikliklerin şiddetinin hidrojellerde bulunan C16A miktarı ile büyük ölçüde orantılı olduğunu belirtmek önemlidir. Faz geçiş sıcaklığının altındaki sıcaklıklarda, artan kristallik derecesi nedeniyle artan C16A içeriği sonucu modül de yükselişe geçer. Tersine, faz geçiş sıcaklığının üzerindeki sıcaklıklarda, artan C16A içeriği ile modül azalma görülür. Geçiş sıcaklığının altında, tüm hidrojeller için $\tan \delta$ 'da bir azalma eğilimi gözlenmektedir. Bunun nedeni, polimer zincirlerinin daha yoğun bir şekilde paketlenmiş olması ve kristallikteki artışın bir sonucu olarak daha fazla sertliğe sahip olmasıdır. Hidrojellerin T_m değerinin üzerinde, fiziksel çapraz bağlanma noktaları olarak hareket eden kristalin alanlar tersinir şekilde eriyerek nispeten yumuşak hale gelmelerine neden olmaktadır. Bu çerçevede reolojik ölçümlerin de DSC ölçümleri ile uyumlu olduğu görülmüştür. Mekanik testlere geçildiğinde, hem kırılma gerilimi σ_f hem de kırılma gerilmesi ϵ_f , artan C16A içeriği ile artış göstermekte ve %50 mol'de maksimuma ulaşmaktadır. Sonrasında ise, bu değerler %50 mol'den daha yüksek C16A konsantrasyonlarında hafifçe düşmektedir. Sonuç olarak, %50 mol C16A içeren hidrojel maksimum $50,8 \pm 0,7$ MPa Young modülü E ve $8,15 \pm 0,7$ MJ.m⁻³ tokluk W değerleriyle üstün mekanik özellikler sergilemektedir. Dolayısıyla, genel bulgular C16A'nın optimum %50 mol seviyesinde dengeli bir şekilde dahil edilmesinin bu hidrojellerde üstün mekanik özellikler elde etmek için çok önemli olduğunu

göstermektedir. Hidrojelin ağ yapısı, mekanik özelliklerini etkileyen farklı hızlardaki strese farklı tepkiler verir. Daha hızlı deformasyon (yani daha yüksek gerilme hızı $\dot{\epsilon}$) hidrojenleri daha sert ve güçlü (daha yüksek E , σ_y ve σ_f , ancak daha düşük ϵ_f) hale getirir. Akma gerilimi σ_y , polimer segmentlerinin akma noktasında bir enerji bariyerini aşması gerektiğini varsayan Eyring teorisine göre $\dot{\epsilon}$ gerinim oranının logaritması ile doğru orantılıdır. Bu etkiyi değerlendirmek için, $\dot{\epsilon}$ gerinim hızı $0,95 \times 10^{-2} \text{ s}^{-1}$ ile $4,54 \times 10^{-2} \text{ s}^{-1}$ aralığında tutularak mekanik test gerçekleştirilmiştir. Hidrojellerin modülü E 'nin, akma gerilmesi σ_y 'nin ve kırılma gerilmesi σ_f 'nin artan gerilme hızıyla arttığını, kopma gerinimi ϵ_f 'nin ise azaldığını ortaya koymaktadır. Ayrıca bu analiz, $2,3 \pm 0,2 \text{ nm}^3$ 'lük bir aktivasyon hacmi V_a 'yı işaret etmektedir. Bu, deformasyon sürecine dahil olan polimer segmentlerinin uzunluğunun, yani Kuhn uzunluğunun, iki komşu karbon atomu arasındaki bağın uzunluğuna eşdeğer olan yaklaşık 1,3 nm olduğunu gösterir. Bu, test edilen gerinim oranlarında her bir bağın mekanik tepkide önemli bir rol oynadığının bir göstergesidir.

Sıcaklığa bağlı olarak tersinir şekilde uyarılabilir alanların varlığı göz önünde bulundurulduğunda, hidrojenlerin şekil hafıza özellikleri bükme testinden yararlanılarak ortaya kondu. Hidrojellerin açılma şekil geri kazanım oranı $R_r(\theta)$, sıcaklıktaki her 1 °C'lik artış için bükülmüş numunenin geri kazanım açısı kaydedilerek hesaplanmıştır. Hidrojellerin açılma sabitleme oranı $R_f(\theta)$ ise C16A mol yüzdeleri 40, 50, 60 ve 70 olan numuneler için sırasıyla %98, %98,8, %99,2 ve %99,4 olarak hesaplanmıştır. Kayda değer şekilde, 40 mol C16A içeren numune hariç tüm numuneler %100'lük bir şekil geri kazanım oranı elde etmiştir. Bu yetenek hidrojenlerdeki yüksek kristaliteye bağlanabilir. Özellikle, test edilen tüm numuneler için 34 °C – 36 °C civarında şekil geri kazanım oranında önemli bir artış gözlenmiş ve C16A miktarının artması daha hızlı bir açılma geri kazanımla sonuçlanmıştır. Numunelerin kendini iyileştirme davranışı bir dizi deneyle gözlemlenmiştir. Numuneler bir jilette çizilmiş ve ardından 10 dakika boyunca 60 °C'lik sulu bir ortama daldırılmıştır. Sonuçlar C16A içeriği ile iyileşme derecesi arasında açık bir korelasyon olduğunu ortaya koymaktadır. %50 mol C16A içeren numunede çizik tamamen iyileşmemiş olsa da genişliği ortalama 82,3 μm 'den 44,8 μm 'ye önemli ölçüde azalmaktadır. %60 mol C16A ile ilgili olarak, çizik neredeyse iyileşmiş ve genişliğin başlangıçtaki 29,9 μm 'den ortalama olarak sadece 5,7 μm 'ye düştüğünü göstermiştir. Bunun da ötesinde, %70 mol C16A içeren numunenin, 109,6 μm genişliğindeki çizik ortadan kalkmasıyla tamamen iyileştiği ortaya konmuştur. Hidrojellerin kendini iyileştirme yetenekleri hem iyileşme öncesi hem de iyileşmiş durumları göz önünde bulundurularak gerçekleştirilen mekanik testler üzerinden değerlendirilmiştir. Artan C16A içeriğinin E , σ_f , ϵ_f verimliliklerini önemli ölçüde artırdığı bulunmuştur. E verimlilikleri %50 mol için %91,8, %60 mol için %97,4 ve %70 mol C16A için %99,5'tir. σ_f verimlilikleri %50 mol için %83,9, %60 mol için %87,8 ve %70 mol C16A için %88,5'tir. ϵ_f verimliliği başlangıçta %50 mol C16A'da (%11,5) geride kalmış, ancak daha yüksek C16A içeriği ile iyileşme göstererek %60 mol için %63,2'ye ve %70 mol için %74'e ulaşmıştır. Genel olarak, C16A miktarının oranının artırılması hidrojel iyileşmesini son derece arttırmıştır.

Sonuç olarak, bu tez, ayarlanabilir mekanik özelliklere, vücut sıcaklığına hızlı cevap veren ve kendi kendini iyileştirme davranışına sahip yeni bir 4D-baskılı hidrojel örneğin, başarıyla hayata geçirmiştir. Bu tez çalışması, hidrofobik ve hidrofilik monomerleri stratejik olarak bir araya getirerek ve oranlarını optimize ederek, bu işlevler arasında arzu edilen bir denge kurmayı başarmıştır.

1. INTRODUCTION

One can envision a future in which bandages heal themselves, contact lenses deliver targeted drugs directly to the eyes, and soft robots adapt to their environment. The development of a new class of materials, hydrogels, has the potential to revolutionize a range of fields, offering a significant step forward in scientific research. Hydrogels are a specialized type of polymer, essentially a macromolecule built from smaller units, that possess these remarkable properties.

Polymers have been profoundly impacted in shaping our world for centuries. These macromolecules are formed by repeated connections of subunits called monomers. Polymers can be tailored to exhibit a remarkable range of properties, offering flexibility like rubber, rigidity comparable to metal or concrete for structural purposes, or transparency like glass for situations demanding clear visibility (Hsissou et al., 2021). The versatility of these materials positions them for use in numerous applications. Their widespread adoption in specific usage, including clothing (Li et al., 2024), adhesives (Lancker et al., 2024), and electronics (Lolaeva et al., 2024), is already evident.

Besides, scientific advancements have led to the creation of specialized materials with extraordinary capabilities. One such example is hydrogels, which are elastic networks of natural or synthetic components that can absorb and hold large amounts of water (Ullah et al., 2015) while preserving their structure. In order to develop more advanced and specialized materials, research in the field of hydrogels is actively seeking to create materials that can respond and adapt to environmental stimuli (Neuhaus et al.; Walther, 2019), so-called “smart materials” or “stimuli-responsive materials” (Tiwari & Kobayashi, 2013).

While smart hydrogels offer remarkable capabilities, their potential applications are often limited by two key factors: mechanical strength (Billiet et al., 2012) and the actuation speed (Li et al., 2020) of their shape-memory function. The inherent mechanical strength limitations of conventional hydrogels can be effectively addressed through the strategic design of polymeric systems that utilize the synergistic effect of

hydrophobic interactions and hydrogen bonding (Zhang et al., 2018; Chang et al., 2018; Xia et al., 2017). This innovative method opens up the possibility of developing mechanically robust gels via physical crosslinking, eliminating the need for chemical cross-linking and its potential toxicological concerns (Coombes et al., 2002) associated with introducing reactive species into living systems. Yet the development of mechanically robust hydrogels is crucial, a significant limitation of hydrogel-based actuators lies in their response rate, particularly when the mechanism of shape recovery hinges on the diffusion of water molecules (Jian et al., 2022). Diffusion-based actuation in hydrogels often suffers from slow response times, alternative stimuli like temperature, light, and electric fields can induce significantly faster actuation (Wang et al., 2022; Hu et al., 2020). Among the alternatives, temperature-induced shape-memory systems stand out due to their simplicity and biocompatibility. Unlike methods utilizing pH or UV (ultraviolet) light, which can potentially damage cells, temperature variations within a range between 4 °C and 37 °C are well-tolerated, making them ideal for biological applications. (Stroganov et al., 2014; Zheng et al., 2021).

Smart hydrogels often require innovative approaches for their fabrication. In this context, 3D printing, a transformative technology that empowers a paradigm shift in production by eliminating the need for traditional molds and tooling, offers a powerful tool for the fabrication of hydrogels. This precise, layer-by-layer deposition allows for the creation of complex structures with tailored properties, significantly enhancing the functionality of the smart materials (Chen et al., 2019). 3D printing encompasses a diverse range of technologies, each utilizing different materials in various states. Common methods include fused deposition modeling (FDM) which employs solid filament materials (Bakar et al., 2010), selective laser sintering (SLS) that utilizes powdered polymers (Mazzoli, 2012), and stereolithography (SLA) which relies on liquid resins (Bagheri & Jin, 2019). In the scope of this thesis study, an SLA type 3D printer is selected for hydrogel preparation due to its high precision and suitability to acrylic-based precursors. Basically, the prepared liquid precursors were poured into the tank of SLA printer at room temperature (25 ± 2 °C) and printing process is initiated. Once the printing process is accomplished, the samples were scraped from the printer's platform and immersed into a large volume of water in order to facilitate physical-crosslinking.

This technology enables the creation of customized pharmaceuticals and medical devices with high precision, thereby providing researchers with a valuable tool for advancing the field of personalized medicine (Beg et al., 2020). The growing number of tissue engineering patents reflects the intensifying research into the potential of 3D printing for advanced medical applications (3D Printing Technology Insight Report, 2014). Moreover, the estimated market size of 3D-printed pharmaceuticals, which is projected to reach 522 million U.S. dollars by 2030 in a moderate growth scenario, further highlights the growing interest and potential of this technology in drug delivery (3D Printed Drugs Market Size, Share, Competitive Landscape and Trend Analysis Report by Spritam and Other Drugs Potential Market by Scenario- Global Opportunity Analysis and Industry Forecast, 2017-2030, n.d.).

The boundaries of 3D printing are expanding with the incorporation of smart materials. This allows for the creation of printed components that can change their shape or properties over time (Khoo et al., 2015). This innovative approach has given rise to a new concept called 4D printing (Tibbits, 2014). Theoretically speaking, 4D printing approach introduces the concept of time as a fourth dimension to the spatial dimensions of length, width and depth (Trenfield et al., 2019). These rearrangements of matter mainly occur as a consequence of external stimuli. When exposed to the stimuli, these materials undergo significant shape and property transformations. This can include swelling, stretching, bending, twisting, or rolling, depending on the pre-designed geometry of the material (Falahati et al., 2020). Thus, this novel technology enables printed components to undergo transformation or adaptation following their printing. The potential for this technology to facilitate the development of advanced biomedical applications is considerable.

Following their creation on 3D printers, these objects, composed of smart materials, can transform in response to their environment by various environmental factors, including temperature, solvent, pH, electrical or magnetic fields, or light. Among these, temperature is the most widely utilized stimuli in responsive polymer systems. Temperature variations offer a particularly attractive approach for triggering material responses due to their ease of control and broad applicability across various systems (Gil & Hudson, 2004). In the initial stage, the 4D printed structure is heated to a temperature (T_{high}) exceeding the critical temperature (i.e. glass transition temperature or melting temperature). At this elevated temperature T_{high} , the material exhibits a more

pliable nature. This allows for the application of a controlled strain, essentially stretching the structure to a predetermined extent. While maintaining the stretched configuration, the temperature of the structure is lowered to another temperature (T_{low}) that falls below the critical temperature. As the material cools, it regains stiffness and essentially locks into the stretched shape due to the presence of external stress. Once the temperature reaches T_{low} , the external stress that was maintaining the stretched shape is removed. Remarkably, the material retains its newly acquired, elongated form even in the absence of the applied stress. To return the structure to its original shape, the process is reversed. The structure is simply reheated to the high temperature (T_{high}) in a stress-free state. At this elevated temperature, the material regains its flexibility, allowing it to return to its original configuration. This thermo-responsive approach offers a versatile method for programming temporary shapes into 4D-printed objects. By carefully selecting a material with a suitable critical temperature and precisely controlling the heating and cooling cycles, it is possible to achieve a desired temporary shape and subsequently revert to the original configuration. In this study, the control over the actuation temperature of hydrogels is well maintained with the C16A content in hydrogels.

In summary, this research addresses the key limitations of traditional hydrogels by developing a novel polymer with a remarkable combination of shape-memory, self-healing capabilities, and superior mechanical properties. Notably, these tough hydrogels achieve rapid actuation (less than 30 seconds) with a tunable actuation temperature, enabling precise control over their shape-shifting behavior. This research employed a liquid-based 3D printing technique (stereolithography apparatus, SLA) for hydrogel fabrication. These unique properties position this material as a promising candidate for various advanced applications. While this research focuses on establishing a fundamental understanding of these functionalities, the potential for future exploration extends to diverse fields beyond the scope of this work. For instance, the tunable actuation temperature opens exciting possibilities in areas like drug delivery systems and biomedical applications in terms of targeting a specific body part. Additionally, further investigation could explore the suitability of this material for biomedical applications, leveraging its unique properties for the design of innovative devices with controlled shape changes triggered by specific temperatures.

2. THEORETICAL PART

True to their name, polymers, also known as “macromolecules”, reflect their structure by drawing their name from the Greek words “poly” (many) and “meres” (parts) (Namazi, 2017). The 1920s saw a paradigm shift in the understanding of polymers. While the prevailing understanding held them as mere aggregates, Hermann Staudinger revolutionized the field by proposing the concept of macromolecules. These large molecules, according to Staudinger, were formed by long chains of subunits linked by strong covalent bonds (Young & Lovell, 2011).

Expanding on the definition of macromolecules presented so far, it is important to recognize that their skeletal structures can vary. The arrangement of building blocks within a polymer is known as its skeletal structure. While some polymers possess a linear arrangement, resembling a chain with two ends, many others exhibit more complex, non-linear configurations. Figure 2.1 provides a visual representation of these diverse structures.

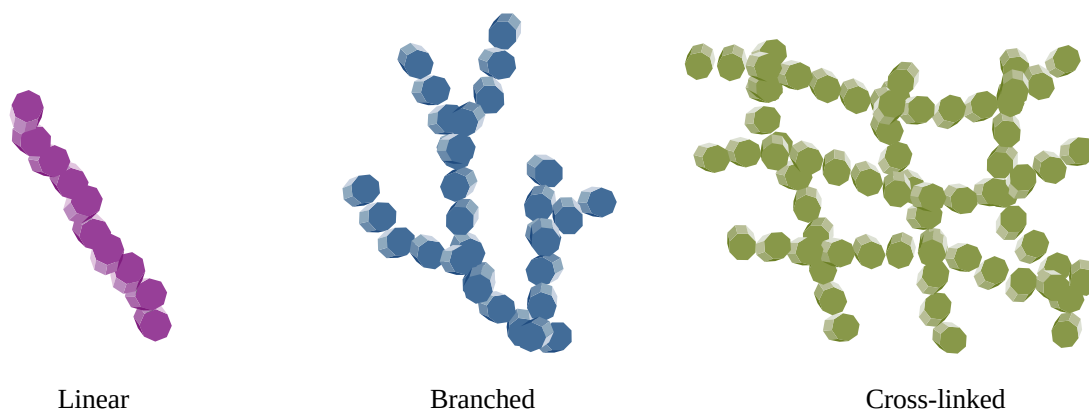


Figure 2.1 : From left to right, skeletal structures of linear, branched, and cross-linked polymer structures.

Branched polymers are characterized by extensive side chains branching off at junction points along the main chain and they are characterized by the size and quantity of these branches. Cross-linked polymers deviate significantly from their linear counterparts. They possess an intricate three-dimensional structure where individual chains are interconnected through a network of junction points and interactions with

other chains. The extent of crosslinking is quantified by crosslink density or degree of crosslinking, which directly relates to the number of junction points present per unit volume. The skeletal structure has a profound impact on the performance of the polymer. For instance, linear polyethylene melts at a temperature 20 °C higher than branched polyethylene due to the higher packing density in linear chains (Young & Lovell, 2011; Vega et al., 2022).

As a result of interactions between the chains, polymers can exist in two fundamental arrangements: crystalline and amorphous. In the crystalline phase, polymer chains exhibit a highly ordered, regularly folded structure. This arrangement resembles a neatly stacked arrangement of building blocks. Conversely, the amorphous phase features chains in a more random, entangled configuration. These two phases are present in varying degrees within most polymers, so-called semicrystalline polymers. Furthermore, some polymers are entirely amorphous, but no polymer exists in its entirety in a crystalline form. In Figure 2.2, polymer structures in the increasing crystallinity order are represented schematically.

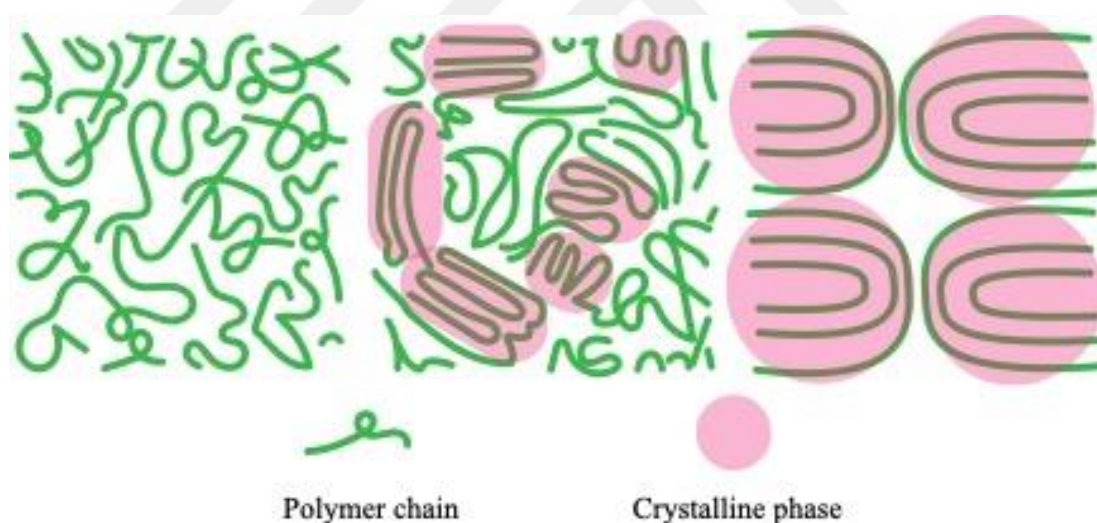


Figure 2.2 : Schematic illustration of polymer structure, amorphous to highly-ordered crystalline structure with an increase in crystallinity from left to right.

2.1 Hydrogels

The advancement of scientific knowledge has led to the development of materials with remarkable properties. One such material is the hydrogel, which is a three-dimensional network formed from natural or synthetic building blocks. These materials have the ability to absorb significant amounts of water without losing their shape (Ullah et al.,

2015). The pioneering work of Wichterle and Lim on crosslinked 2-hydroxy-ethyl methacrylate hydrogels has captivated the attention of biomaterial scientists for decades (Wichterle & Lim, 1960). The distinctive characteristics of hydrogels, including low interfacial tension, tissue-mimicking softness, high permeability for small molecules, and controlled release of encapsulated molecules, have established them as a significant area of research in a range of biomedical fields (Ebara et al., 2014).

A dry hydrogel is a three-dimensional network of polymer chains that are cross-linked together either chemically or physically (Ullah et al., 2015). The cross-linked chains prevent the polymer from dissolving in water, yet they allow water to diffuse into the gel. Thus, when a dry hydrogel is immersed in water, the water molecules diffuse into the pores of the hydrogel, causing it to swell and increase in volume depicted in Figure 2.3. The degree of swelling of a hydrogel is dependent upon the properties of the hydrogel, including the type of polymer chains, and the degree of cross-linking. The swelling/deswelling of a hydrogel is a reversible process. If a swollen hydrogel is removed from water, it will eventually dry out and return to its original size. The removal of a swollen hydrogel from the water will result in its eventual drying out and return to its original size. The rate of swelling and deswelling is dependent on the same factors that affect the degree of swelling.

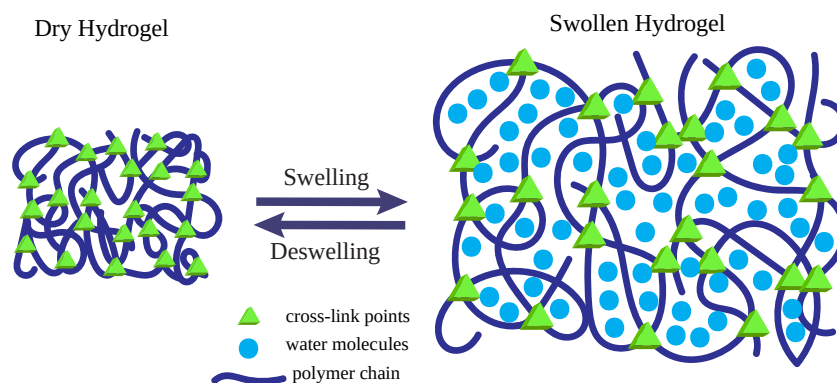


Figure 2.3 : Illustration of swelling and deswelling in hydrogels.

The Flory-Rehner theory provides a framework for understanding how neutral hydrogels (those without ionic groups) swell when exposed to water (Holback et al., 2011). Within the framework of the Flory-Rehner theory, a crosslinked polymeric gel immersed in a solvent achieves equilibrium through a delicate interplay between two opposing thermodynamic forces. The first force arises from the favorable free energy

of mixing between the polymer network and the solvent molecules (Peppas, 2000). This force drives the expansion of the gel. Conversely, the elasticity of the polymer chains inherently resists this expansion, exerting a retractive force (Peppas, 2000). At equilibrium, these two forces become equal, establishing a stable state for the swollen gel. Equation (2.1) expresses the underlying thermodynamics of the system through Gibbs free energy.

$$\Delta G_{total} = \Delta G_{elastic} + \Delta G_{mixing} \quad (2.1)$$

The total Gibbs free energy (ΔG) can be decomposed into two principal contributions: $\Delta G_{elastic}$ and ΔG_{mixing} . The energetic burden associated with the deformation of polymer chains during swelling is represented by $\Delta G_{elastic}$. Specifically, it quantifies the elastic resistance of the network to expansion. Conversely, ΔG_{mixing} reflects the favorable interaction between the polymer chains and the solvent molecules. As a consequence, neutral hydrogels achieve equilibrium swelling due to a balance between these two opposing forces: the attractive force for solvent mixing and the internal contractile force of the polymer network (Holback et al., 2011).

The versatility of hydrogels is evident in their successful applications as wound dressing (Li et al., 2020; Ribeiro et al., 2009), superabsorbent (Kabiri & Zohuriaan-Mehr, 2004; Pourjavadi et al., 2006), and tissue engineering scaffolds (Naahidi et al., 2017; Mawad et al., 2012). One particularly promising application of hydrogels is in drug delivery systems (Ashley et al., 2013; Ischakov et al., 2013) since these networks not only protect medications from harsh environments but also offer controlled drug release through a unique mechanism. The hydrogel structure is capable of adapting to environmental changes, which allows for the controlled release of encapsulated drugs. These responsive hydrogels, which exhibit properties resembling those of a sensor, can undergo reversible transitions in their volume or state (gel to liquid) upon subtle shifts in the surrounding conditions (Ebara et al., 2014). This responsiveness, stemming from the hydrogel's ability to adapt to its environment, justifies the term “smart”.

Smart hydrogels are fascinating materials that can undergo changes in their properties in response to a variety of external stimuli. These stimuli can be physical, such as temperature or light, or chemical, such as changes in acidity or the presence of specific molecules (Peppas & Khare, 1993; Li & Yew, 2009; Aguilar & Román, 2019). This

versatility makes them highly useful for a wide range of applications, including drug delivery (Shriky et al., 2020), sensors (Zhao et al., 2018), and bioseparation applications (Dong et al., 2005).

Despite the remarkable advancements that have been made in the development of smart hydrogels, their real-world applications are often hindered by limitations in mechanical strength. This factor significantly impacts their suitability for various envisioned uses. For instance, applications in soft robotics or tissue engineering necessitate hydrogels capable of withstanding substantial mechanical stresses. Insufficient mechanical strength can result in catastrophic failure, such as a soft robotic gripper losing its shape or a tissue scaffold fracturing under pressure. Consequently, it is vital to enhance the mechanical properties of hydrogels.

Researchers are actively investigating novel strategies to overcome the limitations of mechanical strength in smart hydrogels. One promising approach focuses on the design of polymeric systems that utilize the synergistic effects of hydrophobic interactions and hydrogen bonding (Zhang et al., 2018; Chang et al., 2018; Xia et al., 2017). This innovative design approach offers a significant advantage: the utilization of physical cross-linking. This eliminates the reliance on chemical crosslinkers, which are susceptible to degradation over time. Moreover, chemical crosslinkers have the potential to release harmful byproducts. Such byproducts may have a negative effect on the surrounding environment, particularly in biomedical applications.

In applications that require impact resistance, physically crosslinked hydrogels offer a distinct advantage over their chemically crosslinked counterparts. In the event of stress, chemically crosslinked hydrogels are prone to concentrating the applied force at the point of impact, thereby increasing the risk of failure. In contrast, physically crosslinked hydrogels exhibit a more even distribution of energy dissipation (Abdurrahmanoglu et al., 2009). This mechanism contributes to their superior mechanical strength, as illustrated in Figure 2.4.

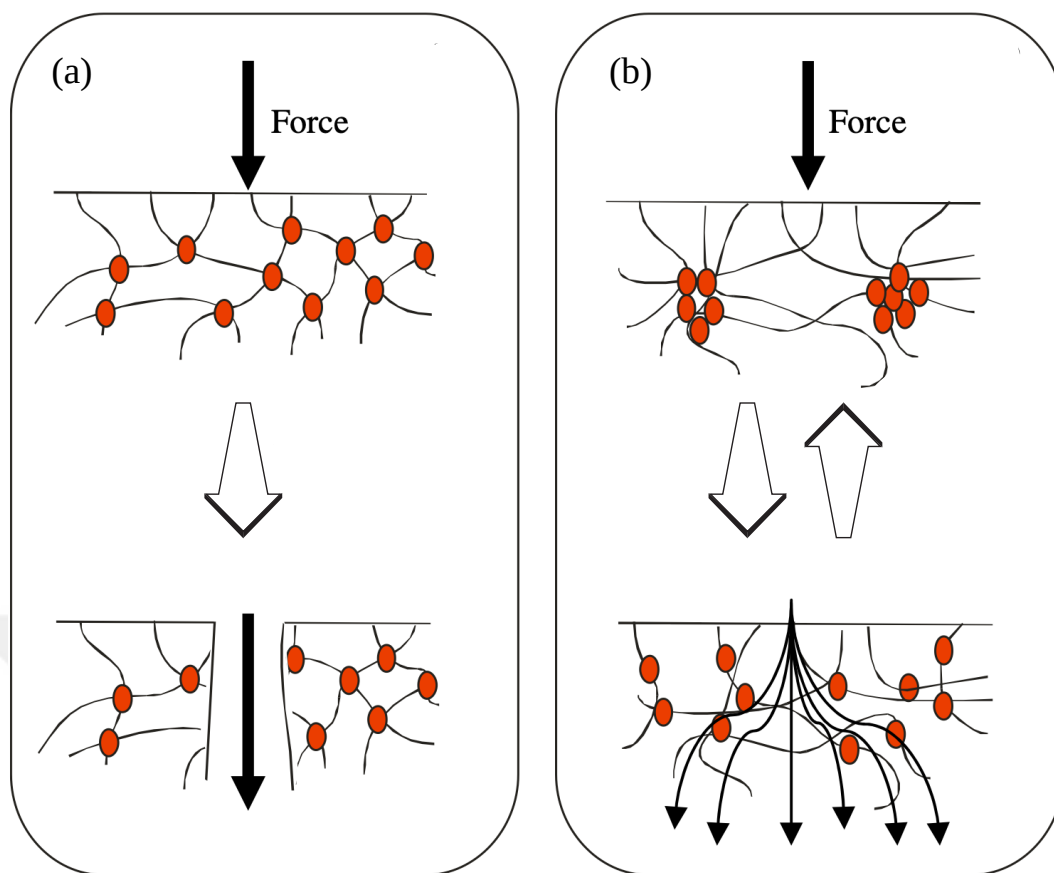


Figure 2.4 : Cartoon for comparison of (a) chemically crosslinked and (b) physically crosslinked hydrogel structures under stress exposure (Abdurrahmanoglu et al., 2009).

Beyond mechanical strength, another critical factor impacting the practical applications of smart hydrogels is the actuation rate of their shape-shifting mechanism (Li et al., 2020). Smart hydrogels can be triggered by various stimuli, including temperature, light, electric field, or swelling-deswelling mechanisms. While swelling-based responses are inherently slower (Jian et al., 2022), stimuli like temperature, light, or electric field can induce much faster alterations in shape (Wang et al., 2022; Hu et al., 2020).

Among the responsive hydrogels, thermo-responsive hydrogels, in particular, are a widely studied class due to their rapid response (Bromberg, 1998; Ferreira et al., 2018) and straightforward actuation mechanism. These hydrogels exhibit a reversible sol-gel phase transition, or a significant change in volume, triggered by a specific temperature (Park et al., 2019). This critical temperature is designated as the lower critical solution temperature (LCST).

At room temperature, some hydrogels demonstrate complete miscibility with water. However, as the temperature increases, the mutual solubility of the hydrogels decreases. This reduction reaches a critical point known as the lower critical solution temperature (LCST), beyond which the hydrogels undergo phase separation, transitioning into a gel state. Below this LCST, the hydrogel regains its solubility. Conversely, the upper critical solution temperature (UCST) refers to a phenomenon observed in some hydrogels that cooling triggers gel formation, followed by phase separation upon further cooling below a critical point. This dual responsiveness to temperature changes makes these kinds of hydrogels highly useful for various applications.

Poly(*N*-isopropyl acrylamide) (PNIPAAm) is a popular thermoresponsive polymer synthesized from *N*-isopropyl acrylamide (NIPAAm). PNIPAAm exhibits a reversible transition from transparent to opaque at its lower critical solution temperature (LCST), which is approximately 34 °C (Ashraf et al., 2016). This distinctive property renders PNIPAAm a valuable material in a range of biomedical applications, including drug delivery and tissue engineering. PNIPAAm's ability to undergo reversible phase transitions at its lower critical solution temperature enables precise control over cell detachment and attachment processes, making it advantageous for biomedical applications. For instance, a study conducted by Kumashiro et al. (2010) demonstrated the use of PNIPAAm polymers covalently grafted to culture dishes in the field of tissue engineering. A variety of cell types were cultivated and proliferated on the culture dishes at a temperature of 37 °C, which is the body temperature. By simply lowering the temperature below 32 °C (to 20 °C), the cells spontaneously detached from the surface within an hour, eliminating the need for the use of harsh enzymes. Figure 2.5 illustrates this detachment in comparison to traditional culture surfaces. The detachment is contingent upon the phase transition of PNIPAAm. Above the lower critical solution temperature, PNIPAAm becomes hydrophobic and transforms to a viscous gel that attaches firmly to cells. Conversely, following the cooling process below its lower critical solution temperature LCST, PNIPAAm undergoes a hydrophilic transformation, resulting in the detachment of cells into a single sheet, which is ideal for the transplantation of cells in a neat and orderly manner.

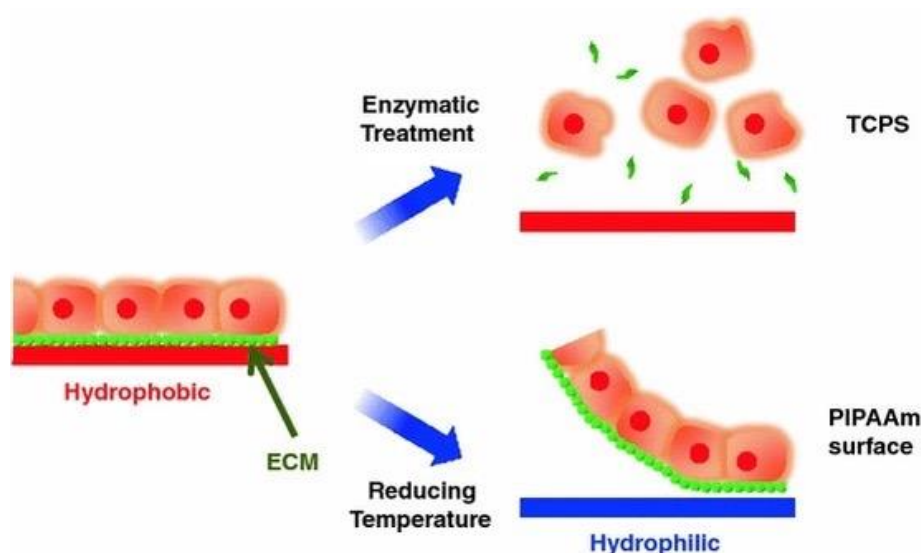


Figure 2.5 : Cell attachment-detachment control via thermo-responsive PNIPAAm compared to enzymatic treatment (Kumashiro et al., 2010).

The chemical composition of the polymer network dictates how hydrogels respond to their environment. A defining characteristic of thermo-responsive hydrogels is their ability to undergo a transition from a solid to a liquid state upon cooling, and vice versa upon heating (Osada & Matsuda, 1995). These materials adopt a new shape when heated and subsequently cooled. Furthermore, reheating above the transition temperature enables them to revert to their original form.

The temperature-triggered actuation of hydrogels can also be dependent on the melting transition temperature of their crystalline domains (Liu et al., 2020; Bilici et al., 2016; Inomata et al., 2012). For instance, Figure 2.6 illustrates the heating/cooling-triggered shape fixing and shape recovery: a dog-bone-shaped sample's shape fixing upon cooling in (a) and the angular recovery of a curved sample upon heating above its actuation temperature in (b). The shape-memory behavior of these types of gels is often attributed to their hydrophobic composition. *N*-substituted acrylamides represent a significant class of temperature-responsive polymers. In a study by Shiblee et al. (2018), they were able to synthesize shape-memory gels that were composed of a blend of hydrophilic and hydrophobic monomers. *N*, *N*-dimethyl acrylamide (DMAA), the hydrophilic component, readily interacts with water. In contrast, the shape-memory properties were derived from the incorporation of hydrophobic crystalline monomers, stearyl acrylate (SA) and lauryl acrylate (LA). The transition temperatures of these gels, which range from 29 to 49.5 °C, were influenced by the specific content of SA and LA monomers incorporated during synthesis (Shiblee et al., 2018).

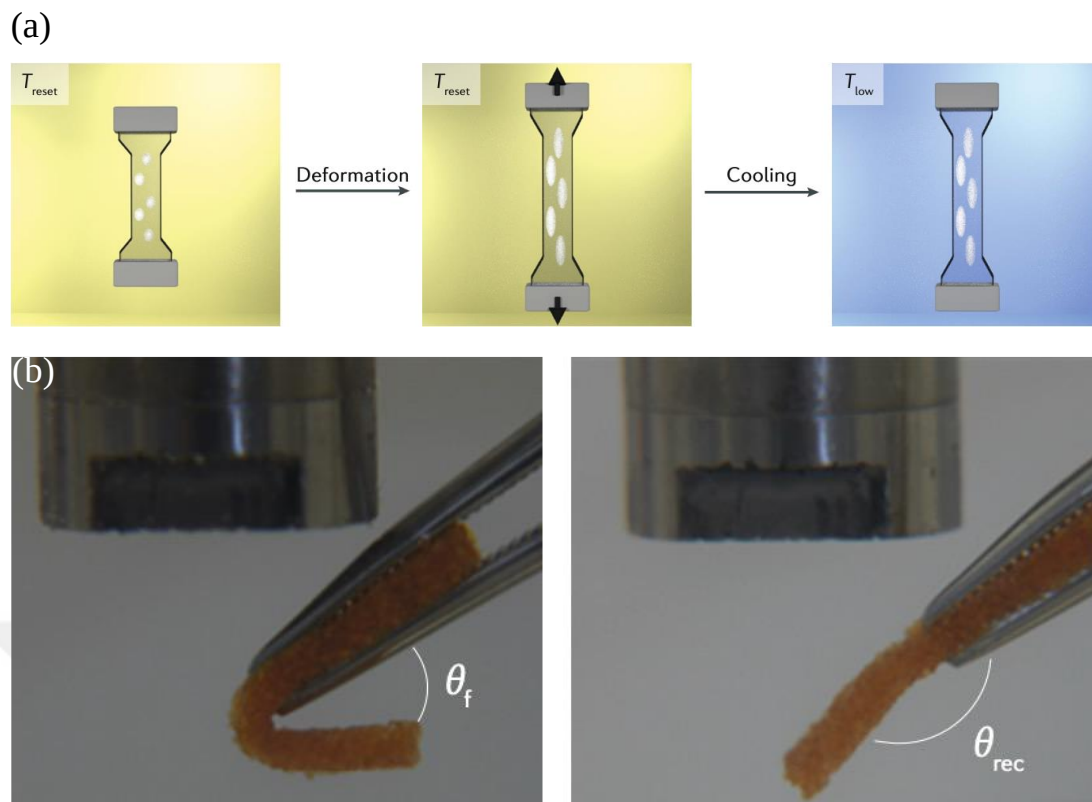


Figure 2.6 : Shape-memory behavior: (a) Shape fixing of the dog-bone sample, (b) angular recovery of a curved sample (Lendlein & Gould, 2019).

Building upon existing knowledge, our research group has demonstrated that hydrogels comprised of *N*-alkyl acrylates having longer side chains exhibited a lower loss factor ($\tan \delta$) compared to their shorter-chained counterparts (Tuncaboylu et al., 2012; Okay, 2019). This loss factor $\tan \delta$ is an indication of the energy dissipated within the material during deformation. Lower $\tan \delta$ values indicate a more elastic material, suggesting that *N*-alkyl acrylates with longer alkyl chains result in hydrogels with increased elasticity compared to their methacrylate counterparts (Okay, 2019).

To exploit these elasticity benefits and create effective hydrogels, this study employs hexadecyl acrylate (C16A), as a hydrophobic monomer. The long alkyl side chains, comprising 16 carbons, facilitate the formation of hydrophobic interactions within the hydrogel. Beyond its contribution to mechanical strength, C16A monomer offers a significant practical advantage: its relatively low melting point of 18 °C. This ensures that the resin, i.e. the precursor, remains in a liquid state at room temperature. This aligns well with the most commercially available resins, thus eliminating the need for pre-printing heating.

In addition to C16A, hydrophilic monomers are also incorporated into the hydrogels in order to further enhance their mechanical strength. The incorporation of hydrophilic monomers introduces hydrogen bonding, with *N, N*-dimethyl acrylate (DMAA) acting as a hydrogen bond acceptor and methacrylic acid (MAA) serving as a hydrogen bond donor. In hydrogels containing both hydrophilic and hydrophobic moieties, the delicate balance between these moieties is significant in defining the material's properties (Kasiński et al., 2020).

Smart hydrogels encompass a diverse range of materials, including self-healing hydrogels, which possess the remarkable ability to repair themselves upon damage. The process of self-repair is facilitated by the interplay of covalent bonds or non-covalent interactions (Devi V. K. et al., 2021). Hydrogels formed through non-covalent interactions are typically more flexible and possess more self-healing than their covalent counterparts (Saunders & Ma, 2018; Jeon et al., 2016). This is due to the fact that the temporary bonds in these gels can be readily dissociated, allowing the material to adapt and repair itself. In contrast, hydrogels formed through covalent bonds are highly stable but lack the flexibility and self-healing abilities of their non-covalent counterparts.

To obtain the self-healing ability in this thesis study, large hydrophobic segments of C16A and hydrogen bonding between DMAA and MAA were incorporated. The hydrogen bonds act like microscopic switches for the hydrophobic parts of the hydrogel. These switches allow the hydrophobic domains to break apart and reconnect easily. This makes the fractured hydrogel self-heal quickly and efficiently (Fang et al., 2020).

2.2 3D Printing

3D printing represents a revolutionary technology in the field of additive manufacturing, characterized by its methodical layer-by-layer construction of objects from digital designs. This contrasts strictly with traditional subtractive manufacturing approaches, which involve the removal of material from solid blocks (Guzzi & Tibbitt, 2019). The additive nature of 3D printing allows for the creation of complex geometries with unprecedented speed and efficiency, thus representing a pivotal advancement in manufacturing and prototyping.

The process begins with the creation of a detailed three-dimensional (3D) model using specialized computer-aided design (CAD) software. The digital blueprint is then sliced into numerous thin layers, effectively converting the model into a series of instructions in the form of X, Y, Z coordination that guide the 3D printer. The printer employs a variety of techniques, dependent on the utilized material, including extrusion of melted polymers, selective sintering of powdered materials, and curing of liquid resin through light exposure. These methods permit the printer to deposit or solidify materials layer by layer in a precise manner, thereby gradually building up the final object.

To broaden the applications of smart hydrogels, researchers are actively developing techniques for precisely programming their structures. This control over structure unlocks the potential to create hydrogels with targeted functionalities, expanding their reach in various fields. 3D printing, also known as additive manufacturing, is a promising technology with the potential to revolutionize this process. It offers precise control over hydrogel properties, enabling the creation of complex structures with intricate geometries, internal channels, and finely tuned mechanical properties. These capabilities are crucial for mimicking the intricate structures and functionalities of natural tissues and organs.

Research by Noor et al. (2019) represents a significant leap forward in the field of 3D printing of hydrogels for biomedical applications, as they were able to create entire, extrusion-based thick cardiac patches with built-in blood vessels as shown in Figure 2.7. This breakthrough showcases the potential of 3D printing to produce personalized tissues that closely resemble native anatomical structures, offering new avenues for effective cardiac repair strategies. The study's novelty lies in the creation of thick, vascularized, and perfusable cardiac patches. This was achieved by utilizing a patient-specific bioink strategy. In this approach, cells were reprogrammed into stem cells, and the extracellular matrix (ECM) was processed to create a personalized hydrogel. Subsequent mixing and differentiation of these components resulted in functional cardiac patches. Notably, these patches were designed to include built-in blood vessels which are open cellular lumens of $\approx 300\text{ }\mu\text{m}$ in diameter that were intended to resemble the patient's native vasculature. This was done in order to facilitate the crucial blood flow that would occur upon implantation. This single-step process was intended to mirror the natural development of blood vessels that have been observed during organoid transplantation. The study successfully demonstrates the feasibility of 3D

printing fully personalized, functional cardiac patches using a patient's own cells and biomaterials. The patches are not only tailored to the patient's unique biochemical and cellular profile but also closely replicate their anatomical features. Furthermore, this approach enables the creation of patches with thicknesses of several millimeters, which is a critical aspect for clinical applications in cardiac repair.

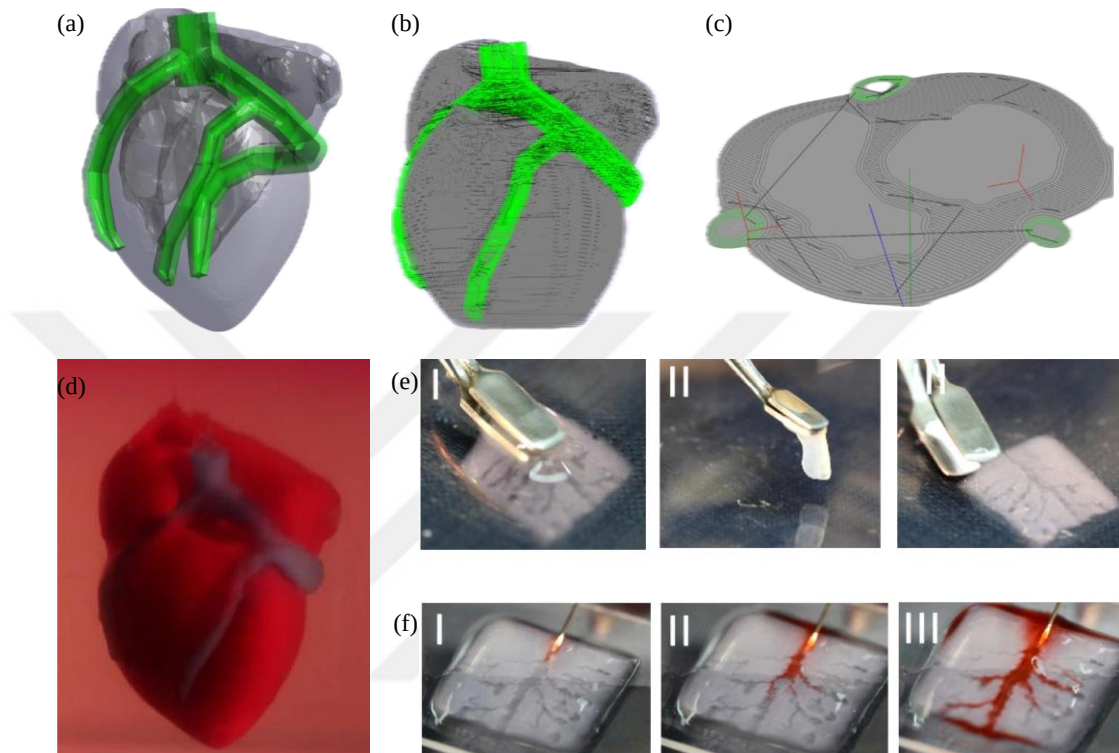


Figure 2.7 : (a) 3D model of the heart, (b) Sliced output of the printer program, (c) a layer/slice of the 3D model, (d) printed heart in the gelatin support medium, (e) robust cardiac patches keeping their integrity, (f) perfusion in the patch (Noor et al., 2019).

Other than the intricate and personalized design opportunity, 3D printing offers a multitude of advantages, particularly in terms of environmental impact. Compared to traditional subtractive manufacturing methods, which remove material to create the desired shape, 3D printing utilizes an additive approach, building the object layer by layer. This significantly reduces waste material, making it a more sustainable production method (Peng, 2016). As illustrated in Figure 2.8 (Chen et al., 2020), the subtractive manufacturing discards excess material, while additive manufacturing builds the object in an efficient manner, minimizing waste eventually and further enhancing its reputation as an environmentally friendly manufacturing technique.

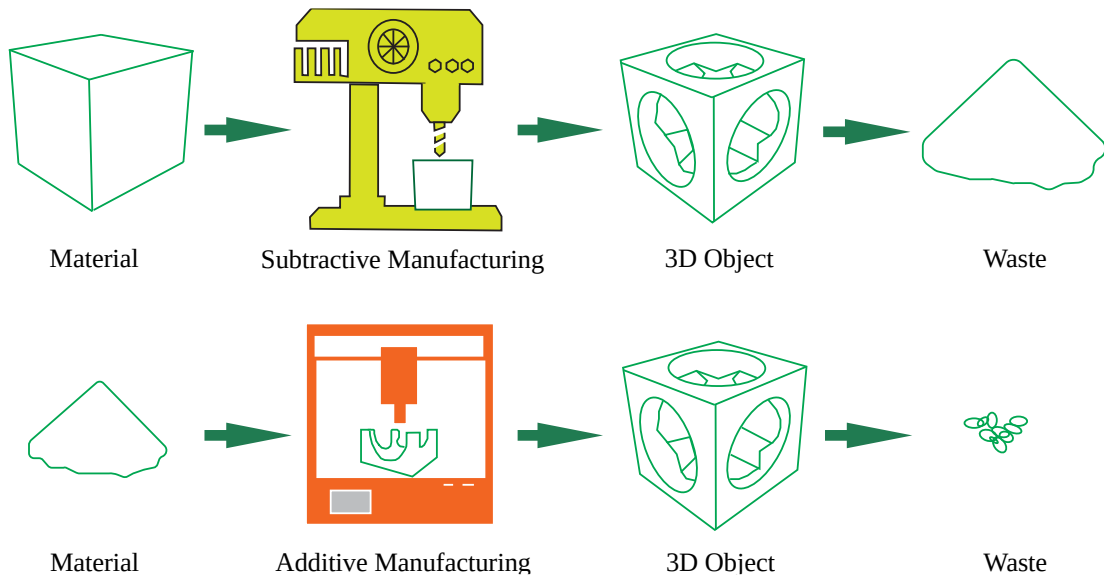


Figure 2.8 : Side-by-side comparison of subtractive manufacturing and additive manufacturing (Chen et al., 2020).

In addition to the advantages of design flexibility and environmental benefits, 3D printing offers economic advantages. In 3D printing, the necessity for complex tooling is also eliminated, thereby reducing costs and production time (Le et al., 2018). Furthermore, the capacity to manufacture components on demand, near the point of use, eliminates the necessity for unnecessary transportation. These factors make 3D printing a compelling choice for low-volume, complex productions. The projected market size for 3D printing products and services is estimated at a staggering 37.2 billion U.S. dollars, with materials alone reaching 3.78 billion U.S. dollars (*Global 3D Printing Industry Market Size*, 2022). This underscores the growing impact of 3D printing.

The transformative potential of 3D printing technology is revolutionizing various sectors, with a particularly profound impact on the medical field. This revolution is facilitated by the exponential growth of patents in tissue engineering (*3D Printing Technology Insight Report*, 2014), a direct result of 3D printing's capabilities. The ability to fabricate personalized drugs and medical devices positions 3D printing as a powerful instrument for medical advancement (Beg et al., 2020). This impact is further underscored by the projected market size of 3D-printed pharmaceuticals, expected to reach a staggering \$522 million by 2030 under a moderate growth scenario (*3D Printed Drugs Market*, 2017).

The field of 3D printing comprises a diverse range of techniques, distinguished by the materials employed (plastics, metals, or resins), the technology used for deposition, the mechanism of layer formation, and the final product's characteristics (surface finish, texture, geometry, and mechanical properties) (Trenfield et al., 2019). The American Society for Testing and Materials (ASTM) has established a classification system that categorizes these techniques into seven distinct groups: binder jetting, directed energy deposition, material extrusion, material jetting, powder bed fusion, sheet lamination, and vat photopolymerization (Jiménez et al., 2019).

For instance, in a material extrusion technique, fused deposition modeling (FDM), thermoplastic polymer filaments undergo a precisely controlled thermal process. FDM utilizes a spool of filament, typically a thermoplastic polymer, which is fed into a heated chamber (liquefier) by a drive mechanism as illustrated in Figure 2.9. This process transforms the filament from a solid state to a semi-molten state, enabling its extrusion through a precisely controlled nozzle (Bakar et al., 2010). The nozzle meticulously deposits the material onto a build platform in a predetermined pattern, layer by layer. A crucial aspect of FDM is the solidification mechanism. As the molten filament exits the nozzle, it fuses with the previously deposited layers. The success of this process heavily relies on the thermoplastic properties of the filament. However, post-extrusion cooling can introduce internal stresses within the printed object. This occurs due to the transition from the filament's glass transition temperature (T_g) to the platform temperature (Ngo et al., 2018). Phase changes during cooling can lead to volumetric shrinkage, which can negatively impact the mechanical integrity of the layers (Ngo et al., 2018). To mitigate this issue, an elevated platform temperature is employed. Ideally, the platform temperature should be close to the T_g of the filament material. This reduces distortion and, in some cases, eliminates it entirely. Consequently, filaments with a lower T_g and a more linear shrinkage rate are generally better suited for FDM printing (Sood et al., 2010). Additionally, for optimal results, it's recommended to maintain the melting process temperature slightly above the material's melting point. Cost-effective and straightforward processing (Ngo et al., 2018), and reusability of printed parts (Maines et al., 2021) are the main benefits of FDM.

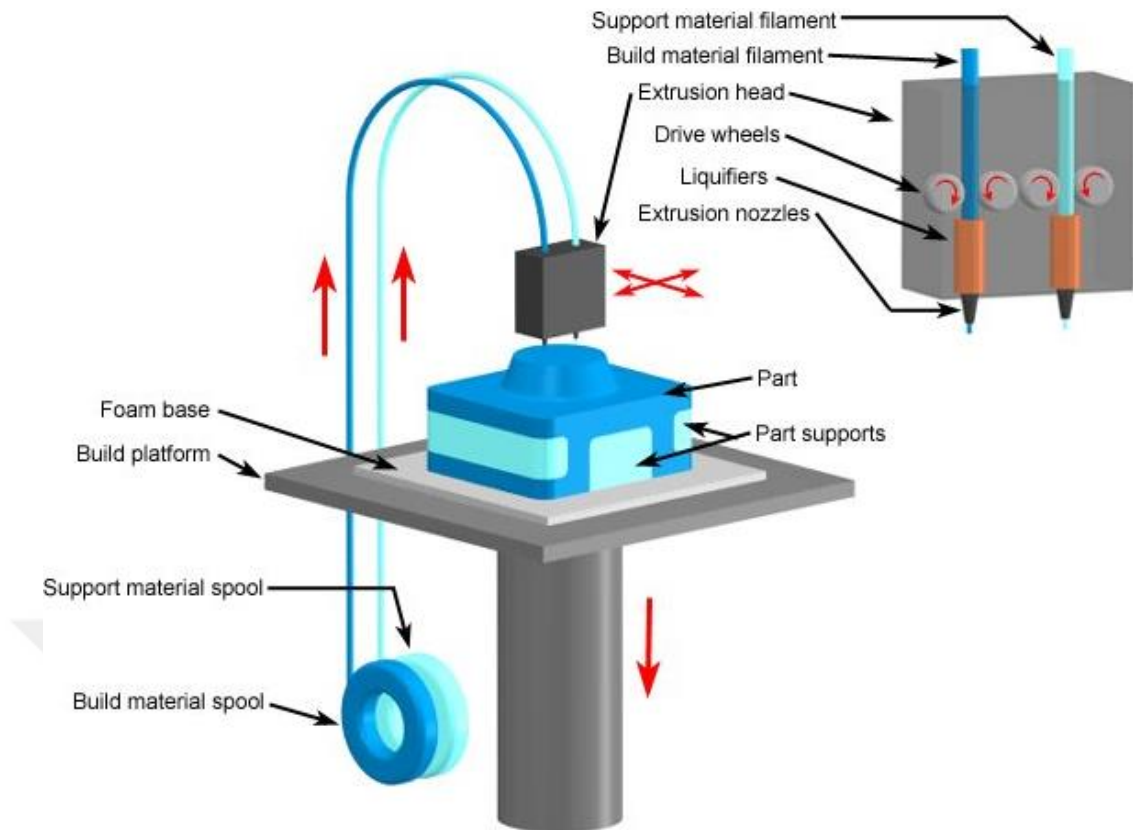


Figure 2.9 : The method of operation in FDM (Dvořák et al., 2018).

As for the technique based on sintering powdered material, powdered bed fusion (PBF) can be given as an example. Powder Bed Fusion (PBF) is a versatile additive manufacturing technology that builds objects layer by layer from a bed of powdered material. PBF's working mechanism is depicted in Figure 2.10. The process unfolds as follows: a roller mechanism meticulously distributes a layer of powder across a build platform that undergoes a precise, incremental rise. Subsequently, a heat source, typically a laser or electron beam can be utilized for irradiation, fusion, and melting of powder particles and then it selectively scans the powder layer, fusing the particles according to a pre-defined three-dimensional computer-aided design (CAD) model (Dev Singh et al., 2021). This initial layer forms the foundation of the object. The build platform then lowers by a predetermined distance corresponding to the desired layer thickness. Fresh powder is deposited on top of the first solidified layer. The heat source then meticulously scans this new layer, selectively fusing the powder particles onto the existing layer, effectively bonding them together. This cycle of powder distribution, heat source scanning, and fusion continues layer by layer with remarkable precision until the entire physical part is complete. PBF technique can be well-suited

for alloys, polymers, metals, ceramics, or a mixture of these materials (Kumar et al., 2023). High precision can be achieved with the PBF method (Velu et al., 2021).

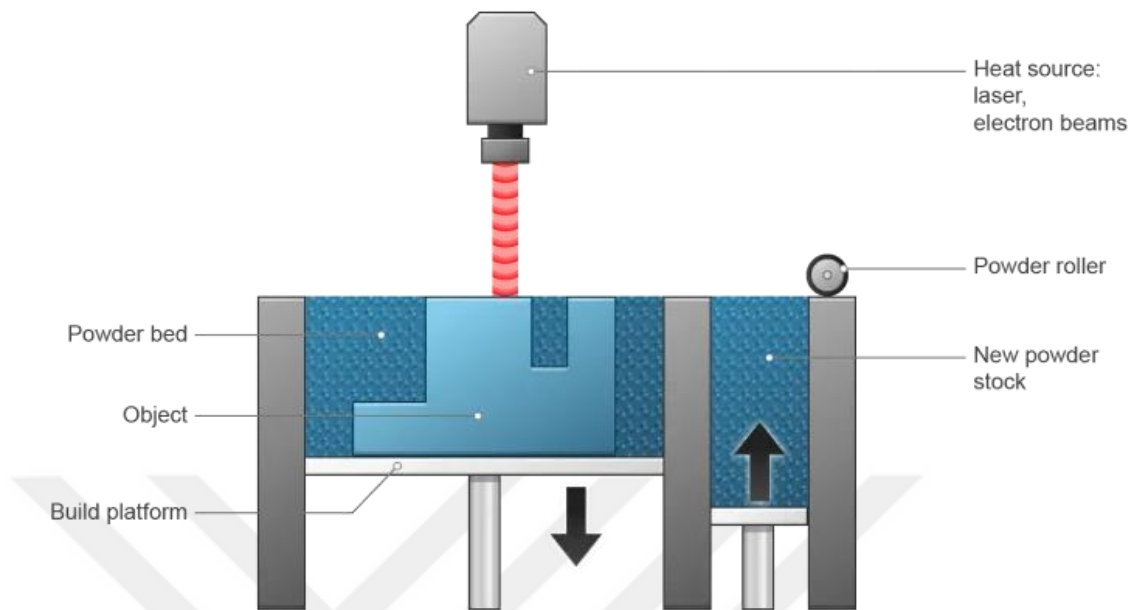


Figure 2.10 : The method of operation in PBF (Kumar et al., 2023).

Stereolithography (SLA) is a liquid-based additive manufacturing method that is known for its exceptional printing resolution and intricate detail. In this process, a vat containing a photocurable resin is subjected to a precisely controlled laser beam under a transparent interface. As illustrated in Figure 2.11 (Lakkala et al., 2023), this laser beam then cures the liquid resin layer by layer, solidifying it into a two-dimensional pattern on the build platform as the building platform moves away. To initiate the curing process, SLA relies on a photoinitiator or photoinitiator system, which absorbs light and triggers a chemical reaction that promotes chain growth (Bagheri & Jin, 2019). The most commonly utilized SLA resins are based on acrylic and methacrylic compounds (Taormina et al., 2018). These resins undergo free-radical photopolymerization. The application of light initiates the breaking of weak bonds within the resin, thereby generating free radicals. Subsequently, the free radicals facilitate the formation of covalent bonds between the resin molecules, thereby conferring the material with its solid plastic state (Taormina et al., 2018). The properties of these resins, including the number of functional groups and molecular weight, significantly influence the behavior of the resin during printing (Taormina et al., 2018).

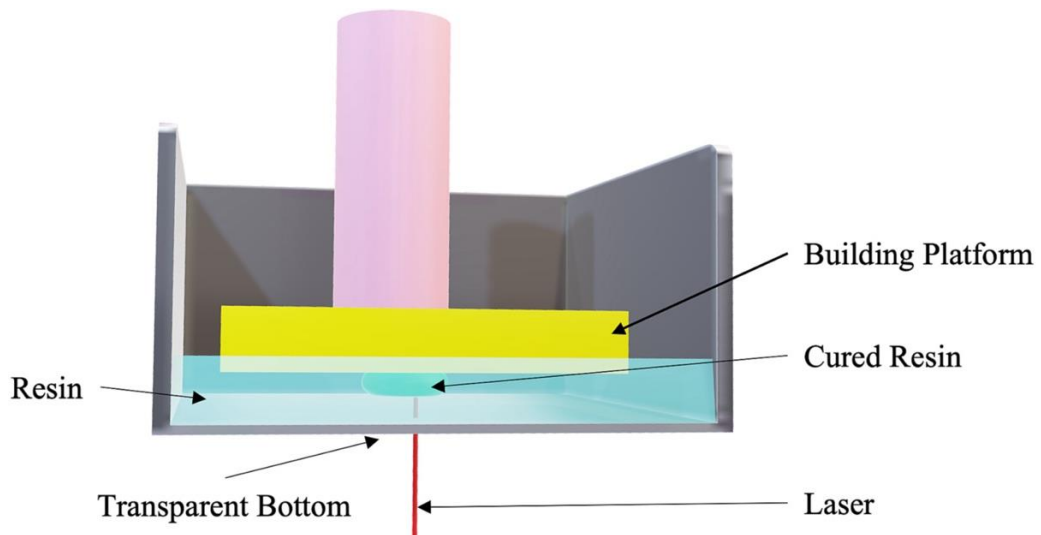


Figure 2.11 : Printing mechanism of an SLA printer (Lakkala et al., 2023).

While SLA is highly adept at producing intricate components with resolutions as fine as 10 μm , it does have certain limitations. The availability of raw materials may be constrained in comparison to other methods (Wang et al., 2017), and the printing process may be slower due to the photopolymerization rates (Manapat et al., 2017). Additionally, prolonged exposure to UV light during printing may cause degradation reactions, potentially impacting biological applications (Manapat et al., 2017).

In order to minimize the damage to cells during printing applications, Klak et al. investigated the effects of ultraviolet (UV) light with wavelengths of 365 nm and 405 nm on DNA in swine pancreatic islets which are clusters of cells. The results demonstrated a notable difference in the extent of damage. The level of DNA damage in islets exposed to the 405 nm wavelength range from 2% to 6%, with the degree of damage increasing with longer exposure times (10 to 300 seconds) (Klak et al., 2020). Conversely, islets treated with the 365 nm wavelength exhibited considerably higher levels of damage, reaching up to 50% (Klak et al., 2020). These findings serve to illustrate the superiority of the 405 nm wavelength in minimizing cellular damage.

2.3 4D Printing

The concept of 4D printing was first proposed in 2013 by Professor Tibbits at MIT's Self-Assembly Lab (Tibbits, 2014). This builds upon the concept of 3D printing by introducing a novel element: the capacity of printed objects to undergo transformation or self-assembly over time. This innovation essentially incorporates time as a fourth dimension, complementing the familiar spatial dimensions of length, width, and depth

(Trenfield et al., 2019). In essence, 4D-printed objects can undergo dynamic alterations in their form, properties, or functionality in response to external stimuli (Trenfield et al., 2019). Such stimuli can result in a variety of large-scale material changes, including swelling, stretching, bending, twisting, or rolling, depending on the object's design (Falahati et al., 2020). 4D printing mainly distinguishes itself from 3D printing through the incorporation of smart materials, as illustrated in Figure 2.12.

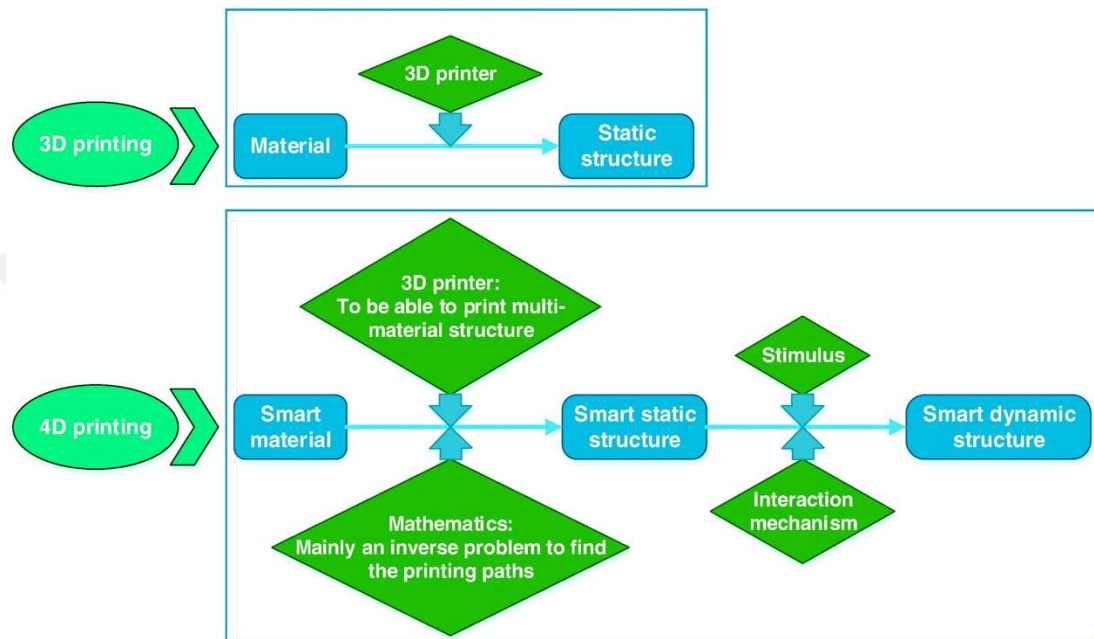


Figure 2.12 : Contrasting 3D and 4D printing workflows (Momeni et al., 2017).

The concept of 4D printing is perfectly aligned with the properties of smart hydrogels, which exhibit shape-shifting upon encountering external stimuli. The considerable potential of 4D printing has prompted a significant body of research into integrating smart, responsive materials with 3D printing technology. This introduces a time-based element to their functionality, offering a new dimension to their utility. This opens up a vast array of potential applications, particularly in the field of biomedicine (Malekmohammadi et al., 2021). In this context, factors such as biocompatibility, response temperature, and the rate of shape change become crucial considerations (Malekmohammadi et al., 2021).

Hydrogels offer a compelling combination of properties that make them ideal especially for biomedical applications. These readily synthesized materials boast excellent biocompatibility, tunable characteristics, exceptional alignment, and affordability (Malekmohammadi et al., 2021). As a result, hydrogels hold significant promise as interface materials across various biomedical fields, from non-invasive

diagnostics and implant therapies to cell manipulation and implantable devices (Peppas et al., 2006).

To create a mechanically robust smart hydrogel with self-healing, shape-memory functions, and toughness, this research focuses on the development of an acrylate-based hydrogel bearing both hydrophilic and hydrophobic moieties. For this purpose, hydrophilic *N, N*-dimethyl acrylamide (DMAA), and methacrylic acid (MAA) monomers were incorporated with hydrophobic hexadecyl acrylate (C16A) monomer. This aforementioned synergistic effect of hydrophobic interactions and hydrogen bonding was utilized to reinforce the gels mechanically. The hydrophobic domains formed by C16A determine the actuation temperature of the shape-memory function and the extent of self-healing in the hydrogel, thus resulting in a temperature-responsive and self-healable hydrogel in the absence of any chemical crosslinker and solvent. Moreover, the actuation temperature was successfully arranged with the content of C16A, and body temperature-responsiveness was achieved. The hydrogel's self-healing, shape-memory, and temperature-responsive properties offer promising applications in targeted drug delivery and tissue engineering. In drug delivery, the shape change can trigger drug release, while in tissue engineering, the self-healing ability can mimic the dynamic nature of natural tissues.



3. EXPERIMENTAL PART

3.1 Chemicals

3.1.1 Monomers

Hexadecyl acrylate (C16A, TCI, 296.50 g/mol), *N,N*-dimethyl acrylamide (DMAA, Sigma-Aldrich, 99.13 g/mol), and methacrylic acid (MAA, Merck, 86.09 g/mol) monomers and diphenyl (2,4,6-trimethylbenzoyl) phosphine oxide (TPO, Sigma-Aldrich, 348.37 g/mol) photoinitiator were used as received. The chemical structures of the monomers are given in Figure 3.1.

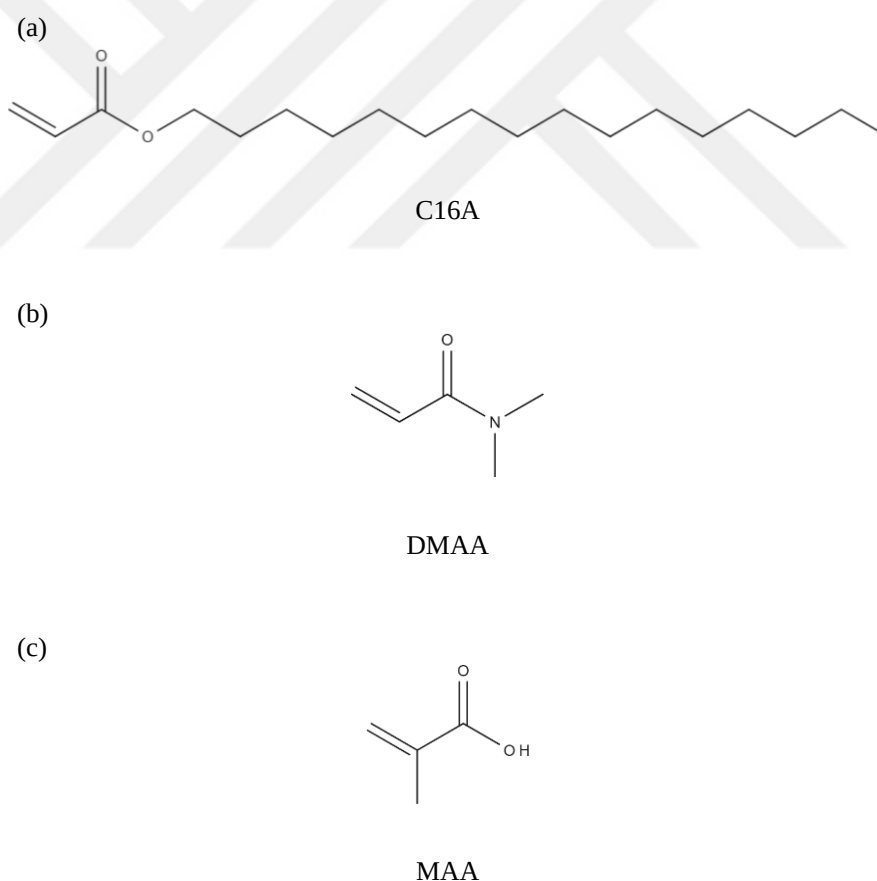


Figure 3.1 : The chemical structures of monomers: (a) hexadecyl acrylate (C16A), (b) *N,N*-dimethyl acrylamide (DMAA), (c) methacrylic acid (MAA).

3.1.2 Photoinitiator system

The polymerization of the monomers was carried out by using diphenyl (2,4,6-trimethylbenzoyl) phosphine oxide (TPO, Sigma-Aldrich, 348.37 g/mol) as a photoinitiator. The molecular structure of TPO is given in Figure 3.2.

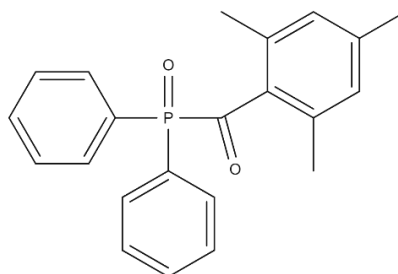


Figure 3.2 : The chemical structures of photoinitiator diphenyl (2,4,6-trimethylbenzoyl) phosphine oxide (TPO).

3.2 Tools and Instruments

3.2.1 Resin printer

This study employed a commercially available stereolithography (SLA) resin printer, the Creality Halot-One, which utilizes a 405 nanometer (nm) wavelength light source for vat-photopolymerization of the monomers. Due to the combination of a commercial printer and a non-commercial resin formulation, it was necessary to optimize the printing parameters before proceeding with the hydrogel printing process. Printing parameters can be listed as bottom exposure time, exposure time, motor speed, and light-off delay.

The bottom exposure time is the curing time for the initial layers printed directly on the steel platform. It differs from the standard exposure time because the resin interacts differently with the platform compared to other layers. A longer bottom exposure time creates a stronger base for the print. Exposure time, on the other hand, is the standard curing time for all layers except the bottom layers. It is also necessary to balance the motor speed. A speed that is too fast may result in breakage, whereas a speed that is too slow may not be sufficient to effectively peel the printed item from the resin vat. The light-off delay is the brief interval that permits the resin to settle prior to being cured by ultraviolet light, thereby facilitating the precision of the resulting prints.

In order to achieve optimal printing results for the hydrogel monomers, the aforementioned parameters were systematically adjusted during the printing of a

dummy model. A comparison of the printed dummy model's details subsequently identified the optimal parameters. The optimal parameters were determined to be a bottom exposure time of 70 seconds, an exposure time of 10 seconds, a motor speed of 3 millimeters per second, and a light-off delay of 8 seconds. This approach ensured that the printed hydrogel structure was sufficiently cured. The key determinant of printability was identified as exposure time. Figure 3.3 displays multiple examples of printed models arranged in increasing exposure times from left to right (7, 8, 9, and 10 seconds). A distinct enhancement in print quality is apparent upon comparison, particularly noticeable in the middle section of the printed models, indicating a gradual improvement in printability.

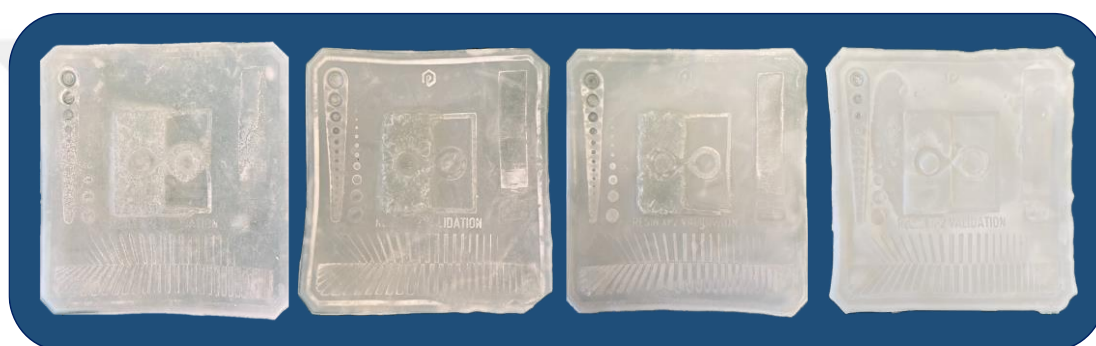


Figure 3.3 : Validation of printing parameters through printed dummy models. From left to right: printed dummy models with exposure times of a 7, 8, 9, and 10 seconds.

3.2.2 Analytical balance

The weight of the C16A monomer, TPO, and synthesized hydrogels were recorded with Precisa 205A analytical balance.

3.2.3 Digital compass

The length of the hydrogel samples was recorded by using a Mitutoyo Absolute Digimatic digital compass with a range from 0 to 150 millimeters (mm).

3.2.4 Vacuum oven

The hydrogels were dried in Nucleon NVE Series Vacuum Gauge at 30 °C under vacuum.

3.2.5 Differential scanning calorimetry

The thermal characteristics of hydrogels were evaluated using Perkin-Elmer Diamond DSC 4000 differential scanning calorimetry (DSC). During testing, nitrogen gas at a

flow rate of $19.8 \text{ mL}\cdot\text{min}^{-1}$ was circulated through the cell to prevent sample interaction with the environment.

3.2.6 Rheometer

Rheological tests were carried out using a Bohlin Gemini 150 rheometer (Bohlin Instruments, UK) equipped with a Peltier temperature control unit. A parallel plate (diameter= 20 mm) was employed during the tests. A water trap was also used to prevent solvent vaporization at elevated temperatures. Temperature-sweep oscillation tests were carried out to assess the rheological behavior of the printed hydrogels in response to temperature.

3.2.7 Light microscope

Olympus CX31 light microscope equipped with QImaging MicroPublisher 3.3 RTV image analyzing system was used to monitor the surface of the hydrogels.

3.2.8 Mechanical test device

The mechanical properties of hydrogels were examined with a Zwick-Roell mechanical test device. The device was equipped with a 500 N load cell and used to perform unidirectional tensile tests on dog-bone-shaped specimens printed in compliance with the ISO 527 standard.

3.2.9 Fourier-transform infrared spectroscopy (FTIR)

The structure analysis of hydrogels was accomplished with Agilent Technologies Cary 630 FTIR spectrometer equipped with ATR. The samples were placed on top of an ATR accessory, and the spectra were recorded in the wavenumber range of $4000\text{--}400 \text{ cm}^{-1}$.

3.2.10 X-ray powder diffraction (XRD)

The crystallinity of hydrogels was examined using Bruker D8 Advance X-ray powder diffraction (XRD) instrument.

3.3 Experimental Procedure

Without any pre-treatment, UV polymerization of the monomers with photoinitiator TPO was conducted at $23 \pm 2 \text{ }^{\circ}\text{C}$ in the absence of a solvent and a chemical crosslinker.

Since the hydrogels were comprised of both hydrophilic constituents such as DMAA and MAA, and hydrophobic constituent C16A monomer, the effect of each group of monomers on hydrogel properties was evaluated by designing 2 sets of gelation experiments. Thus, there are primarily two experimental parameters, namely the mole fraction of DMAA in the DMAA-MAA mixture (x_{DMAA}), and the mole percentage of C16A in the overall precursor mixture (C16A mol%) which are calculated by,

$$x_{DMAA} = \frac{\text{mole of DMAA}}{\text{moles of DMAA} + \text{MAA}} \quad (3.1)$$

$$C16A \text{ mol}\% = \frac{\text{mole of C16A}}{\text{moles of C16A} + \text{DMAA} + \text{MAA}} \times 100. \quad (3.2)$$

In order to determine the optimal properties, the first set of experiments was designed using various mole ratios of DMAA to MAA at a constant C16A amount of 40 mol % as indicated in Table 3.1.

Table 3.1 : Synthesis conditions of hydrogels at various DMAA to MAA mole ratios. C16A content was fixed at 40 mol %.

x_{DMAA}	DMAA:MAA Mol Ratio	DMAA (mol%)	MAA (mol%)
0	0:1	0	60
0.1	1:9	6	54
0.2	1:4	12	48
0.33	1:2	20	40
0.4	2:3	24	36
0.5	1:1	30	30
0.67	2:1	40	20
0.8	4:1	48	12
0.9	9:1	54	6
1	1:0	60	0

In the second set of experiments, the amount of C16A was varied from 10 to 70 wt% at a fixed DMAA mole fraction ($x_{DMAA} = 0.5$), as indicated in Table 3.2.

Table 3.2: Synthesis conditions of hydrogels at various C16A contents. The mole ratio of DMAA to MAA was fixed at 0.5.

C16A (mol%)	DMAA:MAA Mol Ratio	DMAA (mol%)	MAA (mol%)
10	1:1	45	45
20	1:1	40	40
30	1:1	35	35
40	1:1	30	30
50	1:1	25	25
60	1:1	20	20
70	1:1	15	15

Typically, the monomer mixtures with various compositions were prepared by first melting the C16A monomer at 35 °C on a magnetic stirrer, and then mixing it with the liquid DMAA and MAA monomers. After the addition of 2 wt. % TPO (compared to the total monomers) and the mixture being cooled to room temperature (23 ± 2 °C), the precursor mixtures were poured into the vat section and underwent bulk polymerization through an SLA printer with a layer thickness of 50 μm . The bottom exposure time and overall exposure time parameters of the printer were set at 70 seconds and 10 seconds, respectively. After printing each layer, the printing program allowed the printed part to cool down for 5 seconds before printing the following layer. Once the printing process was finished, the cured hydrogel samples were delicately removed from the printer's platform using a steel spatula. Following the printing process, the samples were treated with 405 nm laser light for 10 minutes and then submerged in a large volume of water for two days to promote physical cross-linking. Besides printing, syringes were also used as templates to enable rapid and same-sized hydrogel preparation. After preparing the precursor, syringes were simply loaded with the liquid and placed into the printer's vat to be cured for five minutes using the printer's "Clean" function. The Clean function is commonly used to cure any residual resin within the vat, yet in our study, it also served as a UV lamp to cure samples in syringes. Four repetitions of the Clean function were performed on the syringes, with alternating rotations to ensure full polymerization. After the curing process, the syringes containing hydrogel samples were stored in a refrigerated environment at 2 °C overnight to remove the samples easily from the syringes. The assessment of swelling

and rheological properties was conducted on samples acquired from 1 mL and 20 mL syringes, respectively.

3.4 Hydrogel Characterization

The synthesized hydrogel samples were immersed in an excess amount of distilled water for 2 days to promote physical cross-linking, namely hydrogen bonds, prior to any characterization process.

3.4.1 Swelling measurements

Swelling measurements of hydrogels were conducted in water using the same procedures. All measurements of swelling were conducted using a minimum of three samples, and the standard deviation was calculated accordingly.

The as-prepared mass of the samples was recorded as m_0 and measured using an analytical balance before immersing in water. After immersing samples in excess water, the masses of the specimens were collected daily until they reached equilibrium, which took 2 days. After reaching equilibrium, the mass of the swollen samples was recorded as m_{swl} . Afterwards, samples were dried in an oven at 30 °C for 2 weeks under vacuum. Lastly, the mass of the dried samples was recorded as m_{dry} . The swelling properties of the samples were determined by calculating the weight swelling ratio (q_w), and the relative weight swelling ratio (m_{rel}), which are the swelling ratios with respect to their dry and as-prepared states, respectively. Additionally, the gel fraction (W_g) of samples was calculated which indicates the mass fraction of the incorporated monomers into the 3D network of the hydrogels. Equations (3.3), (3.4), and (3.5) present the mathematical formulas to calculate q_w , m_{rel} , and W_g respectively.

$$q_w = \frac{m_{swl}}{m_{dry}} \quad (3.3)$$

$$m_{rel} = \frac{m_{swl}}{m_0} \quad (3.4)$$

$$W_g = m_{dry} / (m_0 \times C_0) \quad (3.5)$$

Where C_0 is the total concentration of the monomers in the hydrogel.

3.4.2 (ATR)-FTIR measurements

ATR-FTIR analysis was utilized to confirm the functional groups present in the hydrogel networks. The measurements were conducted on five solid polymer specimens obtained after drying of the hydrogels, namely poly(C16A-co-DMAA-co-MAA) terpolymer, poly (DMAA-co-MAA) copolymer, poly (C16A), poly (DMAA), and poly (MAA) homopolymers.

3.4.3 Differential scanning calorimetry (DSC) measurements

DSC analyses were carried out on hydrogels at their equilibrium states in water. The hydrogel specimens that were cut into pieces of around 10 mg were loaded into an aluminum crucible of the DSC instrument and sealed with a sample encapsulation press. The heating-cooling cycles are conducted at a controlled heating rate of 5 °C min⁻¹ ranging from -20 °C to 80 °C. Melting (T_m) and crystallization (T_{cry}) temperatures were determined from endothermic and exothermic peaks in the DSC curves, respectively. Melting enthalpies of C16A ($\Delta H_{m,C16A}$) were calculated from the area under the melting peak while the fraction of crystalline domains (f_{cry}) caused by C16A was calculated using the equation,

$$f_{cry} = \frac{q_w \times w_{C16A} \times \Delta H_{m,C16A}}{\Delta H_{C16A}^0} \quad (3.6)$$

where w_{C16A} , $\Delta H_{m,C16A}$, and $\Delta H_{m,C16A}^0$ are the weight fraction of C16A, the melting enthalpy of specimen (in J.g⁻¹) and the melting enthalpy of 100% crystalline C16A (149 J.g⁻¹), respectively.

3.4.4 X-ray powder diffraction (XRD) measurements

X-ray powder diffraction analysis of hydrogels was conducted to further characterize the crystalline structures of the samples. For this purpose, swollen hydrogel samples were first dried at 30 °C for 2 weeks. They were then ground into powder form using a mortar and pestle. Eventually, powdered samples were put into XRD examination. Scanning of the samples was carried out between $2\theta = 0$ and 45°. XRD analysis was carried out on eight specimens, namely terpolymers with various C16A content between 10 to 70 mol%, and additionally poly(DMAA-co-MAA) copolymer as a control group.

3.4.5 Rheological measurements

Disc-shaped specimens, measuring 20 mm in diameter and 1 mm in thickness, were securely placed between the parallel plates of the instrument to assess the temperature-dependent viscoelastic characteristics of the specimens. The rheometer was then programmed to conduct a cooling-heating cycle in between 65 °C – 25 °C at a fixed angular frequency (ω) of 6.3 rad/s and strain amplitude (γ_o) of 0.1%.

3.4.6 Mechanical tests

Uniaxial tensile testing on a Zwick Roell Z0.5 TH universal testing equipment with a 500 N load cell was used to assess the mechanical characteristics of the printed hydrogels. The specimen shape selected was a dog-bone shape, in accordance with the International Organization for Standardization (ISO 527) standard. Each test was carried out at $23 \pm 2^\circ\text{C}$, and all specimens were water-swollen hydrogels. The nominal stress (σ_{nom}) and strain (ε) of the printed hydrogels were recorded with Zwick/Roell equipment's testXpert III Testing Software. Stress can be described as the force applied on the specimen's initial cross-sectional area (σ_{nom}) or deformed cross-sectional area (σ_{true}), respectively. Additionally, strain is the change in the length of the material as a response to applied stress, whether it is the stretch ratio (λ) or true strain (ε). Collected data was then plotted into $\sigma_{nom} - \varepsilon$ and $\sigma_{nom} - \lambda$ graphs and used to establish the mechanical properties such as Young's modulus (E), fracture stress (σ_f), fracture strain (ε_f), toughness (W).

E , σ_f , ε_f values were calculated from $\sigma_{nom} - \varepsilon$ plot, whereas the W was computed from $\sigma_{nom} - \lambda$ plot. The slope of the linear component of $\sigma_{nom} - \varepsilon$ plot between 2 and 6% strain was chosen to determine E . σ_f is the stress value at the point of rupture, whereas ε_f is the response of the material to σ_f . W was computed as the area under the $\sigma_{nom} - \lambda$ plot until the point of failure. The relation of this σ_{nom} , σ_{true} , λ , and ε definitions for elongation is calculated in the equation (3.7) and (3.8).

$$\varepsilon = \lambda - 1 \quad (3.7)$$

$$\sigma_{true} = \lambda \times \sigma_{nom} \quad (3.8)$$

During uniaxial tensile tests, the sample length was set as 22 mm, and strain rate parameter was fixed at $25.0 \text{ mm} \times \text{min}^{-1}$ ($1.89 \times 10^{-2} \text{ s}^{-1}$). Stretching tests at different

strain rates were also carried out. The results were then repeated for at least three times for each sample and standard deviations were calculated accordingly.

3.4.7 Determination of shape-memory behavior

The shape memory property of hydrogels was evaluated using a bending test. The tests were conducted using rectangular-shaped (40 mm × 12 mm × 1 mm) hydrogel specimens with 40, 50, 60, and 70 mol % C16A and $x_{\text{DMAA}} = 0.5$ under the same conditions. Specimens were heated to 65 °C in a large container. At 65 °C, the rectangular sample was bent to $\theta_0 = 180^\circ$ for a minute at room temperature to maintain a temporary form. Bent rectangular samples were then gradually heated from 25 °C to 65 °C, increasing one degree at a time. Meanwhile, after waiting five minutes at each temperature, the angle between the two edges was recorded as the recovery angle θ_r . Angular shape-recovery efficiency $R_r(\theta)$ and angular shape-fixity ratio $R_f(\theta)$ were calculated accordingly. $R_r(\theta)$ is defined as the ability of the material to return to its original shape while $R_f(\theta)$ is the indicator of the fixing ability of the switching domains of material. $R_r(\theta)$ and $R_f(\theta)$ can be calculated from the equations (3.7) and (3.8), where θ_d , θ_r and θ_f are deformed angle under load, recovery angle and fixed angle after unloading, respectively.

$$R_r(\theta) = \frac{\theta_d - \theta_r}{\theta_d - 180^\circ} \quad (3.9)$$

$$R_f(\theta) = \frac{\theta_f - 180^\circ}{\theta_d - 180^\circ} \quad (3.10)$$

4. RESULTS AND DISCUSSION

The objective of this study was to investigate the properties of hydrogels composed of C16A, DMAA, and MAA monomers. It is crucial to understand the conformational structure of polymers which significantly influences their properties. ATR-FTIR measurements were conducted to observe the conformational structures of polymers composed of various combinations of monomers.

Homopolymers poly(DMAA), poly(MAA), and poly(C16A), a copolymer of poly(DMAA-co-MAA), and poly(C16A-co-DMAA-co-MAA) with x_{DMAA} of 0.5 and 50 mol% C16A were used as specimens. As expected from the polymer compositions, the IR spectra (Figure 4.1) reveal distinct peaks corresponding to the functional groups present in each monomer unit.

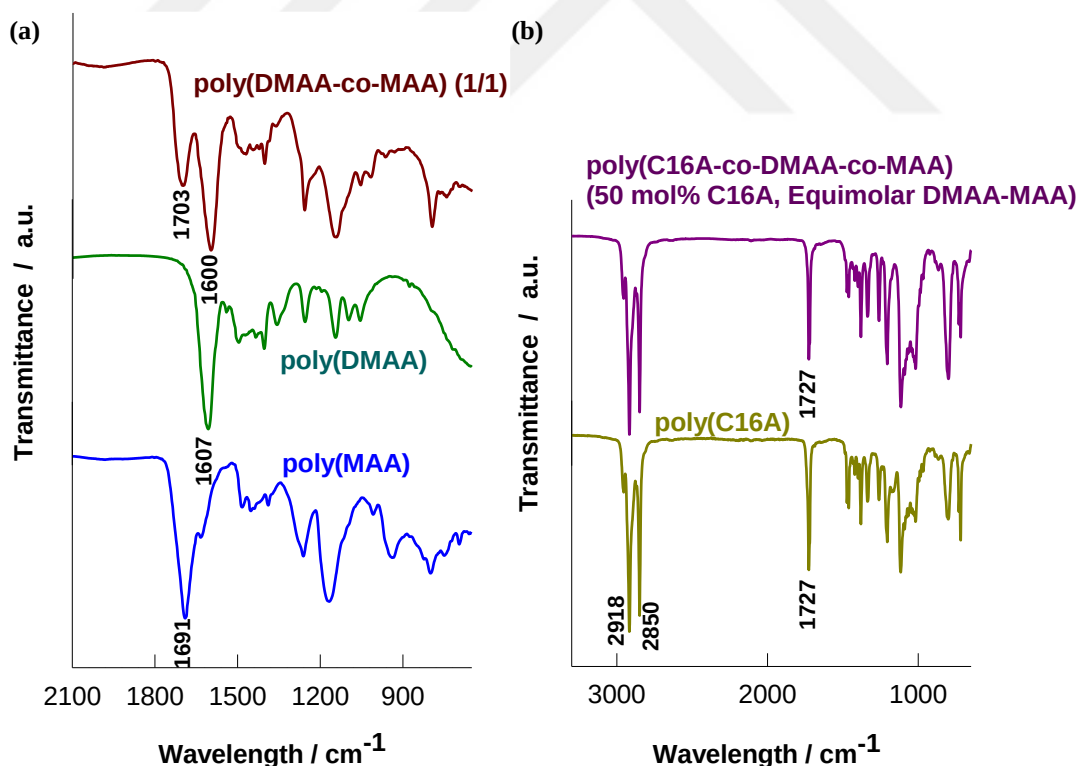


Figure 4.1 : (a) IR spectra of poly(DMAA-co-MAA), poly(DMAA), and poly(MAA), (b) IR spectra of poly(C16A-co-DMAA-co-MAA).

In Figure 4.1(a), IR spectra of copolymer and homopolymers of DMAA and MAA were given. As seen from Figure 4.1(a), C=O stretching of MAA units appears at 1703

and 1691 cm^{-1} for copolymer and homopolymer, respectively (Rufino & Monteiro, 2003). Also, the C=C absorption peak for the DMAA units appears at 1600 and 1607 cm^{-1} for copolymer and homopolymer, respectively (Kodaira et al., 1988).

Continuing with the analysis of the C16A monomer, Figure 4.1(b) presents the IR spectra of the terpolymer and the C16A homopolymer. As expected from the presence of C16A units, both spectra exhibit strong bands at $2850\text{--}2918\text{ cm}^{-1}$, corresponding to the C-H stretching of the side-chain alkyl groups (Mu et al., 2022). Both spectra also reveal a band at 1727 cm^{-1} , likely due to the C=O stretching vibration (Cao et al., 2017).

4.1 x_{DMAA} Effect

The initial experiment investigated the influence of the hydrophilic monomer ratio (DMAA/MAA) on the properties of the hydrogels. While the hydrophobic monomer, C16A, was maintained at a constant 40 mol%, the DMAA/MAA ratio (denoted as x_{DMAA}) was systematically varied from 0 to 1 (0, 0.1, 0.2, 0.33, 0.4, 0.5, 0.67, 0.8, 0.9, 1). These hydrogels were then subjected to differential scanning calorimetry (DSC), swelling ratio, and mechanical testing analyses to understand the impact of x_{DMAA} on their thermal and mechanical properties.

The DSC results revealed that the melting temperature T_m , crystallization temperature T_{cry} , and the fraction of crystallinity f_{cry} remained relatively constant (around $35 \pm 1\text{ }^{\circ}\text{C}$, $26 \pm 2\text{ }^{\circ}\text{C}$, and 0.27 ± 0.18 , respectively) across all the hydrogel samples. This observation can be attributed to the fixed content of the hydrophobic C16A monomer, which likely dominates the thermal behavior.

In contrast, relative weight swelling ratio m_{rel} and swelling ratio q_w exhibited a clear dependence on the x_{DMAA} . As shown in Figure 4.2 (a), both m_{rel} and q_w increased steadily with increasing x_{DMAA} . This trend reflects the enhanced hydrophilicity of the hydrogels with a higher proportion of DMAA units (Oral et al., 2020; Sadik et al., 2019).

Verifying the extent of polymerization is crucial for understanding hydrogel network integrity. Therefore, a bar graph showing the gel fraction of the hydrogels is plotted in Figure 4.2(b). The results indicate that the gel fraction W_g of the hydrogels was unity

(1.003 ± 0.001), which means that the monomer units were completely polymerized and incorporated into the three-dimensional network of the hydrogel.

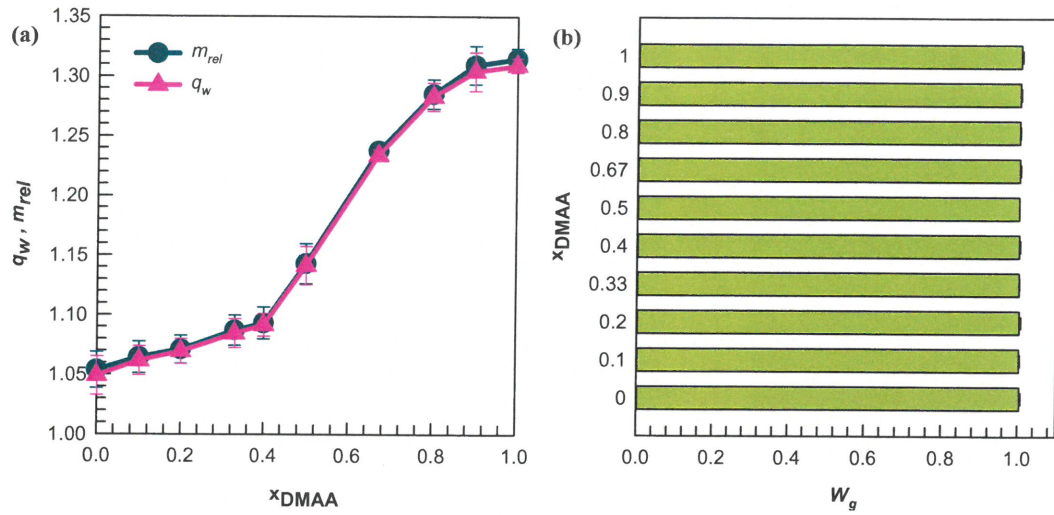


Figure 4.2 : Effect of x_{DMAA} on hydrogel properties. (a) Changes in relative weight swelling ratio m_{rel} and swelling ratio q_w with increasing x_{DMAA} . (b) Gel fraction W_g of the hydrogels at different x_{DMAA} values.

The characterization of the hydrogels, which highlights the influence of the hydrophilic monomer ratio (x_{DMAA}) on swelling behavior and network integrity, requires further investigation into its impact on mechanical performance. The influence of the hydrogel composition on the mechanical performance of hydrogels having varying x_{DMAA} was analyzed with tensile tests at a fixed strain rate of $1.89 \times 10^{-2} \text{ s}^{-1}$. As illustrated in Figure 4.3(a), the stress-strain curves were plotted for 10 samples with different x_{DMAA} . While Figure 4.3(a) shows the stress-strain curves, (b) and (c) represent the change in mechanical parameters as two pairs of $E - \sigma_f$ and $W - \varepsilon_f$, respectively. These graphs provide insights into the mechanical behavior of the hydrogels.

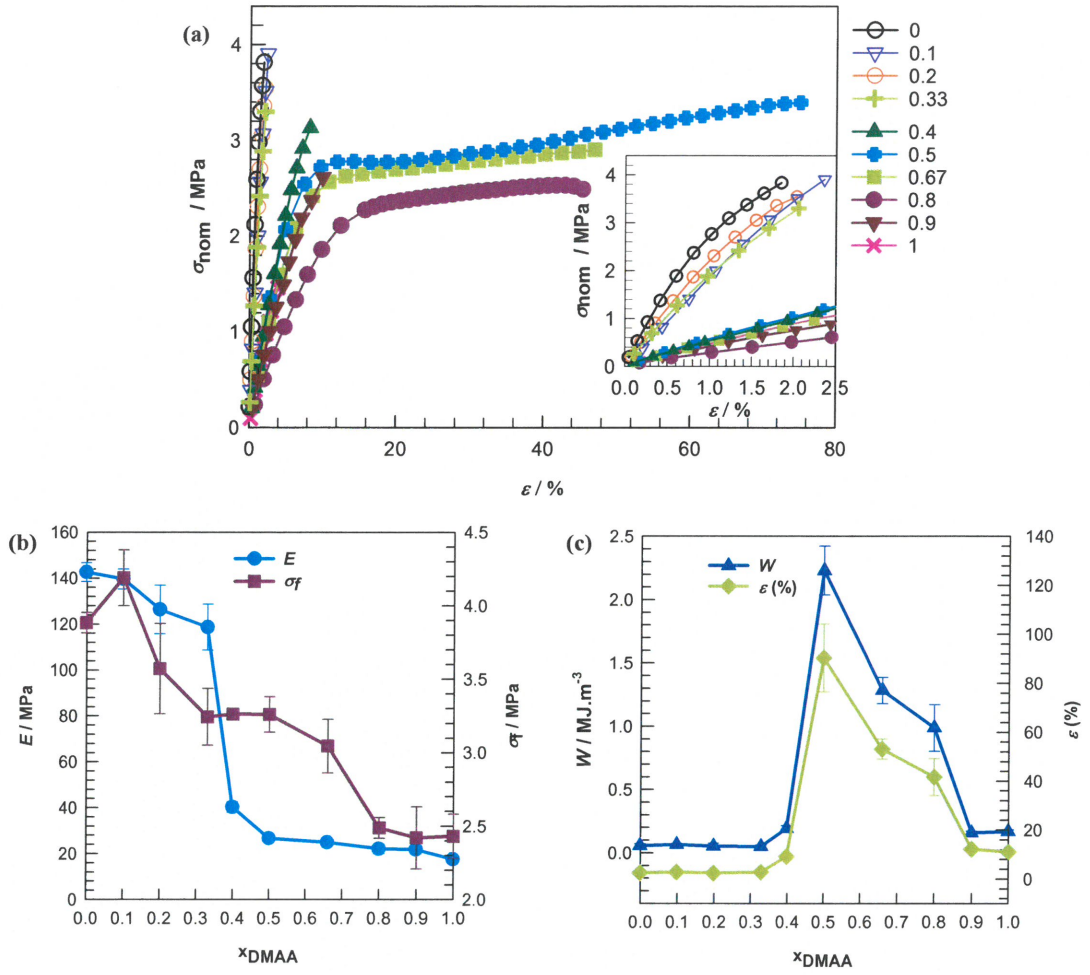


Figure 4.3 : (a) Stress-strain curves of hydrogels with various x_{DMAA} , (b) Young's modulus E , fracture stress σ_f , (c) toughness W and fracture strain ϵ_f of hydrogels with various x_{DMAA} .

The variation in x_{DMAA} represents, significantly influenced the mechanical behavior of the hydrogels as shown in Figure 4.3(a). Samples with x_{DMAA} of 0, 0.1, 0.2, 0.33, 0.4, 0.9, and 1 exhibited brittle characteristics, apparent from their stress-strain curves. In contrast, hydrogels with x_{DMAA} of 0.5, 0.6, and 0.8 displayed ductile behavior, with the sample at x_{DMAA} of 0.5 demonstrating the most pronounced ductility with a 90.3 ± 14 % stretching ratio and 26.7 ± 1.1 MPa Young's modulus E . This brittle-to-ductile behavior.

Figure 4.3(b) further strengthens these observations. Young's modulus E exhibits a remarkable increase, reaching a maximum value of 142.7 MPa at $x_{DMAA}=0$. This represents an 8-fold increase compared to the value obtained at $x_{DMAA}=1$ (17.6 MPa). Interestingly, both toughness W and fracture strain ϵ_f reach their peaks at $x_{DMAA}=0.5$, with values of 2.23 ± 0.2 MJ.m⁻³ and 90.3 ± 14 %, respectively. These trends suggest

that an equimolar ratio of MAA and DMAA ($x_{\text{DMAA}} = 0.5$) leads to the overall highest toughness. This can be attributed to the presence of equal amounts of hydrogen bond donor (MAA) and acceptor (DMAA) units within the polymer network. Facilitated by this complementary pairing, hydrogen bonding likely creates sacrificial bonds, ultimately enhancing the toughness of the hydrogel at this specific composition. As the best overall mechanical properties were obtained with the equimolar hydrophilic monomers, a constant x_{DMAA} of 0.5 was maintained while the C16A mol% was varied between 10 and 70 in the second part of the study.

4.2 C16A Effect

Building upon the influence of the hydrophilic monomer ratio (x_{DMAA}) on hydrogel properties, the effect of the hydrophobic monomer, C16A, was investigated. Swelling measurements were conducted to assess the water uptake behavior of hydrogels with varying C16A mol%. Figure 4.4 illustrates that the parameters of relative weight swelling ratio m_{rel} and weight swelling ratio q_w tend to decrease with increasing C16A mol%. This can be attributed to the reduced available volume of water between polymeric chains at higher C16A concentrations. A gel fraction W_g of unity is also seen for the second set of hydrogels, indicating that all monomers have been successfully incorporated into the polymeric network.

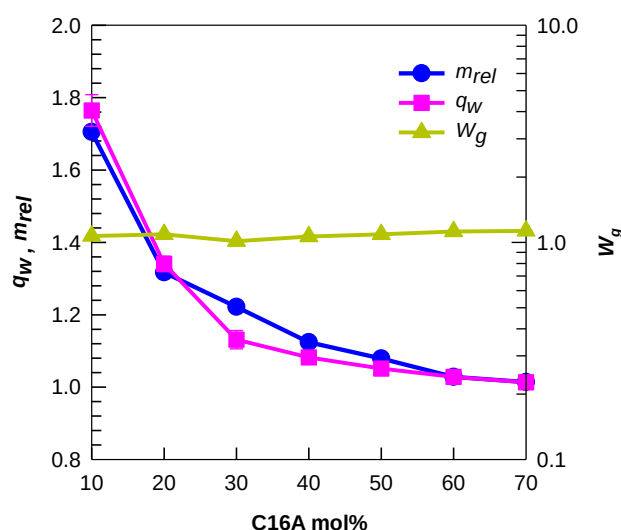


Figure 4.4 : Relative weight swelling ratio m_{rel} ($\square$), weight swelling ratio q_w ($\circ$), and gel fraction W_g ($\triangle$) of hydrogels with varying C16A mol%.

The data on swelling demonstrated a reduction in water uptake with an increase in C16A content. To gain further insight into this finding, differential scanning

calorimetry (DSC) was employed to analyze the thermal properties of the hydrogels. This analysis provides evidence of the existence of crystalline domains, which are primarily formed by the hydrophobic segments of C16A molecules. Figure 4.5 outlines the thermal properties of hydrogels through heating-cooling scans in (a) and (b), respectively, as well as the change in T_m , T_{cry} , and f_{cry} of the samples in (c).

As expected, the melting temperature of the hydrogel's crystalline domains increased proportionally with the rise in C16A mol%. It is important to note that this resulted in a wide range of T_m values, ranging from 34 °C to 40 °C, at 10 mol% and 70 mol% C16A, respectively. The results show that there is a correlation between the concentration of C16A and T_m values of the hydrogels. Also, as shown in Figure 4.5 (c), f_{cry} of the hydrogels increased proportionally with the amount of C16A present. This can be attributed to the formation of highly ordered and stable crystalline structures due to the presence of longer hydrophobic segments within the C16A units. These well-defined crystalline domains require a higher temperature to melt, resulting in the observed increase in T_m .

Building upon the information obtained from the DSC analysis, which indicated the formation of crystalline domains, X-ray diffraction (XRD) measurements were employed. By analyzing hydrogels with varying C16A content, alongside control samples, XRD allows us to investigate the influence of C16A on the crystallinity. This analysis included hydrogels with varying C16A content, alongside control samples of poly(DMAA-co-MAA) (0 mol% C16A) and pure poly(C16A) (100 mol% C16A). Figure 4.6 presented each of the XRD patterns individually. Also, to facilitate a clearer comparison of the crystal structures across the hydrogels, all XRD patterns are combined into a single plot in Figure 4.7.

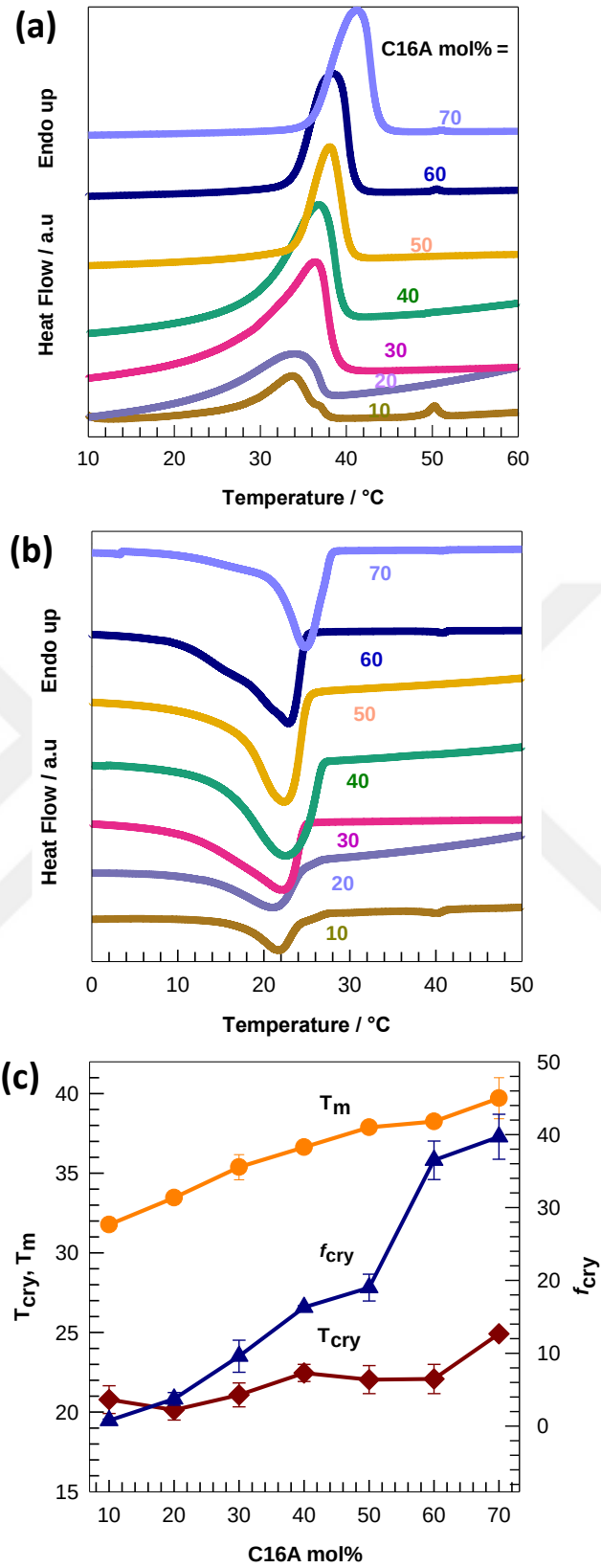


Figure 4.5 : Thermal properties of hydrogels with varying C16A mol%. (a) Heating scans, (b) cooling scans, and (c) changes in melting temperature T_m , crystallization temperature T_{cry} , and fraction of crystallinity f_{cry} of the hydrogels.

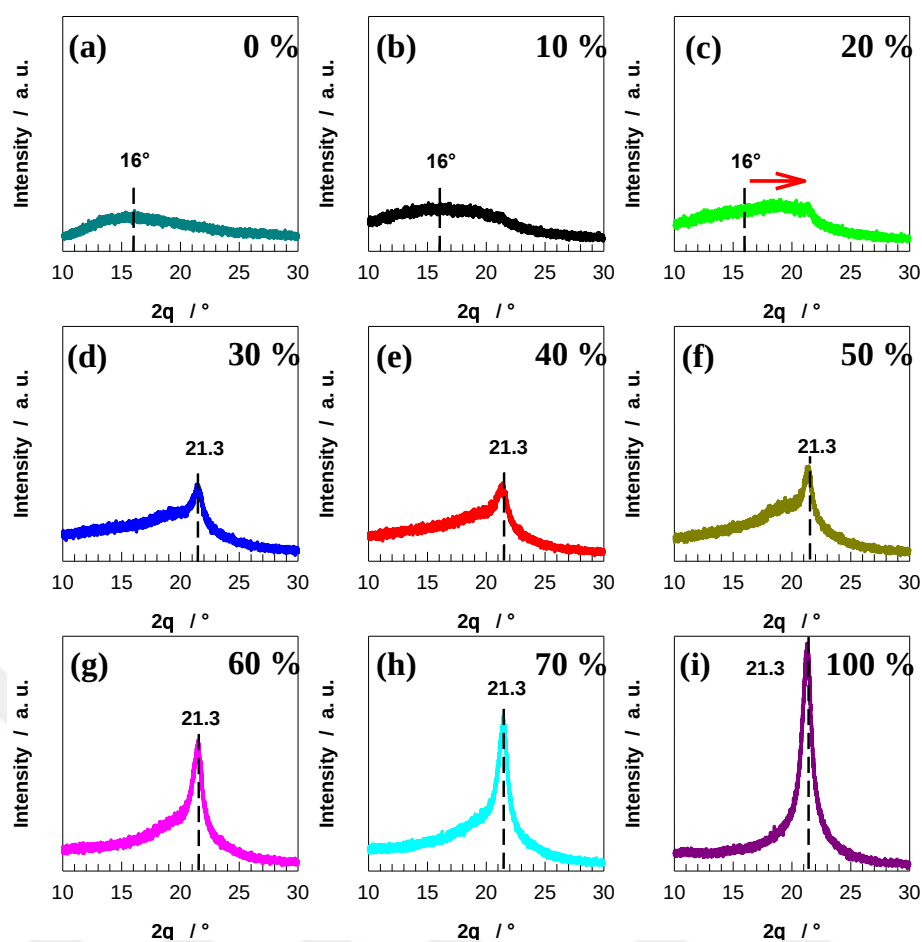


Figure 4.6 : XRD patterns of hydrogels with various C16A mol %= (a) 0, (b) 10, (c) 20, (d) 30, (e) 40, (f) 50, (g) 60, (h) 70, (i) 100.

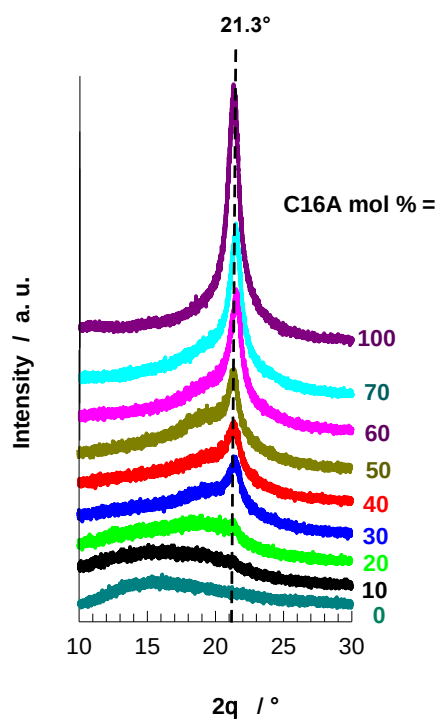


Figure 4.7 : The combined XRD patterns of hydrogels with various C16A content.

As shown in Figure 4.6(a), the XRD pattern of the hydrogel without C16A (0 mol%) exhibits a broad peak around 16° , characteristic of amorphous polymers. This broad peak signifies the lack of long-range order in the polymer chains. Eventually, this amorphous peak gradually diminishes in the patterns from (b) to (c), while a sharp new peak emerges at 21.3° starting from (d). This sharper peak at 21.3° corresponds to a d-spacing of 0.42 nm. This value is typical for a crystalline structure with side-by-side packed, paraffin-like alkyl chains. In our case, it signifies the alignment of the hexadecyl side chains originating from the C16A units within the hydrogel network. The d-spacing of 0.42 nm reflects the average distance between two neighboring hexadecyl side chains when aligned along the same polymer backbone. These observations from XRD, indicating increased crystallinity with higher C16A content, are in good agreement with the DSC results, which showed higher melting temperatures for hydrogels containing more C16A.

Rheologic tests were used to investigate the temperature-dependent viscoelastic properties of hydrogels with C16A mol% ranging from 40 to 70. Figure 4.8 demonstrates the temperature-dependent variation of storage modulus G' in (a) and loss factor $\tan \delta$ in (b).

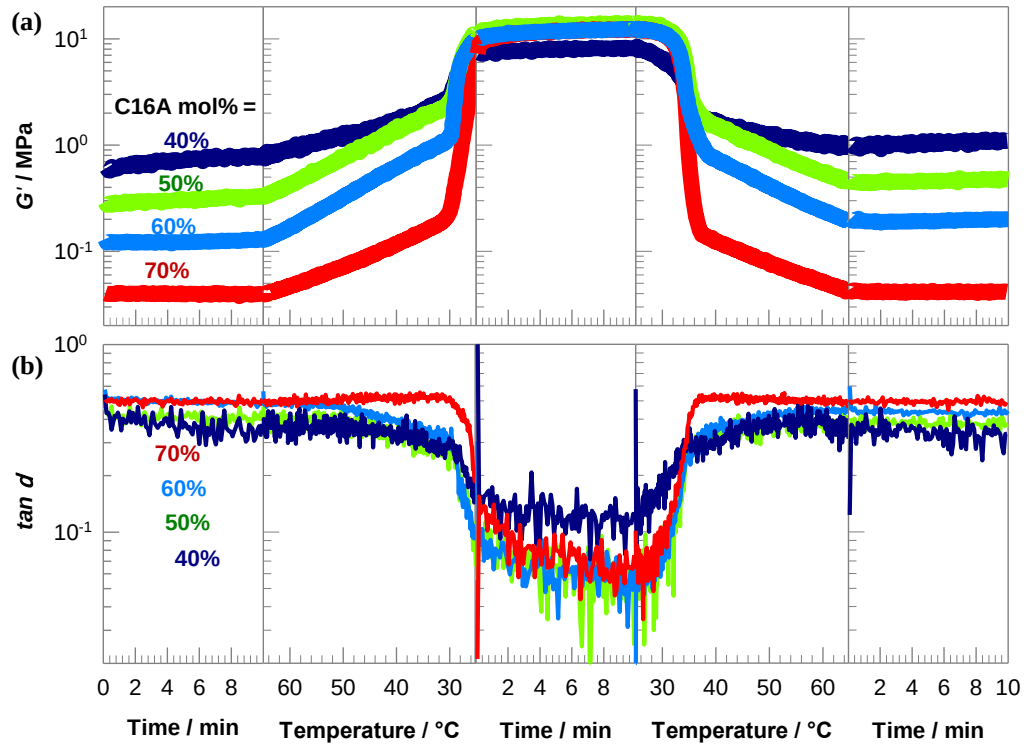


Figure 4.8 : The change in G' (a) and $\tan \delta$ (b) of hydrogels with different C16A contents during a heating-cooling cycle in the range of 25 – 65 °C. $\omega = 6.28 \text{ rad.s}^{-1}$. $\gamma_o = 0.1\%$.

As Figure 4.8 indicates, all tested hydrogels demonstrate clear reversible changes in storage modulus G' and loss factor $\tan \delta$ below and above their phase transition temperatures. It is important to note that the extent of these changes was vastly proportional to the amount of C16A present in the hydrogels. At temperatures below the phase transition temperature ($T < T_m$), the modulus increases with increasing C16A content due to the increasing degree of crystallinity. Conversely, at temperatures above the phase transition temperature ($T > T_m$), the modulus decreases with increasing C16A content.

As seen from Figure 4.8(a), at 65 °C, the G' value of hydrogels remained stable until the sample entered the cooling cycle. As the samples cooled from 65 °C to 25 °C, G' began to steadily increase. After reaching 25 °C, G' was maximized and maintained at the same level. The process was then repeated with a heating cycle to demonstrate its reversibility. These phase changes are supported by the change in $\tan \delta$ as shown in Figure 4.8(b). Below the transition temperature, a decreasing trend in $\tan \delta$ is observed for all hydrogels. This is because the polymer chains are more densely packed and have greater rigidity as a consequence of the increase in crystallinity.

The rheological measurements were also in agreement with the DSC measurements of hydrogels. Above the T_m of the hydrogels, the crystalline domains, acting as physical cross-linking points, are reversibly melted which causes them to become relatively soft.

Tensile data for the second set of hydrogels was also collected using the same method as the first set at a fixed strain rate of $\dot{\epsilon}$ of $1.89 \times 10^{-2} \text{ s}^{-1}$. In Figure 4.9 (a), stress-strain curves of hydrogels with various C16A mol% were shown, while the change in $E - \sigma_f$ and $W - \epsilon_f$ were depicted in (b) and (c), respectively. This combined view facilitates a clear understanding of how these mechanical properties depend on the C16A content.

As observed in Figure 4.9(a), both fracture stress σ_f and fracture strain ϵ_f exhibit an initial increase with increasing C16A content, reaching their maximum at 50 mol%. Yet, these values decrease slightly at C16A concentrations higher than 50 mol%. Consequently, the hydrogel containing 50 mol% C16A demonstrates superior mechanical properties, evident from its maximized Young's modulus E ($50.8 \pm 0.7 \text{ MPa}$), and toughness W ($8.15 \pm 0.7 \text{ MJ.m}^{-3}$) as depicted in Figure 4.9(b) and (c), respectively. Thus, the overall findings suggest that a balanced incorporation of C16A,

at an optimal level of 50 mol%, is crucial for achieving superior mechanical properties in these hydrogels.

The hydrogel's network structure responds differently to stress at varying speeds, affecting its mechanical characteristics. Faster deformation (i.e. higher strain rate $\dot{\epsilon}$) makes hydrogels stiffer and stronger (higher modulus E , yield stress σ_y , and fracture stress σ_f) but less stretchable (fracture strain ϵ_f). In order to evaluate this impact, the strain rate $\dot{\epsilon}$ was adjusted within a range of $0.95 \times 10^{-2} \text{ s}^{-1}$ to $4.54 \times 10^{-2} \text{ s}^{-1}$. As seen in Figure 4.9 (d), observational data reveal that the modulus E , yield stress σ_y , and fracture stress σ_f of hydrogels increase with increasing strain rate, while elongation at break ϵ_f decreases. Furthermore, Eyring theory can aid in elucidating the link between strain rate and the mechanical properties of hydrogels. The yield stress σ_y is directly proportional to the logarithm of the strain rate $\dot{\epsilon}$ according to Eyring theory, which assumes that polymer segments must overcome an energy barrier at the yield point. The relationship between these two factors can be examined by Equation 4.1.

$$\sigma_y = \frac{2kT}{V_a} \ln\left(\frac{\dot{\epsilon}}{\epsilon_o}\right) + \frac{2E_a}{V_a} \quad (4.1)$$

where the activation volume is represented by V_a , while ϵ_o stands for the pre-exponential factor. E_a refers to the activation energy, and kT represents the thermal energy. Figure 4.9 (e) displays the experimental σ_y versus $\dot{\epsilon}$ data on a logarithmic plot with the best fit obtained from Equation 4.1 (solid line). This analysis reveals an activation volume V_a of $2.3 \pm 0.2 \text{ nm}^3$. This indicates that the length of polymer segments involved in the deformation process, namely Kuhn length, was approximately 1.3 nm, which is equivalent to the length of the bond between two adjacent carbon atoms. This is an indication that individual bonds play a significant role in the mechanical response at the tested strain rates.

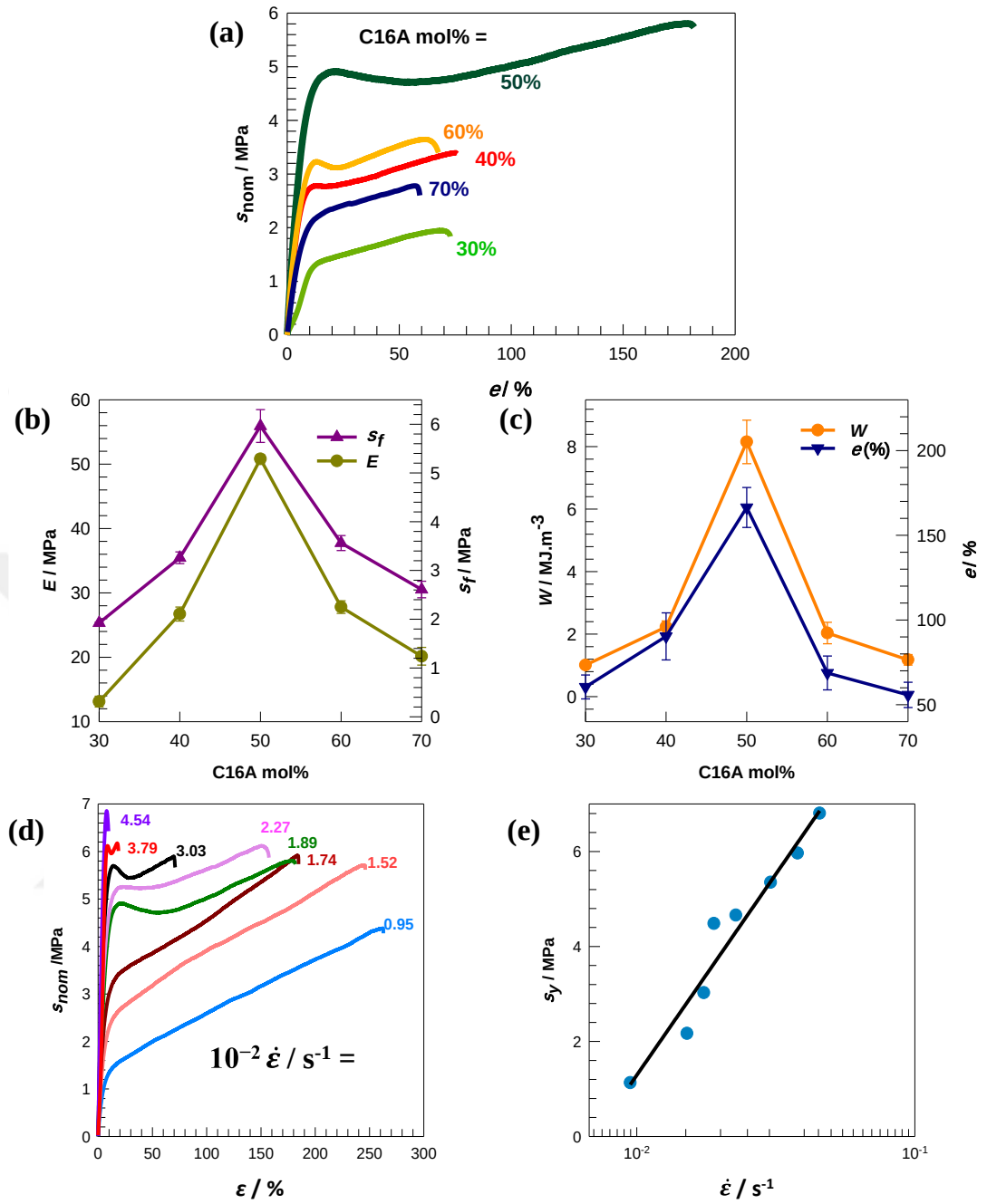


Figure 4.9 : (a) Stress-strain curves of hydrogels at a fixed $\dot{\epsilon}$ of $1.89 \times 10^{-2} \text{ s}^{-1}$, the changes in $E - s_f$ (b) and $W - e_f$ (c) of hydrogels with various C16A contents. (d) Stress-strain curves of hydrogel with 0.5 x_{DMAA} and 50 mol% C16A at various strain rates $\dot{\epsilon}$ as indicated. (e) Yield stress σ_y of the hydrogels plotted against the strain rate $\dot{\epsilon}$ in which symbols represent the experimental data while the solid line depicts the best-fitting line.

Considering the presence of switchable domains, which can act as reversible and inducible domains depending on the temperature, the shape memory behavior of the hydrogels was examined by means of a bending test. The rectangular samples, measuring 40 mm $\times$ 12 mm $\times$ 1 mm, were bent at 65 °C and then cooled to 25 °C while maintaining the applied force, resulting in a fixed temporary shape. The samples were

gradually heated to 65 °C, with each temperature being maintained for 5 minutes. Figure 4.10 shows the shape memory test for the specimen with 50 mol% C16A.

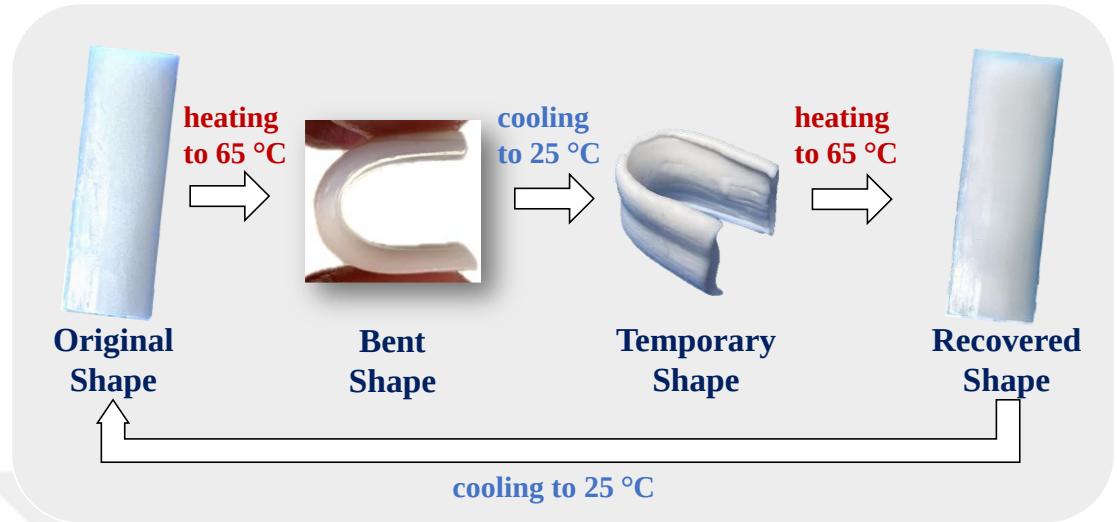


Figure 4.10 : Images of a bending test: shape fixing and shape recovery of a hydrogel sample with 50 mol% C16A.

Angular shape-recovery ratio $R_r(\theta)$, and angular shape-fixity ratio $R_f(\theta)$ of hydrogels were calculated from Equations (3.7) and (3.8) by recording specimen's recovery angle for every 1 °C rise in temperature. Figure 4.11(a) demonstrates the angular shape recovery ratio $R_r(\theta)$ of hydrogels with varying amounts of C16A across a temperature range of 25 °C – 65 °C. Also, in Figure 4.11(b), the calculated angular shape-fixity ratio $R_f(\theta)$ values of specimens were given. Angular shape-fixity ratio $R_f(\theta)$ values were calculated to be 98%, 98.8%, 99.2%, and 99.4% for specimens with C16A mol% of 40, 50, 60, and 70, respectively.

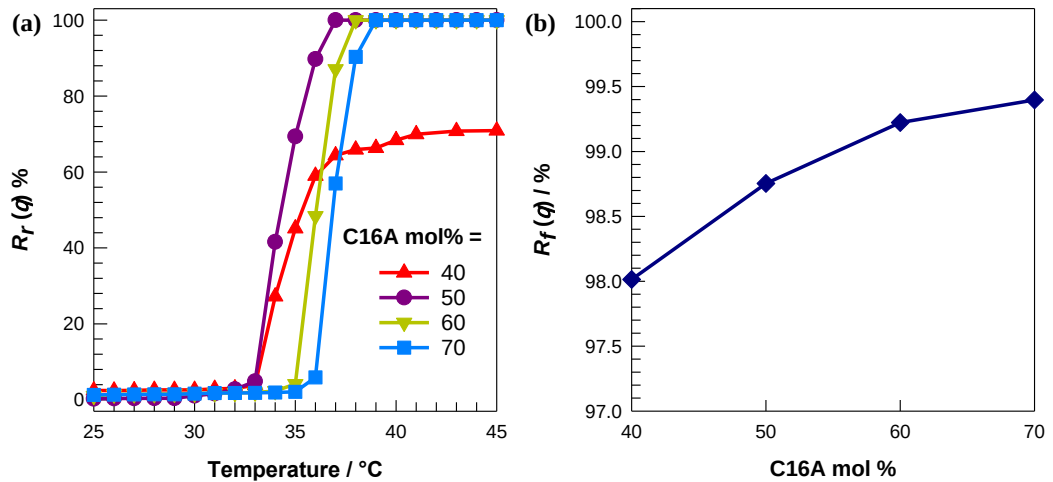


Figure 4.11 : (a) Angular shape-recovery ratios of a rectangular-shaped hydrogel sample (40 mm × 12 mm × 1 mm) as a function of temperature. (b) Angular shape-fixity ratio $R_f(\theta)$ of samples with 40, 50, 60, and 70 C16A mol%.

Figure 4.11(a) highlights the impressive shape recovery performance of the hydrogels. All samples, except for the one containing 40 mol% C16A, achieved a remarkable 100% shape recovery ratio. This exceptional ability can be attributed to the high crystallinity within the hydrogels. Notably, a significant increase in the shape recovery ratio is observed around 34 °C – 36 °C for all tested samples, and increasing the amount of C16A resulted in a more rapid angular recovery. Furthermore, Figure 4.11(b) demonstrates a positive correlation between C16A content and shape-fixity ratio $R_f(\theta)$. As the amount of C16A increases, the hydrogels exhibit a higher shape-fixity ratio. This trend suggests that C16A plays a crucial role in promoting both rapid angular recovery and efficient shape fixation. This observation aligns with the established connection between crystallinity and these shape-memory properties in hydrogels.

Shape-memory behavior of hydrogel was further demonstrated by a paperclip-shaped hydrogel with 50 mol% C16A and 0.5 x_{DMAA} , as seen in Figure 4.12. The paperclip was programmed into a temporary shape and then kept in a 50 °C water medium for under 1 min.



Figure 4.12 : 4D-printed paperclip-shaped hydrogel with 50 mol% C16A and 0.5 x_{DMAA} . From left to right: as prepared state, programmed temporary shape, and recovered permanent shape.

Figure 4.13 also brings the shape-memory properties to life with a captivating illustration. It depicts the shape recovery of a flower-shaped hydrogel containing 50 mol% C16A and 0.5 x_{DMAA} . The figure visually demonstrates the response to an external stimulus: dispensing hot water around 50 °C from a syringe. This triggers a transformation, resembling a blooming flower as the hydrogel visibly regains its original form.

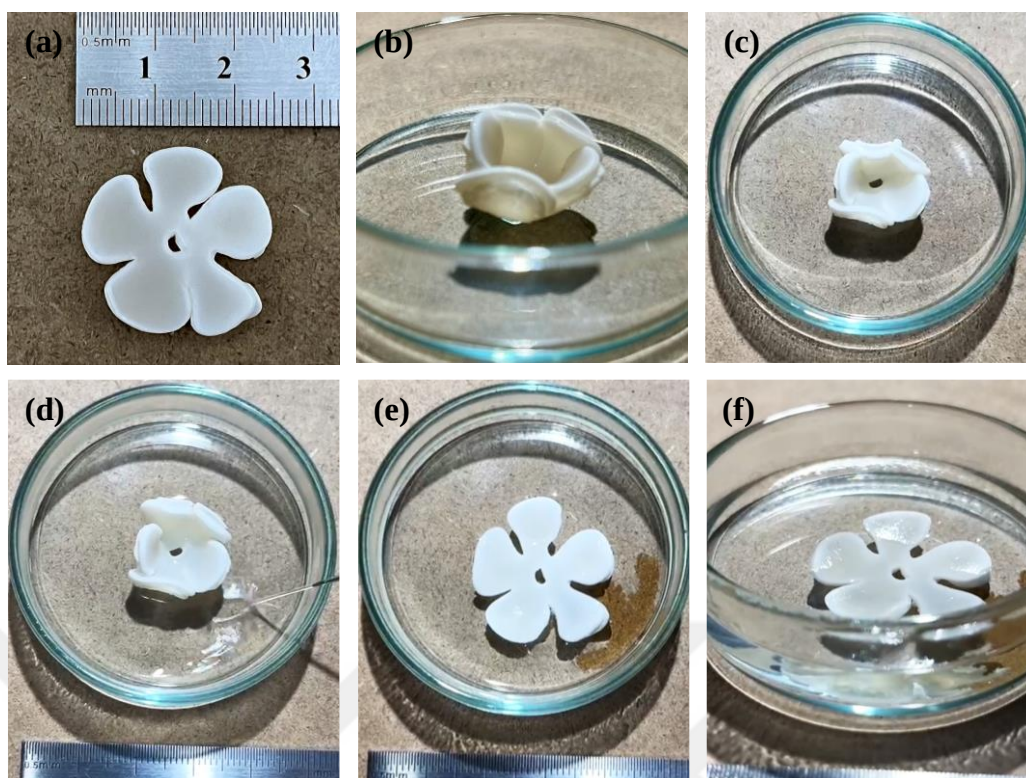
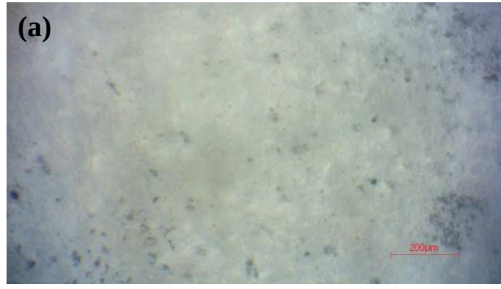


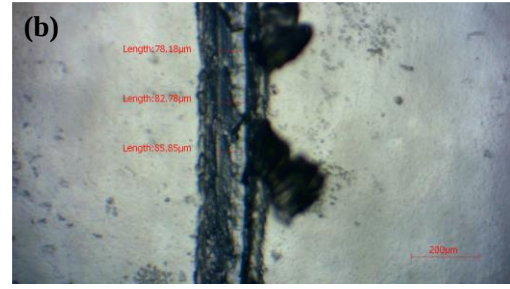
Figure 4.13 : Real-time shape recovery images of petal-shaped hydrogel with 50% C16A and 0.5 $\times$ DMAA. (a) 3D-printed flower permanent shape, (b) programmed temporary shape view from the side and (c) up, (d) transition to permanent shape as dispensing hot water from a syringe, (e) recovered permanent shape view from up and (f) side.

The self-healing behavior of the specimens was observed through a series of experiments. The specimens were scratched with a razor and then immersed in a 60 °C aqueous environment for 10 minutes. Self-healing images of a sample containing 50, 60, and 70 mol% C16A are shown in Figures 4.14, 4.15, and 4.16, respectively. These images were mainly taken of the plain polymer surface before any treatment (a), a scratched specimen surface under 4x magnification (b), and after a 10-minute healing process at 60 °C in water under 4x (c) and 10x magnification (d).

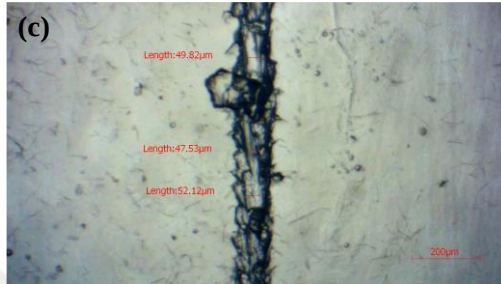
These figures reveal a clear correlation between C16A content and the extent of healing. As shown in Figure 4.14, while the scratch was not entirely healed, its width reduces significantly, from an average of 82.3 μm to 44.8 μm . Regarding 60 mol% C16A, Figure 4.15 shows that the scratch is nearly healed, with the width decreasing to just 5.7 μm on average from the initial 29.9 μm . On top of that, the specimen containing 70 mol% C16A has undergone a complete healing process, as evidenced by the disappearance of the 109.6 μm -width scratches in Figure 4.16.



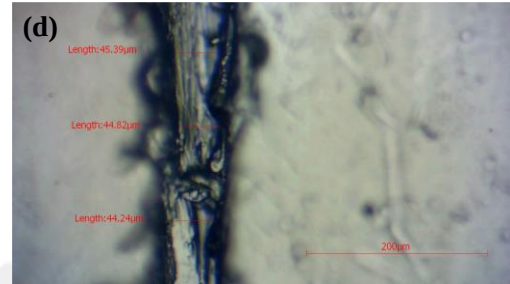
plain sample



scratched sample – 4x



healed sample – 4x

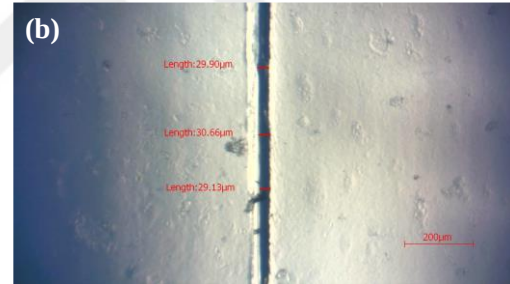


healed sample – 10x

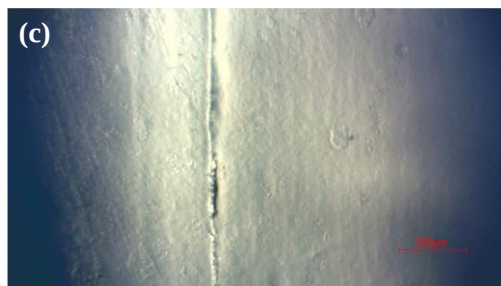
Figure 4.14 : Self-healing images of the specimen with 50 mol% C16A; prior to any treatment (a), a scratched specimen surface under 4x magnification (b), after being submerged in water at 60 °C for 10 minutes under 4x (c) and 10x magnification (d).



plain sample



scratched sample – 4x



healed sample – 4x



healed sample – 10x

Figure 4.15 : Self-healing images of the specimen with 60 mol% C16A; prior to any treatment (a), a scratched specimen surface under 4x magnification (b), after being submerged in water at 60 °C for 10 minutes under 4x (c) and 10x magnification (d).

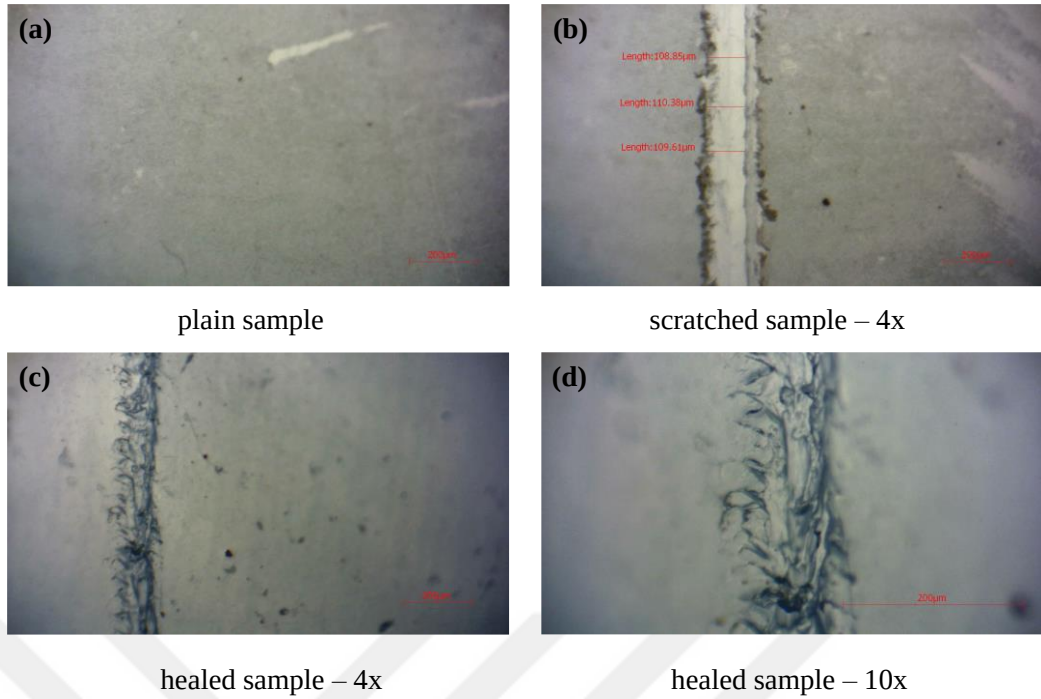


Figure 4.16 : Self-healing images of the specimen with 70 mol% C16A; prior to any treatment (a), a scratched specimen surface under 4x magnification (b), after being submerged in water at 60 °C for 10 minutes under 4x (c) and 10x magnification (d).

A rectangular-shaped hydrogel sample (60 mm × 10 mm × 2 mm) with 70 mol% C16A was used to display self-healing behavior in Figure 4.17. To achieve this, the sample (a) was first cut in half (b), and one part was marked for clarity (c). This was followed by a 10 min immersion in 60 °C water (d), with the specimen being held in place for the first 10 seconds. The healed sample (e) was then bent (f) and twisted (g) to show the integrity of healing. Finally, the sample was stretched to a stretch ratio of $283 \pm 10 \%$ successfully.

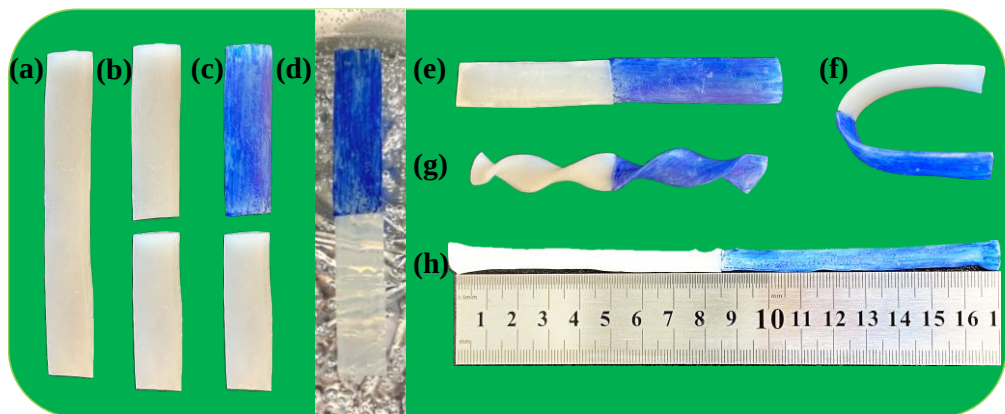


Figure 4.17 : Images of a specimen with 70 mol% C16A before (a) and after cutting (b), after coloring one of its parts with a marker for clarity (c), and after healing at 60 °C for 10 minutes (d). The healed specimen (e) can be twisted (g), bent (f), and stretched to a stretch ratio of $283 \pm 10 \%$.

The self-healing abilities of the hydrogels were explored through an evaluation of their mechanical properties, considering both the virgin and healed states. The plots in Figure 4.18(a) present stress-strain curves for virgin hydrogels (solid lines) and their healed counterparts (dashed lines) for samples containing 50 mol%, 60 mol%, and 70 mol% C16A. The recovery of mechanical properties after healing can be visually assessed by comparing these curves. Accordingly, the efficiency of healing for Young's modulus E , fracture stress σ_f , and fracture strain ε_f is represented in Figure 4.18(b). Overall, increasing the C16A amount significantly improved the efficiencies of E , σ_f , ε_f . Young's modulus E efficiencies were 91.8% for 50 mol%, 97.4% for 60 mol%, and 99.5% for 70 mol% C16A. Fracture stress σ_f efficiencies were 83.9% for 50 mol%, 87.8% for 60 mol%, 88.5% for 70 mol% C16A. Fracture strain ε_f efficiency initially lagged behind at 50 mol% C16A (11.5%) but showed improvement with higher C16A content, reaching 63.2% for 60 mol% and 74% for 70 mol%.

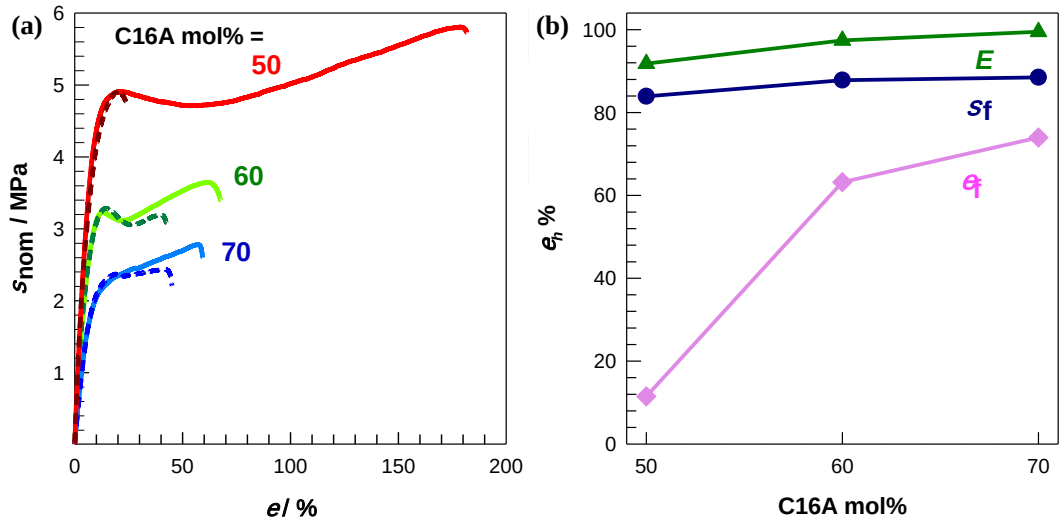


Figure 4.18 : (a) The stress-strain curves for the virgin (solid line) and healed (dashed line) specimens with 50, 60, and 70 mol% C16A, (b) efficiencies of Young's modulus E , fracture stress σ_f , and fracture strain ε_f of specimens.

5. CONCLUSIONS

To conclude, this master's thesis successfully demonstrated 4D-printing of an acrylate-based hydrogel composed of hydrophobic C16A and hydrophilic DMAA and MAA monomers. Utilizing a precisely chosen polymer system and a commercial resin printer with 405 nm UV light, this study highlights the integration of 4D printing for creating novel smart hydrogels. The change in the DMAA:MAA mole ratio revealed a wide range of mechanical characteristics, whether brittle or ductile. By optimizing the DMAA:MAA ratio to 1:1 ($x_{\text{DMAA}} = 0.5$), the hydrogel achieved the highest toughness of $2.23 \pm 0.2 \text{ MJ.m}^{-3}$. The second part of the study revealed precise control over the actuation temperature ($34^\circ\text{C} - 40^\circ\text{C}$) by fine-tuning the C16A content, enabling tailored functionality for a specific environment. This temperature control, achievable at body temperature, positions these hydrogels for pioneering biomedical applications, such as targeted drug delivery.

In terms of rheological properties, the tested hydrogels demonstrated reversible alterations in storage modulus G' and loss factor $\tan \delta$ in response to temperature. The magnitude of these changes increased with increasing C16A content. This reversible rheological behavior is in accordance with the melting and reformation of crystalline domains at the hydrogels' phase transition temperature. Also, shape memory tests showed exceptional performance for 40, 50, 60, and 70 mol% C16A samples having a x_{DMAA} of 0.5, with shape-fixity ratios $R_f(\theta)$ exceeding 98% and rapid actuation times of less than 30 seconds. Actuations of these hydrogels occurred around $34^\circ\text{C} - 36^\circ\text{C}$, as the higher C16A content led to more rapid angular recovery. These results indicate that C16A is crucial for promoting rapid angular recovery and efficient shape fixation, aligning with the established link between crystallinity and shape-memory properties in hydrogels. Except for the sample with 40 mol% C16A, all tested samples achieved a 100% shape recovery ratio. Additionally, hydrogels demonstrated self-healing behavior, which again improved with increasing C16A content, reaching a 100% healing with 70 mol% C16A.

In summary, this research addresses the key limitations of traditional hydrogels by developing a novel polymer with an exceptional combination of shape-memory, self-healing capabilities, and superior mechanical properties. Notably, these tough hydrogels achieve rapid actuation (less than 30 seconds) with a tunable actuation temperature range, enabling precise control over their shape-shifting behavior. These unique properties position this material as a promising candidate for various advanced applications. While this research focuses on establishing a fundamental understanding of these functionalities, the potential for future exploration extends to diverse fields beyond the current scope. For instance, the tunable actuation temperature offers significant possibilities in areas such as drug delivery systems and biomedical applications, particularly in targeting specific body parts. Furthermore, additional research could investigate the suitability of this material for biomedical applications, leveraging its distinctive properties to design innovative devices capable of controlled shape changes triggered by specific temperatures.

REFERENCES

- 3D Printed Drugs Market.** (2017). Allied Market Research. Retrieved May 12, 2024, from <https://www.alliedmarketresearch.com/3d-printed-drugs-market>
- 3D Printed Drugs Market Size, Share, Competitive Landscape and Trend Analysis Report by Spritam and Other Drugs Potential Market by Scenario - Global Opportunity Analysis and Industry Forecast, 2017-2030.** (n.d.). Allied Market Research. <https://www.alliedmarketresearch.com/3d-printed-drugs-market>
- 3D Printing Technology Insight Report.** (2014). Patent Insight Pro. Retrieved April 21, 2024, from <http://www.patentinsightpro.com/techreports/0214/Tech%20Insight%20Report%20-%203D%20Printing.pdf>
- Abdurrahmanoglu, S., Can, V., & Okay, O.** (2009, November). Design of high-toughness polyacrylamide hydrogels by hydrophobic modification. *Polymer*, 50(23), 5449–5455. <https://doi.org/10.1016/j.polymer.2009.09.042>
- Aguilar, M. R., & Román, J. S.** (2019, February 15). *Smart Polymers and Their Applications*. Woodhead Publishing. <https://doi.org/10.1016/b978-0-85709-695-1.50020-0>
- Ashley, G. W., Henise, J., Reid, R., & Santi, D. V.** (2013, January 23). Hydrogel drug delivery system with predictable and tunable drug release and degradation rates. *Proceedings of the National Academy of Sciences*, 110(6), 2318–2323. <https://doi.org/10.1073/pnas.1215498110>
- Ashraf, S., Park, H. K., Park, H., & Lee, S. H.** (2016, April). Snapshot of phase transition in thermoresponsive hydrogel PNIPAM: Role in drug delivery and tissue engineering. *Macromolecular Research*, 24(4), 297–304. <https://doi.org/10.1007/s13233-016-4052-2>
- Bagheri, A., & Jin, J.** (2019, February 20). Photopolymerization in 3D Printing. *ACS Applied Polymer Materials*, 1(4), 593–611. <https://doi.org/10.1021/acsapm.8b00165>
- Bakar, N. S. A., Alkahari, M. R., & Boejang, H.** (2010, December). Analysis on fused deposition modelling performance. *Journal of Zhejiang University-SCIENCE A*, 11(12), 972–977. <https://doi.org/10.1631/jzus.a1001365>
- Beg, S., Almalki, W. H., Malik, A., Farhan, M., Aatif, M., Rahman, Z., Alruwaili, N. K., Alrobaian, M., Tarique, M., & Rahman, M.** (2020, September). 3D printing for drug delivery and biomedical applications. *Drug Discovery Today*, 25(9), 1668–1681. <https://doi.org/10.1016/j.drudis.2020.07.007>

- Bilici, C., Can, V., Nöchel, U., Behl, M., Lendlein, A., & Okay, O.** (2016, September 20). Melt-Processable Shape-Memory Hydrogels with Self-Healing Ability of High Mechanical Strength. *Macromolecules*, 49(19), 7442–7449. <https://doi.org/10.1021/acs.macromol.6b01539>
- Billiet, T., Vandenhaute, M., Schelfhout, J., Van Vlierberghe, S., & Dubruel, P.** (2012, September). A review of trends and limitations in hydrogel-rapid prototyping for tissue engineering. *Biomaterials*, 33(26), 6020–6041. <https://doi.org/10.1016/j.biomaterials.2012.04.050>
- Bromberg, L.** (1998, May 4). Temperature-responsive gels and thermogelling polymer matrices for protein and peptide delivery. *Advanced Drug Delivery Reviews*, 31(3), 197–221. [https://doi.org/10.1016/s0169-409x\(97\)00121-x](https://doi.org/10.1016/s0169-409x(97)00121-x)
- Cao, R., Liu, H., Chen, S., Pei, D., Miao, J., & Zhang, X.** (2017). Fabrication and properties of graphene oxide-grafted-poly(hexadecyl acrylate) as a solid-solid phase change material. *Composites Science and Technology*, 149, 262–268. <https://doi.org/10.1016/j.compscitech.2017.06.019>
- Chang, X., Geng, Y., Cao, H., Zhou, J., Tian, Y., Shan, G., Bao, Y., Wu, Z. L., & Pan, P.** (2018, January 31). Dual-Crosslink Physical Hydrogels with High Toughness Based on Synergistic Hydrogen Bonding and Hydrophobic Interactions. *Macromolecular Rapid Communications*, 39(14). <https://doi.org/10.1002/marc.201700806>
- Chen, C., Wang, X., Wang, Y., Yang, D., Yao, F., Zhang, W., Wang, B., Sewvandi, G. A., Yang, D., & Hu, D.** (2020, September 20). Additive Manufacturing of Piezoelectric Materials. *Advanced Functional Materials*, 30(52). <https://doi.org/10.1002/adfm.202005141>
- Chen, Z., Zhao, D., Liu, B., Nian, G., Li, X., Yin, J., Qu, S., & Yang, W.** (2019, March 25). 3D Printing of Multifunctional Hydrogels. *Advanced Functional Materials*, 29(20). <https://doi.org/10.1002/adfm.201900971>
- Coombes, A., Verderio, E., Shaw, B., Li, X., Griffin, M., & Downes, S.** (2002, May). Biocomposites of non-crosslinked natural and synthetic polymers. *Biomaterials*, 23(10), 2113–2118. [https://doi.org/10.1016/s0142-9612\(01\)00341-6](https://doi.org/10.1016/s0142-9612(01)00341-6)
- Dev Singh, D., Mahender, T., & Raji Reddy, A.** (2021). Powder bed fusion process: A brief review. *Materials Today: Proceedings*, 46, 350–355. <https://doi.org/10.1016/j.matpr.2020.08.415>
- Devi V. K., A., Shyam, R., Palaniappan, A., Jaiswal, A. K., Oh, T. H., & Nathanael, A. J.** (2021, October 31). Self-Healing Hydrogels: Preparation, Mechanism and Advancement in Biomedical Applications. *Polymers*, 13(21), 3782. <https://doi.org/10.3390/polym13213782>
- Dong, J., Chen, L., Ding, Y., & Han, W.** (2005, September 28). Swelling and Mechanical Properties of a Temperature-Sensitive Dextran Hydrogel and its Bioseparation Applications. *Macromolecular Chemistry and Physics*, 206(19), 1973–1980. <https://doi.org/10.1002/macp.200500125>

- Dvořák, V., Votrubec, R., Šafka, J., & Kracík, J.** (2018). Experimental investigation of centrifugal fans for personal protection equipment – effect of used 3D printing technologies. *EPJ Web of Conferences*, 180, 02023. <https://doi.org/10.1051/epjconf/201818002023>
- Ebara, M., Kotsuchibashi, Y., Narain, R., Idota, N., Kim, Y. J., Hoffman, J. M., Uto, K., & Aoyagi, T.** (2014, May 28). *Smart Biomaterials*. Springer. ISBN: 9784431544005
- Ebara, M., Kotsuchibashi, Y., Uto, K., Aoyagi, T., Kim, Y. J., Narain, R., Idota, N., & Hoffman, J. M.** (2014). Smart Hydrogels. *NIMS Monographs*, 9–65. <https://doi.org/10.1007/978-4-431-54400-5>
- Falahati, M., Ahmadvand, P., Safaei, S., Chang, Y. C., Lyu, Z., Chen, R., Li, L., & Lin, Y.** (2020, November). Smart polymers and nanocomposites for 3D and 4D printing. *Materials Today*, 40, 215–245. <https://doi.org/10.1016/j.mattod.2020.06.001>
- Fang, X., Li, Y., Li, X., Liu, W., Yu, X., Yan, F., & Sun, J.** (2020, June 5). Dynamic Hydrophobic Domains Enable the Fabrication of Mechanically Robust and Highly Elastic Poly(vinyl alcohol)-Based Hydrogels with Excellent Self-Healing Ability. *ACS Materials Letters*, 2(7), 764–770. <https://doi.org/10.1021/acsmaterialslett.0c00075>
- Ferreira, N., Ferreira, L., Cardoso, V., Boni, F., Souza, A., & Gremião, M.** (2018, February). Recent advances in smart hydrogels for biomedical applications: From self-assembly to functional approaches. *European Polymer Journal*, 99, 117–133. <https://doi.org/10.1016/j.eurpolymj.2017.12.004>
- Gil, E., & Hudson, S.** (2004, December). Stimuli-responsive polymers and their bioconjugates. *Progress in Polymer Science*, 29(12), 1173–1222. <https://doi.org/10.1016/j.progpolymsci.2004.08.003>
- Global 3D printing industry market size.** (2022, September 29). Statista. Retrieved May 11, 2024, from <https://www.statista.com/statistics/315386/global-market-for-3d-printers/>
- Guzzi, E. A., & Tibbitt, M. W.** (2019, August 18). Additive Manufacturing of Precision Biomaterials. *Advanced Materials*, 32(13). <https://doi.org/10.1002/adma.201901994>
- Holback, H., Yeo, Y., & Park, K.** (2011). Hydrogel swelling behavior and its biomedical applications. *Biomedical Hydrogels*, 3–24. <https://doi.org/10.1533/9780857091383.1.3>
- Hsissou, R., Seghiri, R., Benzekri, Z., Hilali, M., Rafik, M., & Elharfi, A.** (2021, April). Polymer composite materials: A comprehensive review. *Composite Structures*, 262, 113640. <https://doi.org/10.1016/j.compstruct.2021.113640>
- Hu, L., Wan, Y., Zhang, Q., & Serpe, M. J.** (2020, January). Stimuli-Responsive Actuation: Harnessing the Power of Stimuli-Responsive Polymers for Actuation (Adv. Funct. Mater. 2/2020). *Advanced Functional Materials*, 30(2). <https://doi.org/10.1002/adfm.202070012>

- Inomata, K., Terahama, T., Sekoguchi, R., Ito, T., Sugimoto, H., & Nakanishi, E.** (2012, July). Shape memory properties of polypeptide hydrogels having hydrophobic alkyl side chains. *Polymer*, 53(15), 3281–3286. <https://doi.org/10.1016/j.polymer.2012.05.036>
- Ischakov, R., Adler-Abramovich, L., Buzhansky, L., Shekhter, T., & Gazit, E.** (2013, June). Peptide-based hydrogel nanoparticles as effective drug delivery agents. *Bioorganic & Medicinal Chemistry*, 21(12), 3517–3522. <https://doi.org/10.1016/j.bmc.2013.03.012>
- Jeon, I., Cui, J., Illeperuma, W. R. K., Aizenberg, J., & Vlassak, J. J.** (2016, April 9). Extremely Stretchable and Fast Self-Healing Hydrogels. *Advanced Materials*, 28(23), 4678–4683. <https://doi.org/10.1002/adma.201600480>
- Jian, Y., Wu, B., Yang, X., Peng, Y., Zhang, D., Yang, Y., Qiu, H., Lu, H., Zhang, J., & Chen, T.** (2022, December). Stimuli-responsive hydrogel sponge for ultrafast responsive actuator. *Supramolecular Materials*, 1, 100002. <https://doi.org/10.1016/j.supmat.2021.100002>
- Jiménez, M., Romero, L., Domínguez, I. A., Espinosa, M. D. M., & Domínguez, M.** (2019, February 19). Additive Manufacturing Technologies: An Overview about 3D Printing Methods and Future Prospects. *Complexity*, 2019, 1–30. <https://doi.org/10.1155/2019/9656938>
- Kabiri, K., & Zohuriaan-Mehr, M. J.** (2004, June 29). Porous Superabsorbent Hydrogel Composites: Synthesis, Morphology and Swelling Rate. *Macromolecular Materials and Engineering*, 289(7), 653–661. <https://doi.org/10.1002/mame.200400010>
- Kasiński, A., Zielińska-Pisklak, M., Oledzka, E., & Sobczak, M.** (2020, June). Smart Hydrogels – Synthetic Stimuli-Responsive Antitumor Drug Release Systems. *International Journal of Nanomedicine*, Volume 15, 4541–4572. <https://doi.org/10.2147/ijn.s248987>
- Khoo, Z. X., Teoh, J. E. M., Liu, Y., Chua, C. K., Yang, S., An, J., Leong, K. F., & Yeong, W. Y.** (2015, July 3). 3D printing of smart materials: A review on recent progresses in 4D printing. *Virtual and Physical Prototyping*, 10(3), 103–122. <https://doi.org/10.1080/17452759.2015.1097054>
- Klak, M., Gomółka, M., Dobrzański, T., Tymicki, G., Cywoniuk, P., Kowalska, P., Kosowska, K., Bryniarski, T., Berman, A., Dobrzyń, A., Idaszek, J., Świąszkowski, W., & Wszola, M.** (2020, June 25). Irradiation with 365 nm and 405 nm wavelength shows differences in DNA damage of swine pancreatic islets. *PLOS ONE*, 15(6), e0235052. <https://doi.org/10.1371/journal.pone.0235052>
- Kodaira, T., Yang, J. Z., & Aida, H.** (1988). Solvent Effect on the Radical Copolymerization of N-Methyl- and N,N-Dimethylacrylamides and N-Methylmethacrylamide with Methyl Methacrylate. *Polymer Journal*, 20(11), 1021–1029. <https://doi.org/10.1295/polymj.20.1021>
- Kumar, V., Kaur, H., Kumari, A., Hooda, G., Garg, V., & Dureja, H.** (2023, December). Drug delivery and testing via 3D printing. *Bioprinting*, 36, e00298. <https://doi.org/10.1016/j.bprint.2023.e00298>

- Kumashiro, Y., Yamato, M., & Okano, T.** (2010, April 13). Cell Attachment–Detachment Control on Temperature-Responsive Thin Surfaces for Novel Tissue Engineering. *Annals of Biomedical Engineering*, 38(6), 1977–1988. <https://doi.org/10.1007/s10439-010-0035-1>
- Lakkala, P., Munnangi, S. R., Bandari, S., & Repka, M.** (2023, December). Additive manufacturing technologies with emphasis on stereolithography 3D printing in pharmaceutical and medical applications: A review. *International Journal of Pharmaceutics: X*, 5, 100159. <https://doi.org/10.1016/j.ijpx.2023.100159>
- Lancker, B. V., Corte, W. D., & Belis, J.** (2024, April). Investigation of the structural performance of continuous adhesive glass-metal connections using structural silicone and hybrid polymer adhesives. *Engineering Structures*, 304, 117612. <https://doi.org/10.1016/j.engstruct.2024.117612>
- Le, V. T., Paris, H., & Mandil, G.** (2018, August). The development of a strategy for direct part reuse using additive and subtractive manufacturing technologies. *Additive Manufacturing*, 22, 687–699. <https://doi.org/10.1016/j.addma.2018.06.026>
- Lendlein, A., & Gould, O. E. C.** (2019, January 9). Reprogrammable recovery and actuation behaviour of shape-memory polymers. *Nature Reviews Materials*, 4(2), 116–133. <https://doi.org/10.1038/s41578-018-0078-8>
- Li, H., & Yew, Y. K.** (2009, August). Simulation of soft smart hydrogels responsive to pH stimulus: Ionic strength effect and case studies. *Materials Science and Engineering: C*, 29(7), 2261–2269. <https://doi.org/10.1016/j.msec.2009.05.011>
- Li, J. W., Chen, H. F., Liu, Y. Z., Wang, J. H., Lu, M. C., & Chiu, C. W.** (2024, March). Photocurable 3D-printed AgNPs/Graphene/Polymer nanocomposites with high flexibility and stretchability for ECG and EMG smart clothing. *Chemical Engineering Journal*, 484, 149452. <https://doi.org/10.1016/j.cej.2024.149452>
- Li, M., Liang, Y., He, J., Zhang, H., & Guo, B.** (2020, November 24). Two-Pronged Strategy of Biomechanically Active and Biochemically Multifunctional Hydrogel Wound Dressing to Accelerate Wound Closure and Wound Healing. *Chemistry of Materials*, 32(23), 9937–9953. <https://doi.org/10.1021/acs.chemmater.0c02823>
- Li, M., Wang, X., Dong, B., & Sitti, M.** (2020, August 10). In-air fast response and high speed jumping and rolling of a light-driven hydrogel actuator. *Nature Communications*, 11(1). <https://doi.org/10.1038/s41467-020-17775-4>
- Liang, X., Ding, H., Wang, Q., & Sun, G.** (2019, November). Tough physical hydrogels reinforced by hydrophobic association with remarkable mechanical property, rapid stimuli-responsiveness and fast self-recovery capability. *European Polymer Journal*, 120, 109278. <https://doi.org/10.1016/j.eurpolymj.2019.109278>

- Liao, J., Huang, J., Yang, S., Wang, X., Wang, T., Sun, W., & Tong, Z.** (2019, August 22). Ultra-Strong and Fast Response Gel by Solvent Exchange and Its Shape Memory Applications. *ACS Applied Polymer Materials*, 1(10), 2703–2712. <https://doi.org/10.1021/acsapm.9b00654>
- Liu, K., Zhang, Y., Cao, H., Liu, H., Geng, Y., Yuan, W., Zhou, J., Wu, Z. L., Shan, G., Bao, Y., Zhao, Q., Xie, T., & Pan, P.** (2020, May 28). Programmable Reversible Shape Transformation of Hydrogels Based on Transient Structural Anisotropy. *Advanced Materials*, 32(28). <https://doi.org/10.1002/adma.202001693>
- Lolaeva, A. V., Zhivchikova, A. N., Tepliakova, M. M., Sagdullina, D. K., Akkuratov, A. V., & Kuznetsov, I. E.** (2024, April). Facile metal-free synthesis of thiazolo[5,4-d]thiazole-containing conjugated polymers for thin-film electronics. *Synthetic Metals*, 117614. <https://doi.org/10.1016/j.synthmet.2024.117614>
- Maines, E. M., Porwal, M. K., Ellison, C. J., & Reineke, T. M.** (2021). Sustainable advances in SLA/DLP 3D printing materials and processes. *Green Chemistry*, 23(18), 6863–6897. <https://doi.org/10.1039/d1gc01489g>
- Malekmohammadi, S., Sedghi Aminabad, N., Sabzi, A., Zarebkohan, A., Razavi, M., Vosough, M., Bodaghi, M., & Maleki, H.** (2021, October 26). Smart and Biomimetic 3D and 4D Printed Composite Hydrogels: Opportunities for Different Biomedical Applications. *Biomedicines*, 9(11), 1537. <https://doi.org/10.3390/biomedicines9111537>
- Manapat, J. Z., Chen, Q., Ye, P., & Advincula, R. C.** (2017, May 22). 3D Printing of Polymer Nanocomposites via Stereolithography. *Macromolecular Materials and Engineering*, 302(9). <https://doi.org/10.1002/mame.201600553>
- Matsuo, E., & Tanaka, T.** (1988). Kinetics of discontinuous volume–phase transition of gels. *The Journal of Chemical Physics*, 1695–1703. <https://doi.org/10.1063/1.455115>
- Mawad, D., Stewart, E., Officer, D. L., Romeo, T., Wagner, P., Wagner, K., & Wallace, G. G.** (2012, April 12). A Single Component Conducting Polymer Hydrogel as a Scaffold for Tissue Engineering. *Advanced Functional Materials*, 22(13), 2692–2699. <https://doi.org/10.1002/adfm.201102373>
- Mazzoli, A.** (2012, December 19). Selective laser sintering in biomedical engineering. *Medical & Biological Engineering & Computing*, 51(3), 245–256. <https://doi.org/10.1007/s11517-012-1001-x>
- Momeni, F., M.Mehdi Hassani, N. S., Liu, X., & Ni, J.** (2017, May). A review of 4D printing. *Materials & Design*, 122, 42–79. <https://doi.org/10.1016/j.matdes.2017.02.068>
- Mu, M., Huang, J., Chen, Y., Li, Y., Yu, H. Y., & Yao, J.** (2022). Simultaneously Enhance the Heat Storage Capacity and Shape Stability of Poly(hexadecyl acrylate) by Immobilizing It to Cellulose Nanocrystal via In Situ Free Radical Polymerization. *ACS Applied Polymer Materials*, 4(2), 899–907. <https://doi.org/10.1021/acsapm.1c01395>

- Naahidi, S., Jafari, M., Logan, M., Wang, Y., Yuan, Y., Bae, H., Dixon, B., & Chen, P.** (2017, September). Biocompatibility of hydrogel-based scaffolds for tissue engineering applications. *Biotechnology Advances*, 35(5), 530–544. <https://doi.org/10.1016/j.biotechadv.2017.05.006>
- Namazi, H.** (2017, June 16). Polymers in our daily life. *BioImpacts*, 7(2), 73–74. <https://doi.org/10.15171/bi.2017.09>
- Neuhaus, R., Zahiri, N., Petrs, J., Tahouni, Y., Siegert, J., Kolaric, I., Dahy, H., & Bauernhansl, T.** (2020, July 2). Integrating Ionic Electroactive Polymer Actuators and Sensors into Adaptive Building Skins – Potentials and Limitations. *Frontiers in Built Environment*, 6. <https://doi.org/10.3389/fbuil.2020.00095>
- Ngo, T. D., Kashani, A., Imbalzano, G., Nguyen, K. T., & Hui, D.** (2018, June). Additive manufacturing (3D printing): A review of materials, methods, applications and challenges. *Composites Part B: Engineering*, 143, 172–196. <https://doi.org/10.1016/j.compositesb.2018.02.012>
- Noor, N., Shapira, A., Edri, R., Gal, I., Wertheim, L., & Dvir, T.** (2019, April 15). 3D Printing of Personalized Thick and Perfusable Cardiac Patches and Hearts. *Advanced Science*, 6(11). <https://doi.org/10.1002/advs.201900344>
- Okay, O.** (2019). Semicrystalline physical hydrogels with shape-memory and self-healing properties. *Journal of Materials Chemistry B*, 7(10), 1581–1596. <https://doi.org/10.1039/c8tb02767f>
- Oral, C., Yetiskin, B., & Okay, O.** (2020). Stretchable silk fibroin hydrogels. *International Journal of Biological Macromolecules*, 161, 1371–1380. <https://doi.org/10.1016/j.ijbiomac.2020.08.040>
- Osada, Y., & Matsuda, A.** (1995, July). Shape memory in hydrogels. *Nature*, 376(6537), 219–219. <https://doi.org/10.1038/376219a0>
- Osswald, T. A., & Menges, G.** (2012, January 1). *Material Science of Polymers for Engineers*. Hanser Publications. ISBN: 9781569905142
- Park, J. H., Jang, J. W., Sim, J. H., Kim, I. J., Lee, D. J., Lee, Y. H., Park, S. H., & Kim, H. D.** (2019, November 13). Preparation and Properties of Thermoresponsive P(N-Isopropylacrylamide-co-butylacrylate) Hydrogel Materials for Smart Windows. *International Journal of Polymer Science*, 2019, 1–7. <https://doi.org/10.1155/2019/3824207>
- Peng, T.** (2016). Analysis of Energy Utilization in 3D Printing Processes. *Procedia CIRP*, 40, 62–67. <https://doi.org/10.1016/j.procir.2016.01.055>
- Peppas, N.** (2000, July 3). Hydrogels in pharmaceutical formulations. *European Journal of Pharmaceutics and Biopharmaceutics*, 50(1), 27–46. [https://doi.org/10.1016/s0939-6411\(00\)00090-4](https://doi.org/10.1016/s0939-6411(00)00090-4)
- Peppas, N. A., & Khare, A. R.** (1993, July). Preparation, structure and diffusional behavior of hydrogels in controlled release. *Advanced Drug Delivery Reviews*, 11(1–2), 1–35. [https://doi.org/10.1016/0169-409x\(93\)90025-y](https://doi.org/10.1016/0169-409x(93)90025-y)

- Peppas, N., Hilt, J., Khademhosseini, A., & Langer, R.** (2006, May 29). Hydrogels in Biology and Medicine: From Molecular Principles to Bionanotechnology. *Advanced Materials*, 18(11), 1345–1360. <https://doi.org/10.1002/adma.200501612>
- Pourjavadi, A., Kurdtabar, M., Mahdavinia, G. R., & Hosseinzadeh, H.** (2006, July 27). Synthesis and super-swelling behavior of a novel protein-based superabsorbent hydrogel. *Polymer Bulletin*, 57(6), 813–824. <https://doi.org/10.1007/s00289-006-0649-5>
- Ribeiro, M. P., Espiga, A., Silva, D., Baptista, P., Henriques, J., Ferreira, C., Silva, J. C., Borges, J. P., Pires, E., Chaves, P., & Correia, I. J.** (2009, November). Development of a new chitosan hydrogel for wound dressing. *Wound Repair and Regeneration*, 17(6), 817–824. <https://doi.org/10.1111/j.1524-475x.2009.00538.x>
- Rufino, E., & Monteiro, E.** (2003). Infrared study on methyl methacrylate–methacrylic acid copolymers and their sodium salts. *Polymer*, 44(23), 7189–7198. <https://doi.org/10.1016/j.polymer.2003.08.041>
- Sadik, W. a. A., El-Demerdash, A. G. M., Abbas, R., & Gabre, H. A.** (2019). Fast synthesis of an eco-friendly starch-grafted poly(N,N-dimethyl acrylamide) hydrogel for the removal of Acid Red 8 dye from aqueous solutions. *Polymer Bulletin*, 77(8), 4445–4468. <https://doi.org/10.1007/s00289-019-02958-x>
- Saunders, L., & Ma, P. X.** (2018, November 22). Self-Healing Supramolecular Hydrogels for Tissue Engineering Applications. *Macromolecular Bioscience*, 19(1). <https://doi.org/10.1002/mabi.201800313>
- Shiblee, M. N. I., Ahmed, K., Khosla, A., Kawakami, M., & Furukawa, H.** (2018). 3D printing of shape memory hydrogels with tunable mechanical properties. *Soft Matter*, 14(38), 7809–7817. <https://doi.org/10.1039/c8sm01156g>
- Shriky, B., Kelly, A., Isreb, M., Babenko, M., Mahmoudi, N., Rogers, S., Shebanova, O., Snow, T., & Gough, T.** (2020, April). Pluronic F127 thermosensitive injectable smart hydrogels for controlled drug delivery system development. *Journal of Colloid and Interface Science*, 565, 119–130. <https://doi.org/10.1016/j.jcis.2019.12.096>
- Sood, A. K., Ohdar, R., & Mahapatra, S.** (2010, January). Parametric appraisal of mechanical property of fused deposition modelling processed parts. *Materials & Design*, 31(1), 287–295. <https://doi.org/10.1016/j.matdes.2009.06.016>
- Standard Terminology for Additive Manufacturing Technologies.** (n.d.). <https://www.astm.org/f2792-12.html>
- Stroganov, V., Zakharchenko, S., Sperling, E., Meyer, A. K., Schmidt, O. G., & Ionov, L.** (2014, April 7). Biodegradable Self-Folding Polymer Films with Controlled Thermo-Triggered Folding. *Advanced Functional Materials*, 24(27), 4357–4363. <https://doi.org/10.1002/adfm.201400176>

- Taormina, G., Sciancalepore, C., Messori, M., & Bondioli, F.** (2018, April 2). 3D printing processes for photocurable polymeric materials: technologies, materials, and future trends. *Journal of Applied Biomaterials & Functional Materials*, 16(3), 151–160. <https://doi.org/10.1177/2280800018764770>
- Tibbits, S.** (2014, January). 4D Printing: Multi-Material Shape Change. *Architectural Design*, 84(1), 116–121. <https://doi.org/10.1002/ad.1710>
- Tiwari, A., & Kobayashi, H.** (2013, September 30). *Responsive Materials and Methods*. John Wiley & Sons. ISBN: 978-1-118-84290-4
- Trenfield, S. J., Awad, A., Madla, C. M., Hatton, G. B., Firth, J., Goyanes, A., Gaisford, S., & Basit, A. W.** (2019, September 3). Shaping the future: recent advances of 3D printing in drug delivery and healthcare. *Expert Opinion on Drug Delivery*, 16(10), 1081–1094. <https://doi.org/10.1080/17425247.2019.1660318>
- Tuncaboylu, D. C., Argun, A., Sahin, M., Sari, M., & Okay, O.** (2012, November). Structure optimization of self-healing hydrogels formed via hydrophobic interactions. *Polymer*, 53(24), 5513–5522. <https://doi.org/10.1016/j.polymer.2012.10.015>
- Ullah, F., Othman, M. B. H., Javed, F., Ahmad, Z., & Akil, H. M.** (2015, December). Classification, processing and application of hydrogels: A review. *Materials Science and Engineering: C*, 57, 414–433. <https://doi.org/10.1016/j.msec.2015.07.053>
- Vega, J. F., Ramos, J., Souza-Egipsy, V., & Martínez-Salazar, J.** (2022, September). On the melting behaviour of linear polyethylene single crystals in mixtures with homogeneous short chain branched polyolefins. *Polymer*, 256, 125256. <https://doi.org/10.1016/j.polymer.2022.125256>
- Velu, R., Jayashankar, D. K., & Subburaj, K.** (2021). Additive processing of biopolymers for medical applications. *Additive Manufacturing*, 635–659. <https://doi.org/10.1016/b978-0-12-818411-0.00019-7>
- Walther, A.** (2019, November 25). Viewpoint: From Responsive to Adaptive and Interactive Materials and Materials Systems: A Roadmap. *Advanced Materials*, 32(20). <https://doi.org/10.1002/adma.201905111>
- Wang, M., Zhou, L., Deng, W., Hou, Y., He, W., Yu, L., Sun, H., Ren, L., & Hou, X.** (2022, January 18). Ultrafast Response and Programmable Locomotion of Liquid/Vapor/Light-Driven Soft Multifunctional Actuators. *ACS Nano*, 16(2), 2672–2681. <https://doi.org/10.1021/acsnano.1c09477>
- Wang, X., Jiang, M., Zhou, Z., Gou, J., & Hui, D.** (2017, February). 3D printing of polymer matrix composites: A review and prospective. *Composites Part B: Engineering*, 110, 442–458. <https://doi.org/10.1016/j.compositesb.2016.11.034>
- Wichterle, O., & Lím, D.** (1960, January). Hydrophilic Gels for Biological Use. *Nature*, 185(4706), 117–118. <https://doi.org/10.1038/185117a0>

- Xia, N. N., Xiong, X. M., Rong, M. Z., Zhang, M. Q., & Kong, F.** (2017, October 16). Self-Healing of Polymer in Acidic Water toward Strength Restoration through the Synergistic Effect of Hydrophilic and Hydrophobic Interactions. *ACS Applied Materials & Interfaces*, 9(42), 37300–37309. <https://doi.org/10.1021/acsami.7b11230>
- Young, R. J., & Lovell, P. A.** (2011, June 27). *Introduction to Polymers*. <https://doi.org/10.1201/9781439894156>
- Zhang, X. N., Wang, Y. J., Sun, S., Hou, L., Wu, P., Wu, Z. L., & Zheng, Q.** (2018, October 4). A Tough and Stiff Hydrogel with Tunable Water Content and Mechanical Properties Based on the Synergistic Effect of Hydrogen Bonding and Hydrophobic Interaction. *Macromolecules*, 51(20), 8136–8146. <https://doi.org/10.1021/acs.macromol.8b01496>
- Zhang, X. Z., Xu, X. D., Cheng, S. X., & Zhuo, R. X.** (2008). Strategies to improve the response rate of thermosensitive PNIPAAm hydrogels. *Soft Matter*, 4(3), 385. <https://doi.org/10.1039/b713803m>
- Zhao, Y., Lei, M., Liu, S. X., & Zhao, Q.** (2018, May). Smart hydrogel-based optical fiber SPR sensor for pH measurements. *Sensors and Actuators B: Chemical*, 261, 226–232. <https://doi.org/10.1016/j.snb.2018.01.120>
- Zheng, Q., Xu, C., Jiang, Z., Zhu, M., Chen, C., & Fu, F.** (2021, May 31). Smart Actuators Based on External Stimulus Response. *Frontiers in Chemistry*, 9. <https://doi.org/10.3389/fchem.2021.650358>

APPENDICES

APPENDIX A: Tables



APPENDIX A:

Table A.1 : Preparation conditions of hydrogels at various mole ratios of DMAA to MAA. C16A= 40 mol %.

x_{DMAA}	DMAA (g)	MAA (g)	TPO (mg)
0.00	0	5.52	364
0.10	0.63	4.94	363
0.20	1.26	4.36	363
0.33	2.08	3.61	362
0.40	2.48	3.24	362
0.50	3.09	2.68	361
0.67	4.08	1.77	361
0.80	4.86	1.06	360
0.90	5.44	0.53	360
1.00	6	0	360

Table A.2 : Preparation conditions of hydrogels at various C16A contents. $x_{DMAA}=0.5$.

C16A (g)	DMAA (g)	MAA (g)	TPO (mg)
5	7.52	6.53	381
8.28	5.53	4.81	372
10.6	4.13	3.59	366
12.3	3.09	2.68	361
13.6	2.28	1.98	358
14.7	1.64	1.42	355
15.6	1.11	0.97	352

CURRICULUM VITAE

Name Surname : Gamze AYDIN

EDUCATION

- **B.Sc.** : 2021, Istanbul Technical University, Faculty of Science and Letters, Chemistry Department

PUBLICATIONS AND PRESENTATIONS ON THE THESIS:

- **Aydın G.**, Abdullah T., Okay O. 2023. 4D Printing of Body Temperature Responsive Bio-Compatible Material Design. *19th Asian Chemical Congress*, July 9-14, 2023 Istanbul, Türkiye.
- **Aydın G.**, Abdullah T., Okay O. 2023. 4D Baskı Uygulamalarına Yönelik Heksadesil Akrilat Tabanlı Işıklıla Kürlenene Reçineler ve Vücut Isısına Tepki Veren, Biçim Hafızalı ve Kendini Onaran Hidrojeller. International Patent Application Number: PCT/TR2024/050687

